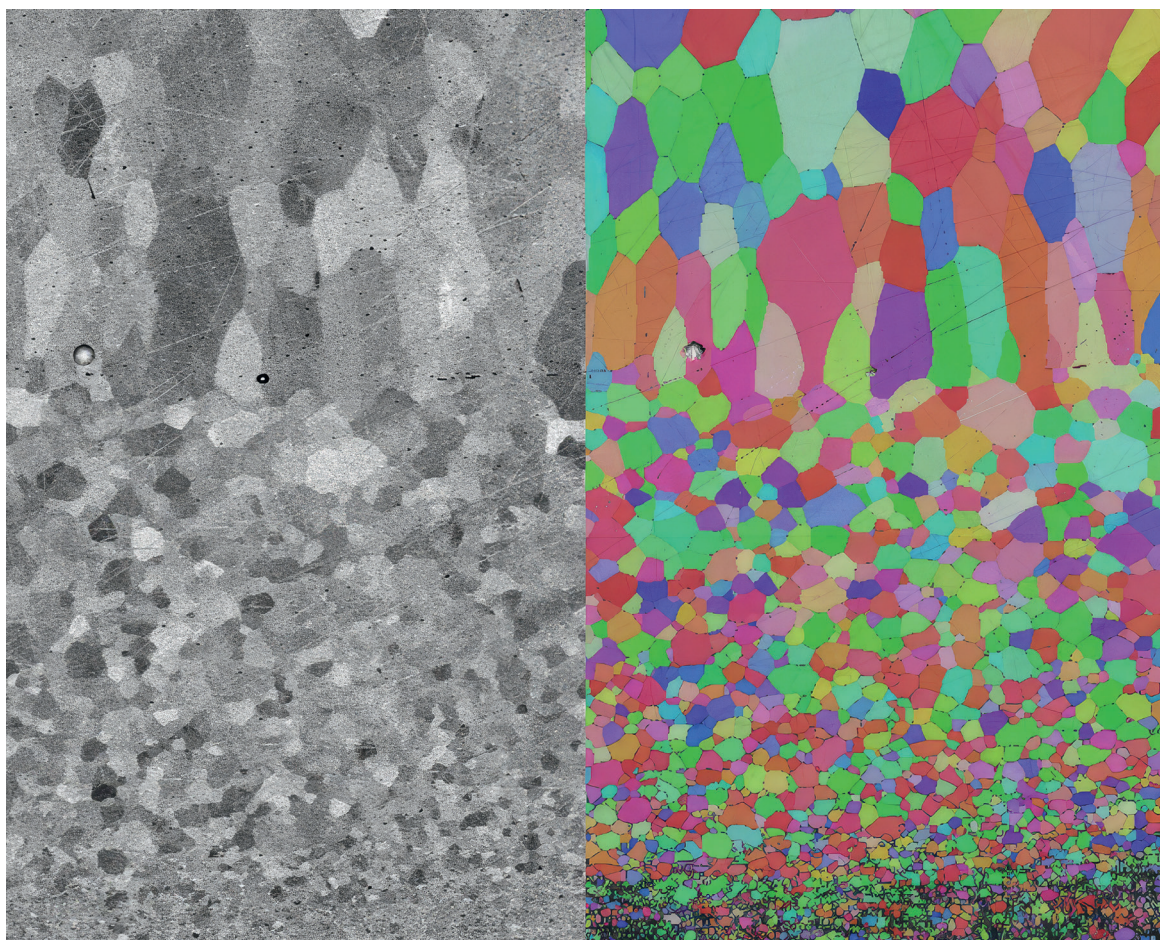




Surface Regeneration of Plasma-Facing Components by Wire-Based Laser Metal Deposition

Jannik Tweer

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List of publications

Parts of the results presented in this work have already been published in the following articles:

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List of symbols and abbreviations

Symbols

α	Plasma incidence angle.....[°]
α_{td}	Thermal diffusivity.....[m ² s ⁻¹]
α_{te}	Thermal expansion coefficient.....[K ⁻¹]
$\dot{m}$	Mass flow rate.....[kg h ⁻¹]
$\dot{V}$	Coolant flow.....[m ³ s ⁻¹]
ϵ	Emissivity.....[-]
η_{arc}	Arc efficiency.....[-]
γ	Surface tension.....[N m ⁻¹]
κ	Thermal conductivity.....[W m ⁻¹ K ⁻¹]
λ	Wavelength.....[m]
μ_d	Dynamic viscosity.....[kg m ⁻¹ s ⁻¹]
μ_k	Kinematic viscosity.....[m ² s ⁻¹]
ν	Poisson's ratio.....[-]
Φ	Radiant flux.....[W]
ϕ	Plasma incidence direction.....[°]
ρ	Density.....[kg m ⁻³]
σ_{SB}	Stefan-Boltzmann constant (5.670373 · 10 ⁻⁸).....[W m ⁻² K ⁻⁴]
σ_S	Standard deviation.....[-]
σ_{therm}	Thermal stress.....[Pa]
σ_{UTS}	Ultimate tensile strength.....[Pa]

LIST OF SYMBOLS AND ABBREVIATIONS

τ_E	Particle confinement time [s]
a	Absorption coefficient [-]
c_p	Specific heat capacity [J kg ⁻¹ K ⁻¹]
A	Area [m ²]
B	Magnetic field [T]
c_3	Wien's constant is ($c_3 = 2.8977685 \cdot 10^{-3}$) [m K]
c	Speed of light [m s ⁻¹]
$d_{ }$	Spot diameter parallel to the deposition direction [m]
D_h	Hydraulic diameter [m]
D_l	Powder layer height [m]
d_R	Maximum resolution [m]
d_s	Spot diameter [m]
d_v	Mean diagonal of the imprint [m]
d_z	Initial position on the z-axis [m]
E_0^*	Normalized equivalent energy density [J mm ⁻³]
E_a	Projected area energy density [J mm ⁻²]
E_M	Elastic modulus [Pa]
E_n	Energy density [J mm ⁻²]
E_t	Line energy input [J mm ⁻¹]
E_v	Volumetric Energy Density [J mm ⁻³]
E	Energy [J]
F_{HF}	Heat flux factor [MW s ^{0.5} m ⁻²]
F	Force [N]
h_{cv}	Convective heat transfer coefficient [W m ⁻² K ⁻¹]
h_d	Hatch distance [m]
H_m	Melt pool height [m]

I_a	Arc current [A]
I_P	Plasma current [MA]
I	Intensity [W m ⁻²]
j	Plasma current density [A m ²]
k_B	Boltzmann constant ($1.38054 \cdot 10^{-23}$) [J K ⁻¹]
k_d	Specific decay constant [s ⁻¹]
K	Ratio between wire feed and scanning velocity [-]
l	Length [m]
$M_{\lambda s}$	Spectral specific radiation [W m ⁻²]
M_{ts}	Radiant exitance [W m ⁻²]
m	Mass [kg]
n_G	Greenwald plasma density limit [m ⁻³]
n_r	Refractive index [-]
n	Particle density [m ⁻³]
Nu	Nusselt number [-]
p_f	Fluid pressure [Pa]
p_P	Plasma pressure [Pa]
p_w	Wetted perimeter [m]
P	Power [W]
Pr	Prantl number [-]
Q_λ	Wavelength dependent radiated power [W]
Q_g	Energy gain factor [-]
q_s	Safety factor [-]
Q	Radiation Energy [J]
q	Rate of heat flux/convection/radiation/internal heat [W]
R_a	Arithmetic average height [m]

LIST OF SYMBOLS AND ABBREVIATIONS

R_q	Root mean square roughness	[m]
r_{wire}	Wire radius	[m]
r_W	Reflection coefficient	[m]
R	Major radius	[m]
r	Minor radius	[m]
Re	Reynolds number	[-]
T_f	Fluid temperature	[°C]
T_v	Vaporization temperature	[°C]
T	Temperature	[°C]
t	Time	[s]
U_a	Arc voltage	[V]
v_f	Fluid velocity	[m s ⁻¹]
v_s	Scanning velocity	[m s ⁻¹]
v_w	Wire feed velocity	[m s ⁻¹]
W_m	Melt pool width	[m]
W_t	Surface waviness	[m]
Z_{eff}	Effective nuclear charge	[C]

Abbreviations

A	Atomic mass number
AM	Additive Manufacturing
Ar	Argon
AT	Articulated Transporter
AUG	ASDEX Upgrade (tokamak)
B	Nuclear binding energy
Be	Beryllium
BRHS	Blanket Remote Handling System

BSE	Backscattered Electron
C	Carbon
CAD	Computer-Aided Design
CFC	Carbon Fibre Composite
CMM	Cassette Multifunctional Movers
CNC	Computerized Numerical Control
Cr	Chromium
CTM	Cassette Toroidal Movers
Cu	Copper
CW	Continuous Wave
D	Deuterium
DED	Direct Energy Deposition
DEMO	DEMONstration Power Plant
dpa	Displacements per atom
DW	Double forged (pure) tungsten
EBAM	Electron Beam Additive Manufacturing
EBSD	Electron Backscatter Diffraction
EDM	Electrical Discharge Machining
ELM	Edge Localized Mode
ERO2.0	Erosion and RedepOsition (MC PWI and impurity transport code)
ETB	Edge Transport Barrier
FEM	Finite Element Method
FPP	Fusion Power Plant
FW	First Wall
H	Hydrogen
HAZ	Heat Affected Zone

LIST OF SYMBOLS AND ABBREVIATIONS

HCPB	Helium Cooled Pebble Bed
He	Helium
HHF	High Heat Flux
HV	Hardness Vickers
IFMIF	International Fusion Materials Irradiation Facility
IGP	ITER Grade Plansee
IPF	Inverse Pole Figure
IR	Infrared
ITER	International Thermonuclear Experimental Reactor (tokamak)
JET	Joint European Torus (tokamak)
JUDITH	Jülich Divertor Test Facility in Hot Cells
LCFS	Last Closed Flux Surface
Li	Lithium
LM	Light Microscopy
LMD-p	Powder-based Laser Metal Deposition
LMD-w	Wire-based Laser Metal Deposition
LOF	Lack Of Fusion Defects
ME	Mounting Equipment
MHD	Magnetohydrodynamics
Mn	Manganese
MPD	Multipurpose Deployer
N	Nitrogen
NA	Numerical Aperture
Ni	Nickel
OFHC	Oxygen Free High Conductivity
PBF-EB/M	Electron Beam Based Powder Bed Fusion

PBF-LB/M	Laser-Based Powder Bed Fusion
PFC	Plasma-Facing Component
PFM	Plasma-Facing Material
PFU	Plasma-Facing Unit
PWI	Plasma-Wall Interaction
RAFM	Reduced Activation Ferritic/Martensitic steel
RH	Remote Handling
RM	Remote Maintenance
RMP	Resonant Magnetic Perturbation
RT	Room Temperature
SE	Secondary Electron
SEM	Scanning Electron Microscopy
SHL	Sheet Lamination
SOL	Scrape-Off Layer
SOLPS	Scrape-Off Layer Plasma Simulation
SXD	Super X Divertor
T	Tritium
Ta	Tantalum
TBR	Tritium Breeding Ratio
TCV	Tokamak à Configuration Variable (tokamak)
TRL	Technology Readiness Level
UTS	Ultimate Tensile Strength
V	Vanadium
VDE	Vertical Displacement Event
W	Tungsten
W7-X	Wendelstein 7-X (stellarator)

LIST OF SYMBOLS AND ABBREVIATIONS

WAAM	Wire Arc Additive Manufacturing
WCLL	Water Cooled Lead Lithium
Z	Nuclear charge number
Zr	Zirconium

Abstract

During operation of high-power nuclear fusion reactors plasma-facing components (PFCs) are exposed to challenging conditions. Due to its properties, pure tungsten is currently the preferred candidate for the plasma-facing material (PFM). However, even for tungsten-armored PFCs the service life remains limited due to surface erosion, neutron damage and overall degradation. The current approach is to extend the service life of PFCs through more durable designs or to replace them once they reach their limit. Since the replacement is both time-consuming and costly, while also severely limiting the availability of the fusion reactor, alternative solutions have to be explored.

In situ local deposition of tungsten on the armor of PFCs using additive manufacturing (AM) could compensate for erosion losses and thus increase the service life of these components. Wire-based laser metal deposition (LMD-w) was evaluated as the most suitable AM process for this task. Following the successful development of a suitable process for the deposition of tungsten coatings on tungsten substrate, the resilience of these coatings was tested under fusion relevant thermal loads. Test components armored with tungsten and coated with LMD-w tungsten were exposed to various thermal loading scenarios in the electron beam facility JUDITH 2. In these thermal loading studies, the LMD-w coatings demonstrated high resilience and excellent heat transfer to the underlying material. Since the coatings exhibit slight manufacturing-related waviness, the influence of wavy topologies on erosion and heat fluxes in a possible DEMO fusion reactor was simulated. Furthermore, it was demonstrated that the waviness can be significantly reduced by remelting the surface via laser. Simulations and experiments were conducted to answer the question of the extent to which the damage expected after the fusion operation can be repaired. It was shown that LMD-w is in principle capable of healing cracks with depths, as they are known to occur in tungsten after edge localized modes (ELMs). In summary, it was proven in the context of this work that erosion losses of the PFM can be successfully compensated and damage be healed by LMD-w and that the LMD-w coatings behave similarly to uncoated tungsten under fusion-relevant thermal loads. This work is the proof of principle for a new concept to extend the service life of PFCs by AM based regeneration of their armor, and thereby provides a potential solution to one of the biggest challenges of nuclear fusion technology. Looking ahead, the acquired insights should be utilized to develop an LMD-w processing unit specially adapted for processing tungsten, which enables optimized process results and is adapted to the environmental conditions within a fusion reactor.

Kurzfassung

Während des Betriebs von leistungsstarken Kernfusionsreaktoren sind die Plasmawandkomponenten (PFCs) anspruchsvollen Bedingungen ausgesetzt. Aufgrund seiner Eigenschaften ist reines Wolfram derzeit der bevorzugte Kandidat für das Plasmawandmaterial (PFM). Doch selbst bei Wolfram gepanzerten PFCs ist die Lebensdauer aufgrund von Oberflächenerosion, Neutronenschäden und allgemeiner Degradation begrenzt. Gegenwärtig werden Ansätze verfolgt, die Lebensdauer der PFCs durch robustere Auslegung zu verlängern oder diese zu ersetzen, sobald sie ihre Belastungsgrenze erreichen. Da ein solcher Austausch sowohl zeitaufwendig als auch kostspielig ist und zudem die Verfügbarkeit des Fusionsreaktors stark einschränkt, müssen alternative Lösungsansätze erforscht werden. Die lokale in situ Deposition von Wolfram auf die Panzerung von PFCs mittels additiver Fertigung könnte Erosionsverluste kompensieren und somit die Lebensdauer dieser Komponenten erhöhen. Das drahtbasierte Laserauftragschweißen (LMD-w) wurde als das am besten geeignete AM-Verfahren für diese Aufgabe bewertet. Nach der erfolgreichen Entwicklung eines geeigneten Verfahrens zur Deposition von Wolframschichten auf Wolframsubstrat wurde die Belastbarkeit dieser Schichten unter fusionsrelevanten thermischen Belastungen getestet. Auf Wolfram gepanzerte Testkomponenten wurden mittels LMD-w Wolfram Beschichtungen aufgebracht und in der Elektronenstrahlanlage JUDITH 2 verschiedenen thermischen Belastungsszenarien ausgesetzt. In diesen thermischen Belastungsstudien zeigten die LMD-w Beschichtungen hohe Robustheit und hervorragende Wärmeübertragung auf das darunterliegende Material. Da die Beschichtungen herstellungsbedingte leichte Welligkeit aufweisen, wurde der Einfluss welliger Topologien auf Erosion und Wärmeflüsse in einem möglichen DEMO Fusionsreaktor simuliert. Darüber hinaus wurde nachgewiesen, dass die Welligkeit durch erneutes Schmelzen der Oberfläche mittels Laser deutlich reduziert werden kann. Um die Frage zu beantworten, inwieweit die nach dem Betrieb eines Fusionsreaktors zu erwartenden Schäden geheilt werden können, wurden Simulationen und Experimente durchgeführt. Es konnte nachgewiesen werden, dass LMD-w grundsätzlich in der Lage ist, Risse mit Tiefen zu heilen, wie sie bekanntermaßen nach der Belastung durch edge localized modes (ELMs) in Wolfram auftreten. Zusammenfassend wurde im Rahmen dieser Arbeit nachgewiesen, dass Erosionsverluste des PFMs durch LMD-w erfolgreich kompensiert und Schäden geheilt werden können, sowie dass sich die LMD-w Beschichtungen bei fusionsrelevanten thermischen Belastungen ähnlich wie unbeschichtetes Wolfram verhalten. Diese Arbeit ist der Proof of Principle für ein neues Konzept zur Verlängerung der Lebensdauer von PFCs durch AM basierte Regeneration ihrer Panzerung und bietet somit eine potenzielle Lösung für eine der größten Herausforderungen der Kernfusionstechnologie. Mit Blick auf die Zukunft sollten die gewonnenen Erkenntnisse genutzt werden, um eine LMD-w Prozesseinheit zu entwickeln die an die Umgebungsbedingungen innerhalb eines Fusionsreaktors angepasst und speziell auf die Verarbeitung von Wolfram ausgelegt ist, um optimierte Prozessergebnisse zu ermöglichen.

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1 Introduction

1.1 Motivation

The rising global energy consumption is reaching a projected energy demand of ~ 10 TW by the second half of the twenty-first century [1]. Covering this extensive demand in a responsible way poses one of the most challenging tasks of our time. Currently, 61 % of global electricity comes from fossil fuels and only 39 % from alternative energy conversion technologies such as hydro energy, nuclear fission, wind and solar [2]. The limited availability of fossil fuels, as well as the resulting environmental pollution and its consequences, are serious problems for the near future. Besides the fact that there are superior applications for fossil fuels than combustion.

Renewable energy, nuclear fission and fusion are the only alternative energy conversion technologies to burning fossil fuels. Nevertheless, each of these alternatives faces its own problems. Renewable energies have a poor needed area to gained energy ratio with energy not necessarily being generated in the location and time it is required. In contrast, nuclear fission possesses a good area needed to energy gained ratio. The biggest issue with nuclear fission, apart from the risk of accidents with devastating consequences, lies in the management of the radioactively activated waste. The third and least developed option is nuclear fusion, which is also the most promising of the alternatives [3].

The next step subsequent to the large experimental fusion reactors such as the Joint European Torus (JET) and ITER ('the way' in latin) will be the construction of a fusion power plant (FPP) [4, 5]. In this context, the European endeavor is the DEMOnstration (DEMO) FPP [6]. One of the greatest challenges in fusion research is the deterioration of the plasma-facing materials as the operating time of the fusion reactor progresses. Due to the demanding conditions that PFCs in high-power fusion reactors are facing, only a few elements and materials can be considered as plasma-facing materials (PFMs). The behavior of various materials for application within the vacuum vessel under the stresses of fusion reactor operation is currently being researched. In this context, research is being carried out to develop increasingly durable materials and components. Tungsten is characterized by a high melting point, high thermal conductivity, comparatively low erosion rate under plasma expo-

sure and low tritium inventory [7–9]. Taking all the requirements for the PFM into account, tungsten stands out as the most promising candidate [10]. Consequently, it is planned to use tungsten as PFM in the next generation of high-power fusion reactors such as ITER and the european DEMO FPP [7, 11–17]. However, over time, the plasma-facing components (PFC) are damaged and erode due to high particle fluxes, high heat fluxes and bombardment by high-energy neutrons [9]. The current strategy is to replace damaged plasma-facing components in a time-consuming procedure [18, 19]. This enormously limits the operating time of fusion reactors and thus reduces the competitiveness of this technology in direct comparison with alternative energy conversion technologies. Another strategy is the in situ repair of damaged PFCs, which so far has not been comprehensively investigated. This work explores the viability of in situ regeneration of tungsten armored PFCs.

1.2 Scope of work

Due to the demanding conditions that PFCs in high-power fusion reactors are facing, only a few elements and materials can be considered as PFMs. Due to its properties, tungsten is currently considered to be the most suitable candidate for PFM. But even with a material as resilient as tungsten, the erosion of PFM cannot be avoided entirely [20, 21]. As a result, PFCs can reach the end of their service life due to advanced erosion. The service life of these PFCs could be extended by compensation of erosion losses through repair using additive manufacturing (AM) technology. If such repairs were carried out in situ, the time-consuming and costly removal of the PFCs could be avoided. Direct energy deposition (DED) processes, which fall under the AM process category, are suitable for this kind of in situ repair [22]. However, several necessary conditions must be considered in the selection of a suitable manufacturing process. Contamination of the reactor vessel with high Z elements such as tungsten must be prevented at all costs. Only low amounts of tungsten are permitted in the plasma core ($< \text{a few } 10^{-5}$), as the plasma would otherwise lose too much energy through radiation cooling and cool down excessively [10]. Therefore, all powder-based processes are considered unsuitable for in situ operation as they have deposition rates of $\sim 50\%$ [22]. In this context, wire-based processes with deposition rates of nearly 100% are suitable [23]. In addition, the deposition system must be designed to be sufficiently space-saving to be inserted through the ports into the reactor cavity and to operate adequately even in more narrow spaces, like the divertor region. The deposition unit must also be sufficiently lightweight to not exceed the maximum load capacity of the robotic arm and to minimize undesirable oscillations of the robotic arm. Even during reactor downtime, electronics inside the reactor vacuum vessel would suffer from damaging radiation due to nuclear decay of the inner wall materials. Thus, these electronics either need to be shielded from radiation or located outside the reactor hot cell. In contrast to electron beam-based processes, in laser-based processes the large laser sources as well as the majority of

electronics can be stationed outside the vessel and the light can be guided to the operation location via glass fibers. For these reasons, the wire-based laser metal deposition process (LMD-w) was chosen as the manufacturing process for the in-situ repair.

This work is intended to provide a proof of principle for the regeneration of the surface of tungsten armored PFCs by LMD-w. This involves both the development of an LMD-w process for tungsten as well as the performance and suitability of these LMD-w coatings for the specific application as PFM of nuclear fusion reactors.

Since the deposition of tungsten on tungsten substrate by LMD-w has not previously been researched, the first step was to investigate the adaptation of the LMD-w process to tungsten for plausibility. Subsequently, the aim was to find suitable process parameters for this tungsten process. The process development was advanced from individual welding beads to entire layers. By deposition of several welding beads next to each other, the layers were generated.

In addition, stacking of these layers to achieve the necessary layer thickness and the post-treatment of layers by laser remelting were researched. The intention behind laser remelting post treatment is the reduction the macroscopic surface waviness after LMD-w coating to prevent leading edges in the subsequent fusion reactor plasma exposure. The influence of macroscopic surface waviness of PFM on heat fluxes, erosion and deposition during operation of the future DEMO FPP were simulated using the ERO2.0 Monte-Carlo code. After the development of an LMD-w tungsten process, the material deposited in this way and the heat-affected zone in the substrate were characterized.

Studies were conducted, to determine how the deposited tungsten layers perform under steady state as well as combined steady state and transient heat fluxes. To achieve this, the tungsten armor of test PFCs was coated using LMD-w and these coated components were then thermally stressed in the electron beam facility JUDITH 2.

Moreover, it was investigated whether PFM damaged by fusion-relevant thermal loads can be healed by application of LMD-w coatings. Tungsten tiles were brazed onto a test component and subjected to combined steady state and transient thermal stresses in JUDITH 2 to generate fusion-relevant damage. Afterwards, it was tested whether the LMD-w and laser remelting process are able to heal these damages. As part of this, the temperature profile during the laser process was simulated using the finite element method (FEM).

1.3 Structure of the work

Theoretical Background

The theoretical background of thermonuclear fusion technology is covered broadly in chapter 2. It also covers the relevant high-power experimental reactors and introduces the European DEMO FPP project. Furthermore, the fundamentals of plasma-wall interaction (PWI) are introduced. In this context, the main challenges of PFCs and their materials are discussed. An overview of the remote maintenance systems for nuclear fusion reactors is also provided.

Chapter 3 provides the theoretical background for AM technology and outlines its potential for fusion technology. Furthermore, relevant studies already carried out on AM in the fusion context are discussed. Finally, an overview of the DED processes is provided and the LMD-w process is introduced. Here, the characteristics of the LMD-w process for lateral wire feeding are discussed.

Test facilities and methods

This is followed by an overview of the various test facilities and utilized methods in chapter 4. The LMD-w facility and the associated diode laser used for material deposition are characterized. The electron beam facility JUDITH 2 which was used for the performed high heat flux (HHF) experiments is described. In addition, temperature measurements using pyrometers and infrared thermography are discussed, which were utilized to monitor the LMD-w process and HHF exposures. The analysis methods used for sample characterization such as scanning electron microscopy (SEM), electron backscatter diffraction (EBSD), profile measurements and hardness measurements are outlined. Finally, the finite element method (FEM) simulations, which were used to simulate temperature profiles, and the Monte-Carlo code ERO2.0, which was used to analyze the influence of the macroscopic wavy surface topologies of LMD-w coated PFCs on heat fluxes, erosion and redeposition in fusion reactors, are discussed.

Adaptation of the LMD-w process for tungsten

The development of an adapted LMD-w process for the deposition of tungsten on tungsten substrate is covered in detail for the respective system setup in chapter 5. First, the process was optimized for the deposition of individual welding beads and subsequently the deposition of layers. These layers were created by positioning several welding beads next to each other. Then stacking of layers and the reduction of the surface waviness by subsequent laser remelting was investigated. The set of process parameters that achieved the most favorable results with the utilized facility setup were provided.

ERO2.0 simulations were performed to investigate the influence of wavy topologies of PFM on heat fluxes, erosion and redeposition under DEMO relevant plasma exposure. It was found that total heat flux, erosion, and deposition values are not

influenced by surface waviness or plasma incidence direction. However, it also turned out that wavy topologies of PFC armor can influence the distribution of heat fluxes, erosion, and deposition, and thus their peak values during plasma exposure. Surface smoothing of wavy topologies can reduce the local concentration of heat flux and erosion, and choosing the deposition direction parallel to the plasma incidence direction can prevent it entirely.

High heat flux testing of coated plasma facing components

Various types of LMD-w coatings were deposited on the surface of tungsten armored test components and exposed to HHFs to examine the performance of these coatings under fusion-relevant thermal loads (see chapter 6). In the course of these experiments, the components were subjected to cyclic steady state thermal loads to simulate the influence of multiple fusion reactor plasma discharges on the LMD-w coatings. The test components were also subjected to combined steady state and transient thermal loads to study the thermal effect of giant edge localized modes (ELMs) and off-normal plasma events on these coatings. The components were thermally monitored during exposure to obtain an insight into their condition during exposure. Furthermore, the samples were examined comprehensively before and after thermal exposure to detect surface modifications and further effects. The LMD-w coated tungsten tiles performed similarly to uncoated references in both the cyclic steady state as well as the combined steady state and transient thermal loading scenarios.

Repair of heat load damaged PFM using LMD-w

For chapter 7, tungsten tiles were damaged by combined steady state and transient thermal loads to obtain a surface state that is expected after edge localized modes (ELMs) have loaded the PFCs. FEM simulations of the laser remelting process were performed to gain insight into the temperature profile in the material during the process. The previously damaged tungsten tiles were subsequently subjected to various scanning strategies of LMD-w and laser remelting. It turned out that both processes are able to heal the damage.

2 Theoretical background of thermonuclear fusion

2.1 Thermonuclear fusion

There are two nuclear reactions, nuclear fission which is used in present nuclear power plants and nuclear fusion. In the nuclear fusion reaction, multiple nuclei combine and form a heavier nucleus. Depending on the binding energy of the nuclei before and after the reaction, energy is either released or absorbed. If the mass of the nuclei is less than that of ^{62}Ni , energy is released during the fusion reaction while energy is absorbed if the nuclei mass is higher. This is due to the fact that ^{62}Ni has the highest binding energy per nucleon (fig. 2.1) [24].

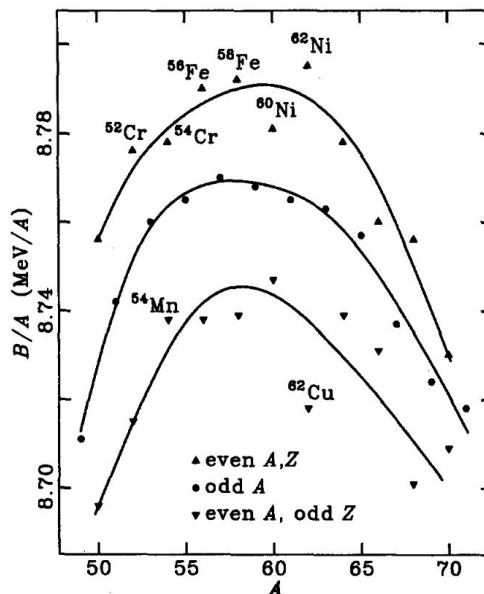


Figure 2.1: Average binding energy B/A per nucleon against its atomic mass number A of the strongest bound nucleons. Z is the nuclear charge number of the atomic nucleus [24].

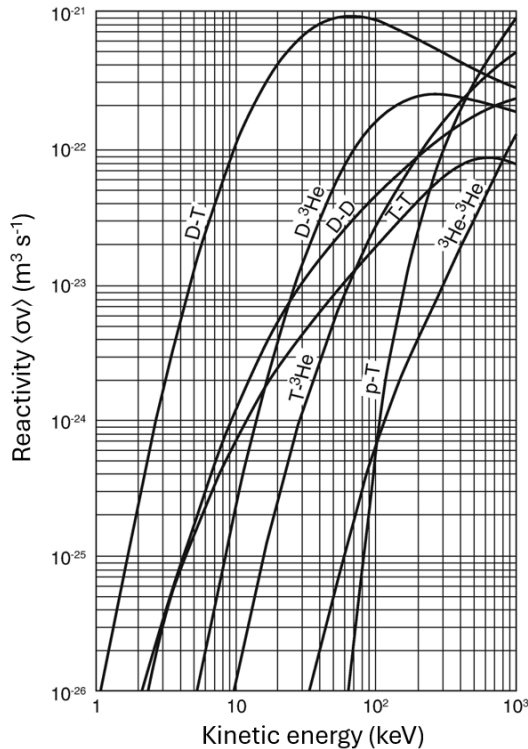


Figure 2.2: The illustration shows the reactivity $\langle\sigma v\rangle$ of high-yield fusion reactions at certain kinetic energies. The deuterium-tritium (DT) reaction features the highest reaction rate at comparatively low temperatures [25].

If we take hydrogen (the three most stable hydrogen isotopes are: Protium ^1H , Deuterium ^2H , Tritium ^3H) as an example: the nucleus of a hydrogen atom consists of only one proton. For two positively charged particles to combine, the coulomb repulsion needs to be overcome. At longer distances the Coulomb force dominates, whereas at the short distance of a few nuclear radii the attractive nuclear force dominates. Even though the quantum tunneling effect facilitates the nuclear reaction, an average kinetic energy of about 10 keV is required to enable fusion in a deuterium-tritium (DT) plasma. As shown in fig. 2.2, the deuterium-tritium fusion reaction requires the lowest kinetic energy to achieve a high reaction rate compared to other important fusion reactions. In the DT reaction, schematically visualized in fig. 2.3, deuterium and tritium combine in a transition state to form ^5He and decay shortly afterwards with a decay width of 0.6 MeV (half life time $\approx 5.5 \cdot 10^{-22}$ s) [26]. The decay releases binding energy of 17.6 MeV, which is distributed on two stable products in the form of a free neutron and a ^4He nucleus according to the inverse of

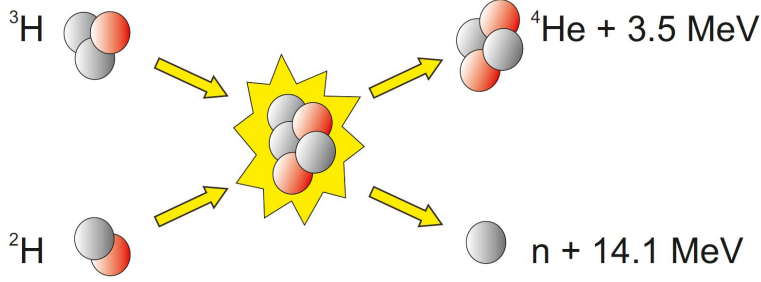
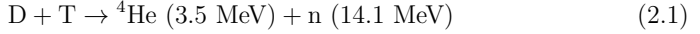
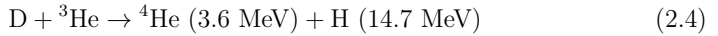
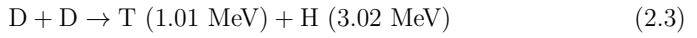
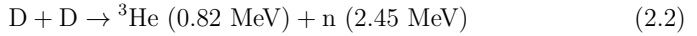


Figure 2.3: Deuterium-tritium fusion: Deuterium and tritium form an intermediate state ${}^5\text{He}$ with a half life time of approximately $5.5 \cdot 10^{-22}$ s [26]. The products of the decay are a stable ${}^4\text{He}$ nucleus and a free neutron with kinetic energies of 3.5 MeV and 14.1 MeV respectively.

their mass ratio:



Listed below are other fusion reactions that can potentially be performed on earth in a controlled manner:



The following equation describes the energy E released during fusion reactions [J], where m is the mass difference [kg] between the combined mass of the reaction partners and the reaction products formed (atomic nuclei and neutrons) and c is the speed of light in vacuum [m s^{-1}] [27]:

$$E = mc^2 [\text{J}] \quad (2.5)$$

Achieving the fusion reactions addressed in fig. 2.2 requires certain conditions that are not easy to achieve. To achieve fusion reactions in particle plasmas, a high triple product $nT\tau_E$ [s keV m^{-3}] is required, where n is the particle density [m^{-3}], T is the kinetic energy [keV] and τ_E is the particle confinement time [s]. The Lawson criterion for achieving ignition combines these parameters and sets a minimum threshold that the triple product must exceed, which is roughly [28–30]:

$$nT\tau_E = \frac{12k_B T^2}{\langle \sigma v \rangle E_\alpha - 4c_1 Z_{eff} (k_B T)^2} \geq 5 \cdot 10^{21} \text{ s keV m}^{-3} \quad (2.6)$$

Here k_B is the Boltzmann constant [J K^{-1}], $\langle \sigma v \rangle$ the reactivity [$\text{m}^3 \text{s}^{-1}$] of the specific fusion reaction (see fig. 2.2), E_α the energy released [J] per fusion reaction (for the DT fusion the energy of the ${}^4\text{He}$ particles) and $4c_1 Z_{eff} (k_B T)^2$ the bremsstrahlung for

an effective nuclear charge Z_{eff} . The aforementioned conditions can be achieved by gravitational, inertial or magnetic confinement. Gravitational confinement is only observed in stars and is achieved by their high mass. In stars, hydrogen is fused into helium, releasing energy. This happens mainly in chain reactions, such as the proton-proton (p-p) and carbon-nitrogen-oxygen cycles [31]. Fusion by gravitational confinement is not an option on Earth compared to magnetic and inertial confinement. The ignition conditions for magnetic confinement fusion are particle densities of $n \approx 2 \cdot 10^{20} \text{ m}^{-3}$, confinement times of $\tau_E \approx 1\text{-}10 \text{ s}$ and a kinetic ion energy of $T \approx 10\text{-}15 \text{ keV}$. Compared to this, the confinement times required for inertial confinement fusion are shorter because the particle densities are higher [32]. However, magnetic confinement is currently considered to be the most promising and most advanced variant for achieving these conditions [29, 33]. Therefore, this work will focus on fusion technology based on magnetic confinement. External heating is required to bring the fuel to the desired kinetic energy and thus achieve ignition in a nuclear fusion reactor. To generate net power in a fusion reactor, the power P_F [W] generated by fusion needs to exceed the power required for the external heating systems P_H [W] (break-even condition):

$$Q_g = \frac{P_F}{P_H} \geq 1 \quad (2.7)$$

Where Q_g is the so-called energy gain factor. With the optimal mixing ratio DT of 50:50 a $Q_g = 0.65$ was achieved in the Joint European Torus (JET) and transiently in a hot ion H-mode regime even a $Q \sim 0.6\text{-}1$ has been measured [34–36]. However, an even higher energy-gain factor is necessary to cover the energy demand of the entire reactor plant, including all systems required for reactor operation. In the long term, the general aim is to strive for a $Q_g = \infty$. This requires the fusion reaction to heat itself through the energetic alpha particles. The internal heating power of the plasma due to collision with the alpha particles P_α [W] accounts for 20 % of the fusion power P_F [W] in the DT reaction (it is assumed that the neutrons leave the plasma without interacting with it). Conduction, i.e. particles leaving the enclosed plasma, and bremsstrahlung are the two effects that have the greatest influence on power loss P_L [W] [30]. For a self-sustaining fusion reaction, the energy delivered by α particles needs to be greater than or equal to those losses. The condition for a self-heating fusion reaction is therefore as follows [37]:

$$P_\alpha \geq P_L[\text{W}] \quad (2.8)$$

In magnetic confinement fusion, one challenge is to protect the plasma from cooling due to loss of its kinetic energy through contact with the cold inner wall of the reactor vessel. This confinement is achieved by manipulating the electrically charged plasma particles through the Lorentz force using specially shaped magnetic fields. The hot inner core of the plasma must be insulated from the cold wall in the most effective way possible. This insulation is achieved by a sophisticated and strong

magnetic confinement. With growing size of a fusion reactor and increasing fusion plasma volume where energy is released, the surface area over which energy is lost rises proportionally slower. Thus, it is desirable to build fusion machines with ever larger plasma volumes and stronger magnetic fields in order to increase the particle confinement time τ_E and the fusion performance. The development towards fusion reactors with ever larger major radii is clearly evident in the comparison illustrated in (fig. 2.4) [38].

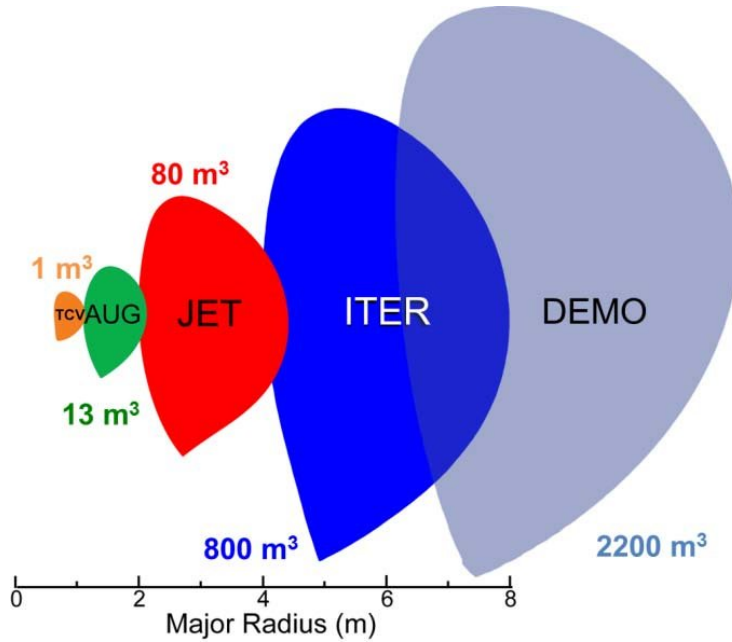
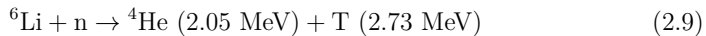


Figure 2.4: The development of the dimensions of relevant fusion reactors in comparison with ITER and DEMO (© [2017] IEEE [38]).

Due to the high reaction rate at comparatively low kinetic energies and the large amount of binding energy released, the DT-reaction is the preferred choice for future thermonuclear fusion reactors (fig. 2.2). The hydrogen isotope and fusion reaction fuel deuterium is contained in seawater at up to 0.02-0.03 at.% and can be easily extracted [39]. Tritium, on the other hand, only occurs sporadically in extremely low concentrations on earth. With a half-life of 12.33 years, tritium decays into ^3He by β -decay [40]. Due to its short half-life time tritium requires artificial production to supply the DT reactor operation. The high energy neutrons emitted in the DT reaction can be used to obtain tritium within the fusion reactor by bombarding a lithium breeding blanket in the inner wall of the reactor and thus generating tritium in a nuclear reaction. The tritium must then be extracted and can be used as fuel

for the fusion reactor. Lithium breeding blankets are currently the most promising candidate, in physical and technical regard, for the production and extraction of tritium in future fusion power plants (FPPs). However, the construction of a resilient, working breeding blanket is by no means trivial due to the extreme conditions within fusion reactors as well as the high diffusion rate of tritium. This poses one of the greatest challenges for fusion technology [14]. Two possible nuclear reactions for breeding tritium are shown below [41, 42]:



The ratio between the number of tritium atoms that are bred T_{breed} and the number of tritium atoms that are consumed by the DT fusion reaction T_{burned} is called TBR and is described by the following equation [43].

$$\text{TBR} = \frac{T_{\text{breed}}}{T_{\text{burned}}} \quad (2.11)$$

For a tritium self-sufficient FPP, the TBR should be ≥ 1.1 [44, 45]. This includes a 10 % safety margin to cover calculation uncertainties in the previously performed simulations for the TBR [46]. This also takes losses of tritium into account. To increase the efficiency of tritium breeding in the blankets, neutron multipliers made of materials such as beryllium are utilized [47].

For several decades, mankind has been researching fusion technology using magnetic confinement. One successful strategy to achieve fusion through magnetic confinement is the use of tokamak devices. The word tokamak is an acronym from Russian and stands for “toroidalnaya kamera ee magnitaya katushka”, or toroidal chamber with magnetic coils [48]. The concept was pioneered in the early 1950s by Soviet scientists Andrei D. Sakharov and Igor J. Tamm [49]. Tokamaks are torodial devices in which strong magnetic fields are generated by magnets in which the plasma is heated, confined and compressed (fig. 2.5). In tokamaks the magnetic confinement is achieved by superimposing a toroidal magnetic field B_T [T] and a poloidal magnetic field B_{pol} [T] creating a toroidal helix [51]. This closed geometry prevents the losses that occur in linear plasma configurations. In the tokamak configuration, energy and particle losses result from transport perpendicular to the magnetic field. The toroidal magnetic field B_T is generated by the toroidal magnetic field coils, while the central solenoid induces a current I_P [A] in the plasma, which generates the main poloidal magnetic field B_{pol} . External poloidal field coils provide additional poloidal magnetic fields for plasma shaping and control. The concept of magnetic confinement can be described as follows. A current of electrically charged particles, with a density j [A m⁻²], interacts with a magnetic field B , resulting in a force $\vec{j} \times \vec{B}$ per unit volume [52]. These forces resulting from $\vec{j} \times \vec{B}$ are determined by the plasma currents I_P , as well as the magnetic fields B . On the other hand, a force acts

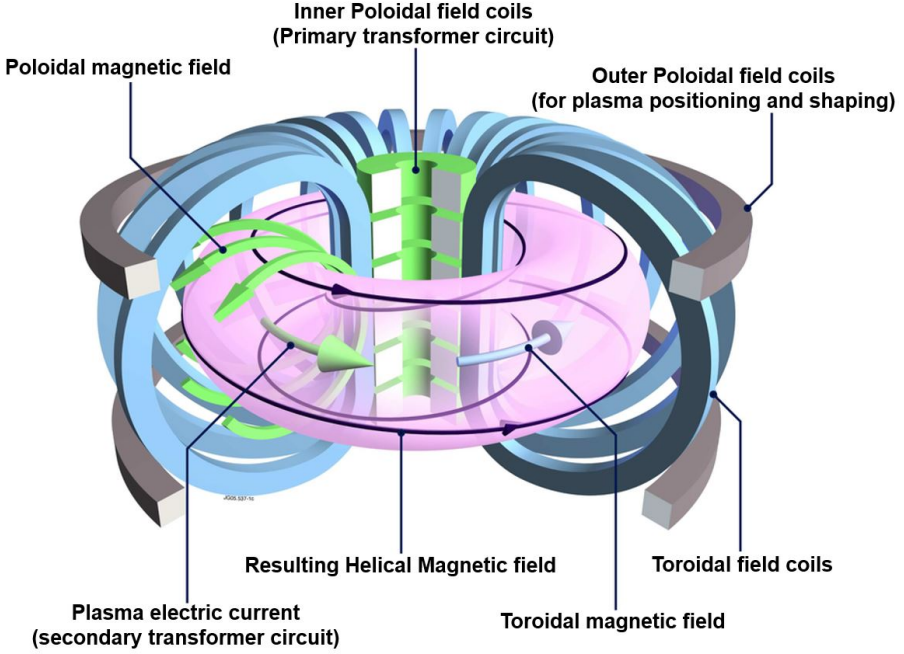


Figure 2.5: The schematic configuration of a tokamak with its toroidal and poloidal magnetic field coils is illustrated. The graphic also shows the electric current induced in the plasma, with the resulting twisted magnetic field lines [50].

that is caused by the gradient of the kinetic plasma pressure ∇p_P . Consequently, for magnetic confinement, it is required that equilibrium between this gradient of the kinetic plasma pressure ∇p_P and the $\vec{j} \times \vec{B}$ forces is achieved and maintained [53]. In tokamaks, the current I_P [A] that is induced in the plasma causes the magnetic field to twist, which stabilizes the plasma and improves the confinement. To induce an electric field or a current in the plasma, a change in the magnetic field is necessary. Since this is only possible to a limited extent, tokamaks are operated in pulses. The helical twist of the magnetic field lines, which is essential for the magnetic confinement, is characterized by the so-called safety factor q_s . The safety factor provides the fraction of the toroidal angle with respect to 2π it takes for a field line achieve the same poloidal position. Alternatively, the safety factor can be described as the ratio between the changes in the local poloidal and toroidal angles along the magnetic field lines. The profile of the safety factor is given in cylindrical approximation for a circular poloidal plasma cross-section in equation 2.12, provided that the minor plasma radius r [m] is small compared to the major plasma radius

R [m]. [53]

$$q_s(r) = \frac{r}{R} \frac{B_T}{B_{pol}} \quad (2.12)$$

For magnetohydrodynamic (MHD) stability at the edge of the plasma, the safety factor should be $q_s \geq 2$ [53]. Controlling the plasma, which consists of ionized particles, using magnetic fields is complicated because an electric current flowing through the plasma generates a magnetic field, which in turn generates another electric current, and so on. Predicting the behavior of plasmas is attempted within the field of MHD. Understanding the plasma behavior is of great importance, because instabilities can cause the plasma to get out of control in a fraction of a second, leading to unwanted collisions of the plasma with the inner wall. Tokamaks have been developed and improved over time due to their relative stability. Tokamaks can be operated in the classic low energy confinement mode (L-mode) or in high-confinement mode (H-mode). Depending on the detailed interpretation, other regimes or operational modes of plasma confinement may exist, but the L- and H-mode are currently the most relevant for large scale fusion devices [54–56]. L-mode is the default mode of operation if not modified by further confinement mechanisms. This L-mode is easy to achieve, but compared to the H-mode, it loses a lot more energy due to a significant level of turbulence. It shows poorer confinement and lower temperatures overall, which makes it more difficult to achieve fusion reactions. In contrast, the H-mode, which can be triggered by external heating or high-power operation, features an edge transport barrier (ETB) that acts as a sharp boundary layer, significantly reducing turbulences at the edge of the plasma, which leads to lower energy and particle losses [57, 58]. However, the H-mode inherently results in MHD instabilities in form of edge localized modes (ELMs), which in turn create their own set of problems (as detailed in section 2.3) [59]. Since high plasma densities can lead to instabilities and disruptions, an empirical scaling law for the upper plasma density limit was developed. This scaling law is known as the Greenwald limit and expressed in equation 2.13 [60].

$$n_G = \frac{I_P}{\pi r^2} \quad (2.13)$$

Here n_G [m⁻³] is the Greenwald plasma density limit, I_P [A] the plasma current and r [m] the minor radius of the specific tokamak. Nevertheless, technological advancements could enable future fusion reactors to exceed the Greenwald limit.

The largest tokamak facilities built to date include the Joint European Torus (JET) in Culham, United Kingdom and the JT-60SA in Naka, Japan [4, 61]. These modern tokamaks like JET and JT-60SA have additional poloidal coils to better stabilize the plasma vertically [62, 63]. The largest tokamak yet, which is still under construction, is the International Thermonuclear Experimental Reactor (ITER) in Cadarache, France, which should achieve a $Q \geq 10$ for the first time and thus lift magnetic fusion technology with the break even to the next level [64]. Among other

aspects, ITER shall be used to study the behavior of a burning plasma heated by a significant fraction from α particles [33, 65]. In addition, tritium breeding blankets are planned to be installed in ITER to get one step closer to a closed tritium fuel cycle [14]. It is planned that this fusion reactor will deliver 500 MW of fusion power at a heating power input of ≤ 50 MW for pulse lengths of ≥ 400 s [66–68].

Another concept for achieving fusion by magnetic confinement is the stellarator, which has also been continuously developed. The configuration of a classical stellarator with toroidal and helical magnetic field coils is illustrated schematically in (fig. 2.6) [69]. The concept of a stellarator was first proposed in 1951 by Lyman

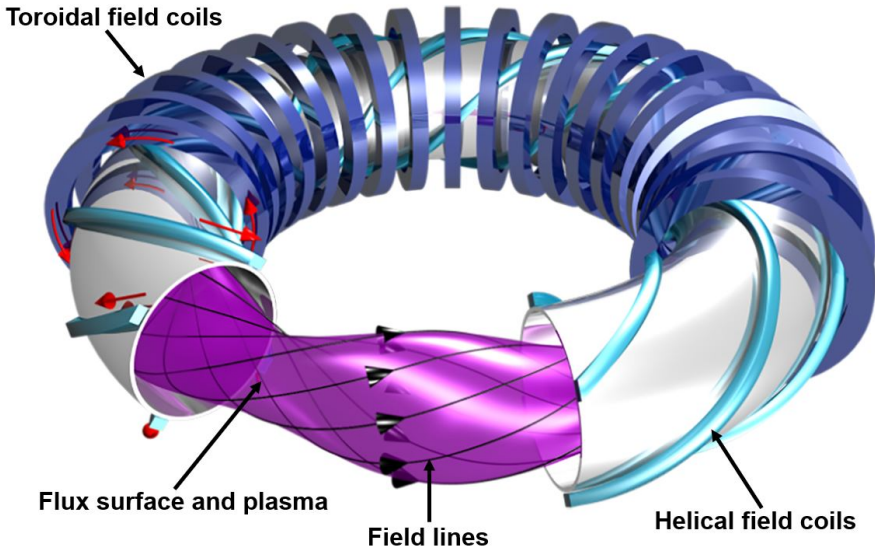


Figure 2.6: The schematic representation of a classic stellarator with its toroidal and helical magnetic field coils with the resulting field lines in the plasma is visualized [69].

Spitzer at Princeton University, USA [51]. The word stellarator comes from the Latin word for star, ‘stella’, and was chosen to conjure up associations with the stars, where nuclear fusion also takes place [70]. In contrast to the tokamak, the stellarator does not require the induction of a current in the plasma to achieve the twisting of the magnetic field and the magnetic confinement. This means that stellarators can theoretically be operated continuously. Furthermore, stellarators as opposed to tokamaks are characterized by the absence of current-driven instabilities [43]. The rotation of the toroidal field in stellarators is achieved by twisting the magnetic axis or by adding a helical field [51]. This helical field is mainly achieved by adding helical magnetic field coils or complex 3D shaped magnetic field coils [43]. However, this comes at the disadvantage that stellarators are more difficult to construct and

build, which has led at the present time to a development gap compared to the tokamak concept. The largest built stellarator to date is Wendelstein 7-X (W7-X) in Greifswald, Germany (with major radius $R = 5.5$ m, minor radius $r = 0.5$ m, and 30 m^3 plasma volume) [71]. In W7-X, the twisted magnetic field is achieved by complex non-planar, modular coils whose geometry has been specifically designed to achieve an optimized magnetic field [72]. This machine is designed to achieve, at $P_H \approx 14$ MW heating power input and pulse lengths of up to 30 minutes, plasma temperatures of $T \approx 5\text{--}11$ keV at plasma densities of $n \approx 3 \cdot 10^{20} \text{ m}^{-3}$ [73].

2.2 Future fusion power plants

Following experimental fusion reactors, the next stage of development towards commercial fusion power plants (FPPs) will be the DEMOnstration power plants, which, in contrast to experimental reactors, will supply energy to the power grid. There are various approaches to the development of FPPs. In the European DEMOnstration power plant project in the context of magnetic fusion, it is currently not absolutely certain whether the tokamak or stellarator design will be used. But due to the developmental lead of the tokamak technology, this approach is currently preferred. Therefore, the European DEMOnstration power plant will most likely be the successor to ITER [74]. Table 2.1 compares the essential data of these machines. In order to achieve a higher fusion power than ITER, which is necessary to achieve the goals of the DEMO project, this reactor should exceed the dimensions of its predecessor. However, there are other ways of achieving this, such as using stronger magnetic fields to compress the plasma further and thus achieve a higher fusion power. Japan, China, Korea, USA, Russia and India are also planning independent efforts after ITER to develop DEMO FPPs [68]. Furthermore, there are currently multiple private companies that are pursuing various concepts for the construction of commercially viable FPPs [75, 76].

In thermonuclear fusion power plants, the energy implanted into the inner wall of the reactor, primarily by neutrons (~ 80 % of the total energy of the DT plasma), is transferred to the coolant and converted into electric energy in turbines [78]. In fig. 2.7, a fusion power plant based on the tokamak concept that utilizes the DT reaction is schematically illustrated. In the depicted fusion power plant, deuterium, tritium and helium are pumped out of the divertor and separated from another before being recycled.

In tritium breeding blankets, neutrons are utilized to obtain tritium from lithium. This tritium then serves as fuel for the DT fusion reaction in the device. These tritium breeding blankets will be stationed in the first wall of a fusion power plant. The TBR (equation 2.11) for a tritium self-sufficient FPP, should be ≥ 1.1 [44, 45].

Table 2.1: This table lists the planned characteristics and performance data for ITER (experimental reactor) and the European DEMO FPP [65, 77].

	ITER	DEMO	Unit
Characteristics			
- Major radius (R)	6.2	9.0	m
- Minor radius (r)	2	2.9	m
- Aspect ratio	3.1	3.1	
- Plasma current (I_P)	15	18.0	MA
- Toroidal field on axis (B_T)	5.3	5.9	T
- Auxiliary heating power - flat top (P_H)	73 (110)	50	MW
Performance			
- Fusion Power (P_F)	500	2000	MW
- Electric Output	none	500	MW
- Neutron wall loading	0.57	1.04	MW m ⁻²
- Burn time	≥ 400	7200	s
- Plasma stored Energy	0.5	1.181	GJ

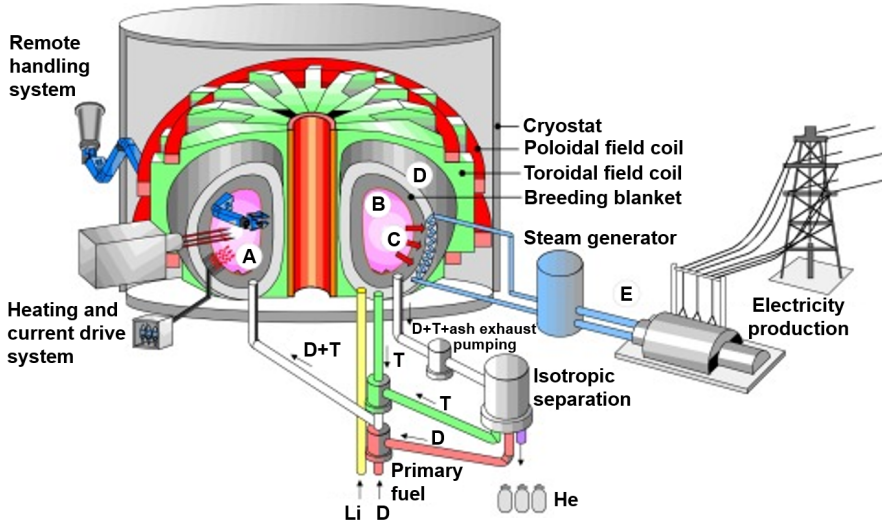


Figure 2.7: Schematic representation of a fusion power plant based on magnetic confinement showing: (A) vacuum chamber, (B) plasma, (C) plasma radiation, (D) tritium breeding blanket, (E) electricity production [78].

2.3 Plasma-wall interaction

With the development towards ever more powerful fusion reactors capable of operation with ever longer burning plasma pulses, the stresses on the plasma facing materials and components are intensifying. The pulse durations in today's tokamak fusion reactors such as JET are mostly tens of seconds, whereas pulse lengths of ≥ 400 s are planned for ITER and as much as 7200 s for European DEMO FPP [4, 65, 77]. The electron, ion and heat fluxes in ITER will only be higher than in earlier fusion reactors by a factor of 2 to 3. But the accumulated particle and energy loads per plasma discharge in ITER will be higher by up to four orders of magnitude. The integrated load on the inner wall of the device caused by a single ITER pulse is comparable to the load applied to the inner wall of JET during a full year of operation [79]. As a result of the longer pulses, the accumulated loads will increase once again in the DEMO FPPs. In plasma-wall interaction, a distinction can be made between heat fluxes and particle fluxes that impinge on the inner wall of fusion reactors.

Thermal loads

The energy deposited on the inner wall of fusion reactors is absorbed by the installed materials. In particular, the plasma-facing components are heated until an equilibrium is reached between the deposition of energy and dissipation by active cooling. In the case of normal operation, this is referred to as steady state heat flux operation. In the area of the first wall of the European DEMO FPP, steady state thermal loads of $\sim 1 \text{ MW m}^{-2}$ are expected, while in the area of the divertor, they can reach up to 20 MW m^{-2} . These steady state thermal loads alone pose enormous challenges for the PFCs and in particular the PFM. In addition, however, there are transient thermal loads that implant enormous amounts of energy in the GW m^{-2} range on the inner wall in comparatively short periods of time [9]. These transient scenarios occur during so-called normal and off-normal events that arise from instabilities and insufficient control of the plasma. An overview of these off-normal events, as exemplified for the ITER reactor with the associated power densities and pulse durations is shown in fig. 2.8.

Edge Localized Modes (ELMs) fall under this category of transient events. These are MHD instabilities that occur in the pedestal region during operation in H-mode. They cause a periodic energy loss at the plasma edge [59]. The pressure gradient at the plasma edge increases up to a certain value in H-mode. Once this threshold is exceeded, plasma is flushed out into the scrape-off layer (SOL). These plasma particles are then carried into the divertor along the magnetic field lines, where they transfer most of their energy to the PFCs [59, 84–86]. ELMs can be classified into three categories according to their characteristics [87].

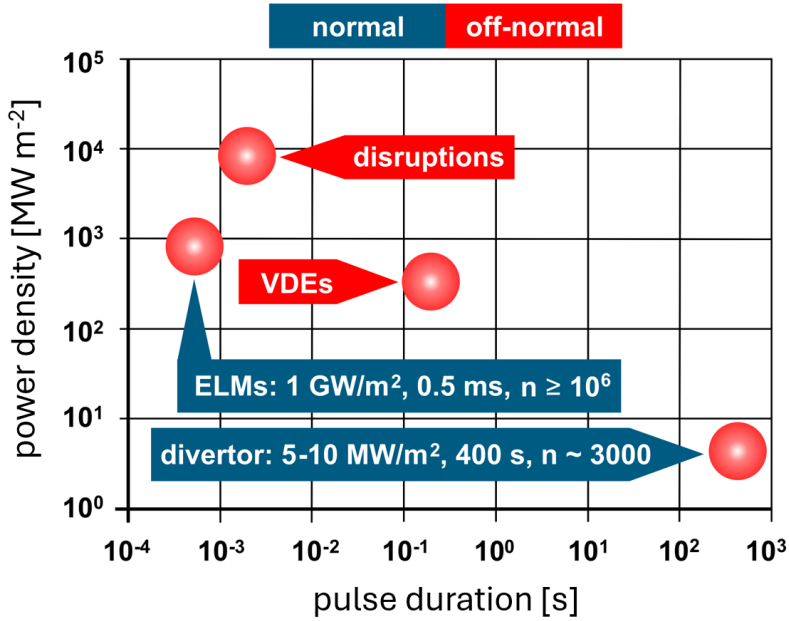


Figure 2.8: Various normal and off-normal events with the associated pulse durations and the power densities impinging on the inner wall of the ITER reactor are plotted. Steady state thermal loads and ELMs are part of the normal divertor loading scenario in H-mode operation. Plasma disruptions and vertical displacement events (VDEs), on the other hand, are considered off-normal events [80–83].

Type-I ELMs, also referred to as giant ELMs, deposit large amounts of energy on the inner wall of the reactor in short periods of time. This effect is caused by a certain amount of the energy stored in the plasma being released into the SOL. These type-I ELMs occur in time spans of 0.2 to 0.5 ms at frequencies of several Hz [85,86]. In a full-performance ITER plasma discharge, where $Q = 10$ will be reached, an energy of ~ 30 MJ can be released by a single unmitigated type-I ELM, resulting in a heat flux of 10 to 15 MJ m⁻² on the divertor PFCs. This amount of energy exceeds the divertor ablation limit by a factor of 20 to 30, which means that measures must be taken to mitigate such ELMs. There is a range of concepts for ELM mitigation, whereby the main concepts for ITER are the use of pellet injection and resonant magnetic perturbation (RMP) fields [88]. For pellet injection, frozen deuterium pellets are shot into the plasma core at the frequency at which giant ELMs occur, in order to cool the plasma core and thereby mitigate the ELM [89–91]. The second mitigation method mentioned involves the use of additional magnetic coils to generate small RMP fields that modify the pedestal plasma in the H-mode in a way that suppresses or mitigates ELMs [92,93].

Type-II ELMs, also known as grassy ELMs, are considerably weaker than type-I ELMs and occur at a higher frequency of several thousand Hz [94].

Type-III ELMs are weak and therefore also called small ELMs. The frequency of their occurrence decreases with increasing plasma heating power until they disappear completely [94].

Vertical Displacement Events (VDEs) are transient off-normal events characterized by the loss of the vertical equilibrium of the plasma, causing the plasma to move up or down in the chamber. This can result in contact between the plasma and the wall, thus depositing high doses of energy $\sim 60 \text{ MJ m}^{-2}$ in the case of ITER on the wall over timescales of 100 to 300 ms. This can lead to severe damage to the PFCs. These VDEs can be triggered by plasma instabilities, control system failures and ELMs. Active control of the magnetic fields, cooling of the plasma by pellet injection and restriction of the operational regimes can help to mitigate or prevent VDEs [95,96].

Plasma disruptions are transient off-normal events where the plasma breaks down due to instabilities caused by electromagnetic forces. Due to the control loss, large amounts of energy of over 30 MJ m^{-2} are deposited on the inner wall of the reactor in time frames of 0.1 to 3 ms at ITER. The sequence of a plasma disruption can be divided into three phases. In the precursor phase, changes occur in the magnetic fields and the plasma. This is followed by the thermal quench phase, in which the temperature of the plasma drops rapidly and the energy confined in the plasma is transferred as thermal energy to the SOL and then to the PFCs. In the last phase, the current quench, the plasma current drops rapidly [85,97].

Particle fluxes

The PFM has to withstand high particle fluxes during plasma discharges of nuclear fusion reactors. It must withstand electron, ion and neutral particle fluxes. At the strike line in the ITER divertor, particle fluxes of $> 10^{24} \text{ m}^{-2} \text{ s}^{-1}$ are expected for the respective particle types mentioned. The ions are transported along the magnetic field lines in the SOL and the separatrix until they encounter the PFCs of the divertor. In a 400 s ITER plasma discharge an ion fluence of $> 10^{26} \text{ m}^{-2}$ is expected to impinge on the PFM of the divertor. At the boundary between plasma and solid matter, a thin electrically charged layer, the plasma sheath, is formed due to the different masses and mobilities of the ions and electrons. The so-called sheath potential in this region causes the charged particles to accelerate shortly before they hit the solid PFM. This means that at a divertor plasma density of 10^{21} m^{-3} with a plasma temperature of 3 eV, a D^+ impact energy of $\sim 15 \text{ eV}$ can be expected due to the sheath potential acceleration. These steady state load conditions can cause

the PFM to heat up to peak temperatures of ~ 2000 °C in the divertor region and ~ 300 °C in the first wall [98–102]. Transient loading scenarios as shown in fig. 2.8 can cause these temperatures to be exceeded.

The ion fluxes to be expected on the first wall are lower than in the divertor region, since the magnetic field lines do not guide the ions directly onto the first wall, but instead impinge on the wall in the form of small ion fluxes based on neoclassical and anomalous transport [30]. At the first wall of ITER, neutral fluxes of D ions in the range of 10^{19} to 10^{21} m $^{-2}$ s $^{-1}$ with typical energies of ~ 8 to 300 eV are expected, according to modelling results. The ion fluxes at the first wall are thus three orders of magnitude lower than those expected in the divertor. In these estimates, the long-range transport in the SOL was taken into account [98,99].

Another aspect of particle fluxes that poses a challenge for PFMs are so-called runaway electrons. These electrons are accelerated to relativistic speeds, escape the magnetic confinement and deposit high energies on impact with the PFCs, where they can cause significant damage. They occur in plasma disruptions during the thermal quench phase, when the high electric fields lead to a large number of high-energy electrons. With increasing energy, the coulomb collision frequency decreases, causing the high-energy electrons to be accelerated ever further until they escape the magnetic confinement (i.e. they run away). In future fusion reactors, the avalanche mechanism can produce further runaway electrons through near-Coulomb collisions of existing runaway electrons with bulk electrons. Therefore, there is great interest in researching experimental methods to control runaway electrons [43].

One of the biggest challenges for PFMs is the bombardment with up to 14 MeV (see equation (2.1)) neutrons. The neutrons can lead to significant volumetric material damages, by causing changes in the lattice structure. Not only by generating defects, but also by transmutation, leading to alteration of the material properties [43, 103]. During the lifetime of the PFCs in ITER, 1 dpa (displacements per atom) is expected for the PFM in the divertor. The PFM of DEMO, on the other hand, is expected to reach 4 to 8 dpa and the heat sink material 5 to 15 dpa during 1.5 to 2 full power years [9, 83, 104, 105].

2.4 Plasma-facing materials

After describing the loading conditions which plasma-facing materials (PFMs) in nuclear fusion reactors are subjected to, this section describes the properties which PFMs should possess. In future FPPs, the PFMs must withstand these challenging conditions for years without failure. This, the consideration of further boundary conditions necessary to achieve a good plasma performance and the minimization

of nuclear waste production severely limits the choice of suitable materials.

High thermal conductivity and high melting/sublimation point

One of the most critical properties of PFMs is their ability to rapidly transfer thermal energy to the underlying cooling structure. This defines also the maximum temperature which occurs during thermal loading and makes thermal conductivity one of the most important physical properties of PFMs. Naturally, due to the high surface temperatures on the PFCs during reactor operation, the melting temperature, or in the case of carbon, the sublimation temperature, must also be high, in order to maintain the solid phase. Tungsten stands out as a potentially suitable candidate for PFM with the highest melting point of all elements ($T_{\text{melt}} = 3422\text{ }^{\circ}\text{C}$) and a high thermal conductivity ($\kappa \approx 183\text{ W m}^{-1}\text{ K}^{-1}$ at RT) [9, 17, 106]. In fig. 2.9, the surface temperatures of tungsten-armored ITER divertor PFCs are plotted schematically for various combined steady state and transient heat fluxes, in front of the background of the preferred operating temperatures of tungsten. The change in

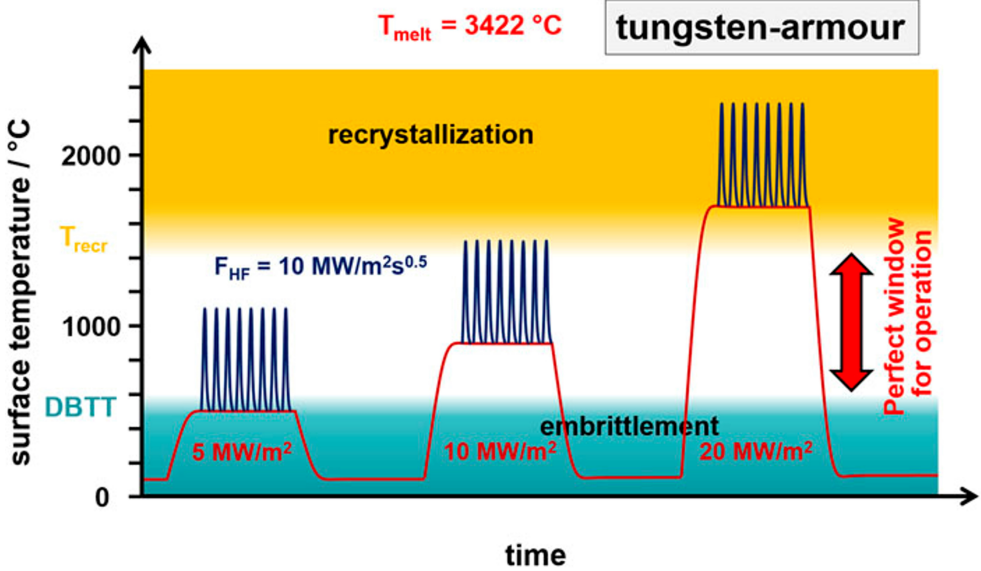


Figure 2.9: In red, the surface temperature change due to the steady state heat fluxes (5, 10 and 20 MW m⁻²) on the surface of tungsten armored ITER divertor PFCs is plotted schematically. In dark blue, the additional surface temperature rise caused by transient heat fluxes in the form of mitigated ELMs with an assumed heat flux factor of 10 MW s^{0.5} m⁻² is plotted. In the background, the white area represents the favorable operational window in regard to the operating temperature of tungsten [9, 107].

the surface temperature of the PFM due to transient thermal loads ΔT [K] on a

semi-infinite body can be approximated using the following equation [108, 109]:

$$\Delta T = 2I \sqrt{\frac{\Delta t}{\pi \kappa \rho c_p}} [\text{K}] \quad (2.14)$$

Where I [W m^{-2}] is the intensity, Δt [s] the pulse duration, κ [$\text{W m}^{-1} \text{K}^{-1}$] is the thermal conductivity, ρ [kg m^{-3}] the density and c_p [$\text{J kg}^{-1} \text{K}^{-1}$] is the specific heat capacity of the PFM. By separation of the material-specific temperature-dependent parameters, the equation becomes:

$$\Delta T = \frac{2}{\sqrt{\pi \kappa \rho c_p}} I \sqrt{\Delta t} [\text{K}] \quad (2.15)$$

The product of the intensity of the thermal load and the square-root of its duration determines the so-called heat flux factor:

$$F_{HF} = I \sqrt{\Delta t} [\text{MW s}^{0.5} \text{m}^{-2}] \quad (2.16)$$

The heat flux factor F_{HF} is used as a parameter to compare different loading conditions. The same heat flux factor F_{HF} results in the same surface temperature increase ΔT for a specific material. How tungsten reacts to the exposure to combined steady state and transient heat fluxes is schematically illustrated in fig. 2.10. When loaded, the material heats up and expands, resulting in compressive stresses and if the stress is high enough, in plastic deformation. When the material cools, shrinkage can cause tensile stresses. With increasing intensity, the changes in the material range from surface elevation, roughening, grain growth and recrystallization to cracking and edge melting. At extremely high heat fluxes, melting, bubble formation, splashing and epitaxial growth occur. Depending on the microstructure and chemical composition of the tungsten material, the damage pattern at different power densities varies [9, 81, 110–114]. Crack formation in PFM during short pulses of intense thermal loading, as experienced during ELMs, can be described by the relationship between thermal stress and material toughness [115, 116]:

$$\sigma_{therm} = \frac{E_M \alpha_{te}}{1 - \nu} \Delta T [\text{Pa}] \quad (2.17)$$

where σ_{therm} [Pa] is the thermally induced mechanical stress, E_M [Pa] the material specific Young's modulus, α_{te} [K^{-1}] the thermal expansion coefficient, ν the dimensionless Poisson's ratio and ΔT [K] the change in temperature. If the inner thermal stresses σ_{therm} exceed the ultimate tensile strength of the material σ_{UTS} plastic deformations and cracks are initiated [115]:

$$\sigma_{therm} > \sigma_{UTS} [\text{Pa}] \quad (2.18)$$

By substitution of equation 2.17 into equation 2.18, the cracking criterion can be written as [115]:

$$\Delta T > \sigma_{UTS} \frac{1 - \nu}{E_M \alpha_{th}} [\text{K}] \quad (2.19)$$

On the left side is the temperature change ΔT caused by transient thermal loads, which is the driving force for cracking. On the right side of equation 2.19 are material properties that represent the resistance of the material to cracking. If the temperature change ΔT is higher than the resistance of the material, cracking or plastic deformation starts. For pure tungsten, using a literature value for the σ_{UTS} of 800 MPa at 200 °C, the maximum tolerable temperature change calculated according to equation 2.19 is only ~ 330 K [106,117]. It should be noted that σ_{UTS} tends to decrease with increasing temperature and neutron irradiation [117].

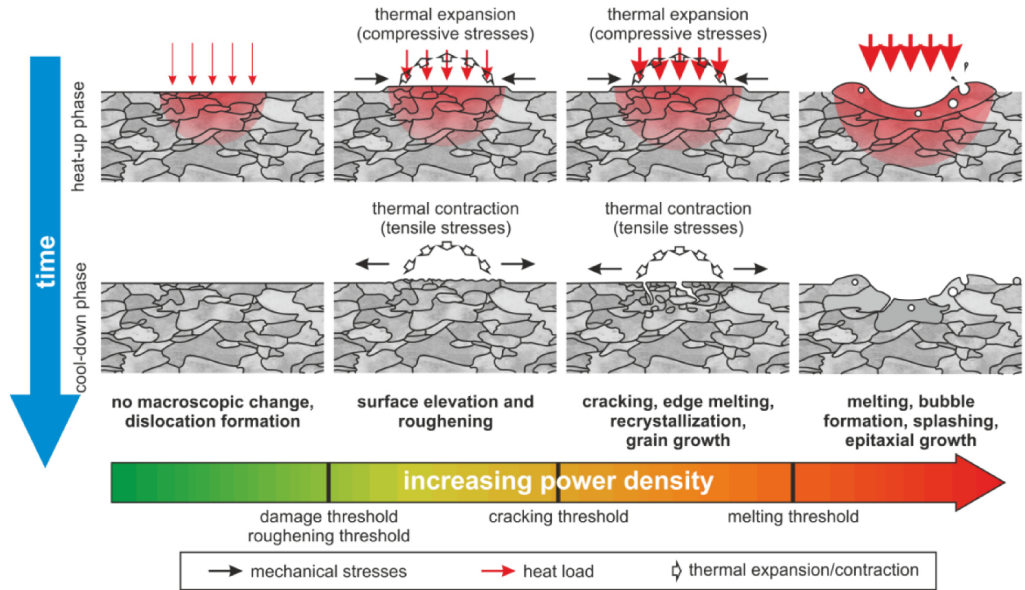


Figure 2.10: The behavior of tungsten when subjected to combined steady state and transient heat fluxes is illustrated. The various damage mechanisms that occur with increasing intensity are schematically illustrated [80,109].

Plasma compatibility

Contamination of the plasma by molten or broken-out PFM should be prevented as much as possible, since particularly materials with a nuclear charge number (Z) can significantly cool the plasma and cause it to collapse. Non-ionized atoms use the kinetic energy of the fuel ions, converting it into radiation through the excitation or deexcitation of electrons. This radiation energy is lost to the plasma and absorbed by the PFCs. Therefore, maximum contamination limits of the elements from the PFM in the plasma must be considered. For the high Z element tungsten, the concentration in the core plasma should be $< \text{few } 10^{-5}$ [10,118]. Low Z elements such as beryllium have the advantage that higher concentrations can be tolerated in the core plasma [119].

Sputtering resistance

Since physical and chemical sputtering have a major influence on the erosion of the PFM, which should be kept as small as possible, a high sputter resistance is an important aspect in the selection of the PFM [120]. The contamination of impurities in the plasma is highly dependent on the sputter threshold and yield of the PFM. Furthermore, tritium can be bound by chemical sputtering, which affects the reactors tritium inventory and should be avoided. CFC exhibits particularly high chemical sputter yields due to hydrocarbon formation. Tungsten is superior to beryllium and CFC in both physical and chemical sputtering due to comparatively low sputter yields.

Neutron activation/transmutation resistance

The collision of high-energy neutrons generated in the fusion reaction with atoms of the materials inside the vacuum vessel can lead to activation and transmutation of these materials. The number of unstable nuclides $N(t)$ can be described by an exponential decay law [121]:

$$N(t) = N_0 e^{-k_d t} \quad (2.20)$$

Where N_0 is the number of nuclides at the time ($t = 0$), t is the elapsed time [s] and k_d is the specific decay constant [s^{-1}]. In the context of neutron irradiation, fig. 2.11 shows the residual activation of various elements after 5 years of exposure to the neutron flux of $10^{15} \text{ cm}^{-2} \text{ s}^{-1}$, as expected in the first wall of DEMO [102, 122, 123]. Due to geometric reasons, the neutron flux is expected to be lower by a factor of 10 in the divertor region, even though the highest heat fluxes are expected there [124].

Neutron irradiation resistance

The material properties of many elements and materials are strongly influenced by the damage induced in the lattice by neutron irradiation, such as the creation of vacancies, transmutation and dislocations [125]. In order to test the properties of materials after neutron bombardment, research currently relies on the exposure of samples in nuclear fission reactors or on proxies like self ion or proton irradiation. To irradiate materials with neutrons of 14 MeV, as they occur in the DT fusion reaction, the International Fusion Materials Irradiation Facility-DEMO Oriented Neutron Source (IFMIF-DONES) facility is currently being built [126]. For example, the physical property of thermal conductivity, which is essential for the application of PFM, decreases significantly after irradiation with neutrons in carbon [127]. In contrast, neutron irradiation has a smaller effect on the thermal conductivity of tungsten. Experimentally obtained data such as that is essential for the selection of a suitable PFM.

Tritium retention

To maximize tritium recycling and achieve a high tritium breeding ratio (TBR),

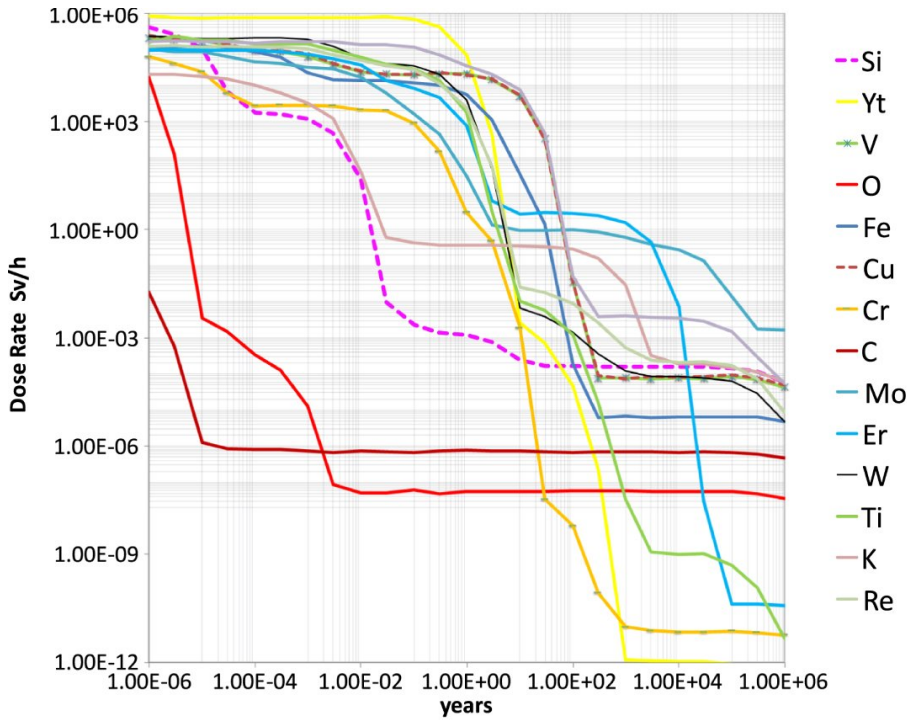


Figure 2.11: This plot visualizes the residual activation of various elements. The degree and duration of the residual activation narrows down the selection of suitable elements. As a point of reference, here are the limits set for exposure to humans. Shielded hands-on level: 2 mSv h^{-1} , hands-on level: $10 \text{ } \mu\text{Sv h}^{-1}$ [102, 122].

tritium retention in the entire vacuum vessel should be kept as low as possible. For the tritium retention of materials, effects such as the change in tritium retention after neutron bombardment must be taken into account. Furthermore, the tritium retention of the inner wall of the vacuum vessel strongly influences the radioactive inventory of the reactor and should therefore be kept as low as possible. It should be mentioned that the fuel retained in the PFM can be partially released by heating and used in the next plasma discharge.

Since no material perfectly fulfills all the requirements that are placed on PFMs, a compromise must be made at this point and the most suitable material selected. The three most promising and most widely researched PFMs to date are listed in Table 2.2 with their respective advantages and disadvantages. Pure tungsten is currently considered to be the most promising candidate for PFM.

Table 2.2: The most relevant candidates for the PFM of large fusion reactors are tungsten (W), beryllium (Be) and carbon fibre composites (CFC). Based on the advantages and disadvantages listed below, tungsten is currently considered the preferred material for future fusion reactors and power plants [128, 129].

Material	Advantages	Disadvantages
W	• high thermal conductivity • highest melting point of all elements • low influence of temperature and neutron irradiation on thermal conductivity • low erosion rate • high thermal shock resistance • low tritium inventory	• poor mechanical properties • high Z (low concentration allowed in the plasma) • brittleness increases under neutron irradiation • high α_{te} mismatch with Cu or steel heat sink • activation under neutron exposure
Be	• high thermal conductivity • low Z (high concentration allowed in the plasma) • oxygen getter (improves vacuum conditions) • no chemical sputtering	• low melting point • toxicity (carcinogenic dust) • thermal conductivity decreases with temperature • high swelling
CFC	• excellent thermal conductivity • low Z (high concentration allowed in the plasma) • high thermal shock resistance • no melting point (instead: high sublimation point)	• thermal conductivity decreases with temperature and neutron irradiation • high tritium inventory • formation of hydrocarbons

2.5 Plasma-facing components

Heat and particle exhaust concepts

Dependent on the heat and particle exhaust concept, PFCs are located in different regions of the fusion reactor. Since these components will experience varying levels of stress dependent on their purpose and location in the reactor, different designs are chosen. To provide an appropriate introduction to PFCs, an explanation of the heat and particle exhaust concepts has to be given. Two key concepts have been developed for the heat and particle exhaust of fusion reactors. These include the limiter and divertor concept (see fig. 2.12). In the limiter configuration, the plasma of the SOL is in direct contact with the inner wall since the PFCs of the limiters reach into the SOL. Consequently, particles leaving the confined plasma and entering the SOL

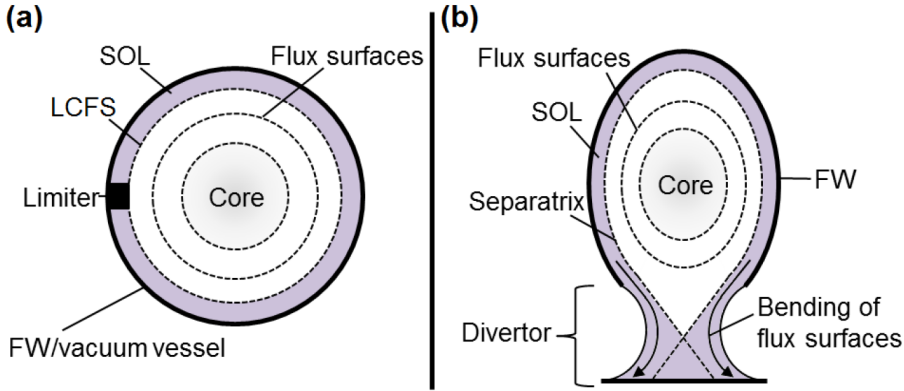


Figure 2.12: The poloidal cross-section of the magnetic geometry for two different heat and particle exhaust concepts is illustrated schematically. (a) shows the limiter configuration where the limiter that protrudes into the plasma determines the last closed flux surface (LCFS) and separates the core plasma from the SOL. (b) displays the magnetic field lines diverted by additional magnetic field coils in the divertor configuration. Here, particles that escape from the core plasma into the SOL are guided along the magnetic field lines into the divertor, where they impinge on the designated PFCs [130].

directly hit the limiter components. Between the inner confined plasma and the SOL lies the so-called last closed flux surface (LCFS) [131]. A serious disadvantage of the limiter configuration is a comparatively high degree of contamination due to erosion of the limiter armor. The divertor concept offers a reduction in the degree of impurities and, consequently, improved plasma performance. In tokamaks featuring divertors, additional magnetic field coils divert the outer magnetic field lines, thereby directing the plasma in the SOL along the magnetic field lines to the PFCs of the divertor. The boundary between the confined plasma and the SOL is called the separatrix. In the divertor the particles are largely neutralized and removed by the vacuum pumps. The openings of the pumps are positioned in the divertor behind the armor at locations with relatively high neutral particle pressure. This way, the fusion product in the form of helium ash and impurities can be removed from the vacuum vessel to achieve a purer plasma. With the single-null concept, the divertor is located at the bottom of the vessel. In the double-null concept, on the other hand, there are two divertors to increase the area affected by the plasma and thereby reduce the loads per area. In this case, one divertor is located at the top and one at the bottom of the vacuum vessel [132]. Additionally, there are further divertor concepts such as the super x divertor (SXD) design where the arms of the divertor are extended to also increase the plasma-wetted area [133]. Depending on the heat exhaust concept of the fusion reactor, a distinction is made between the limiter or divertor PFCs and those in the first wall (FW). The limiter and divertor

components, respectively, have to withstand higher stresses than the components in the FW (see 2.3) and are therefore designed to correspondingly high standards of robustness. Regardless of where the PFCs are located, active cooling is required to prevent overheating [17, 134].

First wall PFCs

For the PFCs of the first wall in ITER and European DEMO, the flat tile design is planned (see fig. 2.13) [17, 135, 136]. On the plasma-facing surface of the first wall of the European DEMO FPP, the incoming steady state heat loads are currently expected to be around 1 MW m^{-2} [137, 138]. Due to the pulsed operation of tokamaks,

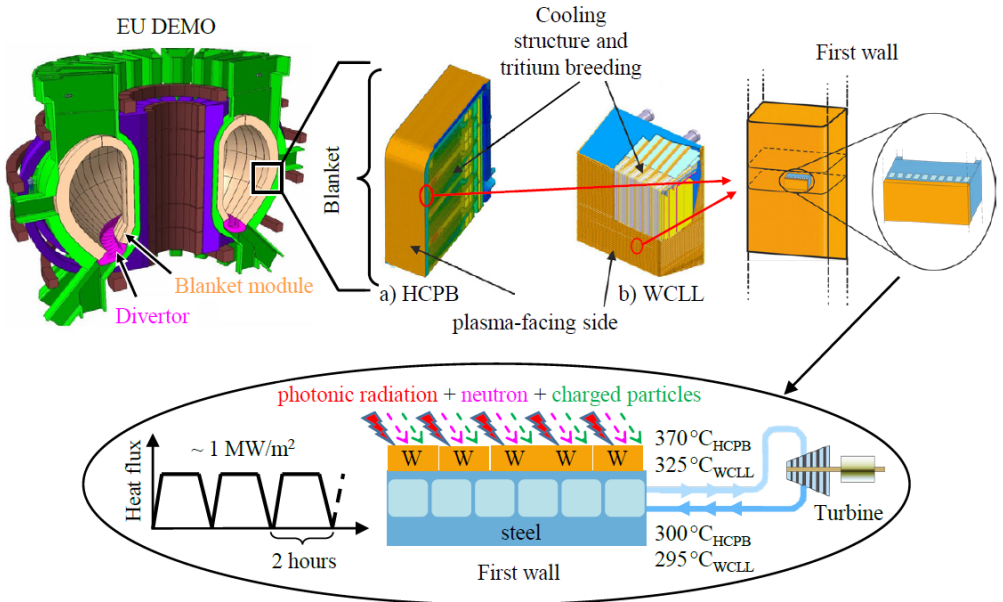


Figure 2.13: The illustration depicts the two concepts for European DEMO blankets, with a) Helium Cooled Pebble Bed (HCPB) and b) Water Cooled Lead Lithium (WCLL) and with a schematic representation of a FW below [17, 137, 139, 140].

these heat fluxes occur cyclically. To withstand these stresses, the flat tile components are most likely armored with 2 mm thick tungsten as PFM, and the underlying structural material is made of a specially developed steel alloy [17]. These materials are joined together by diffusion welding. Reduced activation ferritic/martensitic (RAFM) steels are suitable candidates for the underlying structural material [141]. The most prominent representative of RAFM steels is the alloy EUROFER 97, developed in Europe in 1997. This is a modified DIN X10CrWMoVNB9-2 steel where the high activation elements such as Mo, Nb, Ni, etc. have been removed or replaced

by low activation element variants [142]. The chemical composition of the latest EUROFER 97 development is listed in Table 2.3 [143]. This material is characterized by the following properties that are relevant in the context of the structural material of fusion reactors: Its activation dose (neutron activation) is below the recycling limit after 100 years of storage, comparatively high resistance to damage from neutron irradiation, high creep strength, high temperature strength and high toughness [143]. A major challenge in the development of resilient tungsten-armored PFCs is the joint

Table 2.3: The chemical composition in wt% with the main alloying elements of EUROFER 97 steel is listed below. This is a reduced activation ferritic/martensitic (RAFM) steel, which is one of the most promising candidates for the structural material used for the heat sink of the first wall of nuclear fusion reactors [128, 129].

	C	Cr	Mn	W	Ta	V	N ₂
min	0.09	8.50	0.20	1.00	0.10	0.15	0.015
max	0.12	9.50	0.60	1.20	0.14	0.25	0.045

between the PFM and the underlying structural material. When the plasma-facing surface of PFCs is subjected to high heat fluxes, large temperature gradients arise in the material between this surface and the cooling channel. Furthermore, different properties of tungsten and steel, can lead to large residual stresses in the component. Here, the difference in the coefficient of thermal expansion α_{te} plays a major role: $4.5 \times 10^{-6} \text{ K}^{-1}$ for tungsten and $10.3 \times 10^{-6} \text{ K}^{-1}$ for steel. The difference in α_{te} has an effect on the joint, especially during cyclic hot-cold temperature changes, such as those arising during the pulsed plasma operation of tokamaks. Furthermore, a challenge in the realization of the European DEMO FPP will be, to ensure that the 1200 m^2 of FW surface area is constructed with defect-free joints between PFM and structural material [137]. Moreover, to achieve tritium self-sufficiency in the FPP, it is necessary to install tritium breeding blankets behind the first wall. According to assessment studies, the European DEMO project is currently focusing on two types of breeding blankets [14, 137]. These two concepts are shown schematically in fig. 2.13. The plasma-facing surface of the breeding blankets is the first wall, which is highlighted in orange and shown schematically by fig. 2.13 in the lower area [137]. The challenges mentioned here are only intended to provide an insight into the complex subject of FW PFCs that are expected to be in operation for several years in future FPPs.

Divertor PFCs

The thermal stresses in the divertor area are, with cyclic steady state heat fluxes of ~ 5 to 20 MW m^{-2} on the plasma facing surface of PFCs, even larger than those acting on the FW. Due to the requirements on the PFCs in the divertor area, which

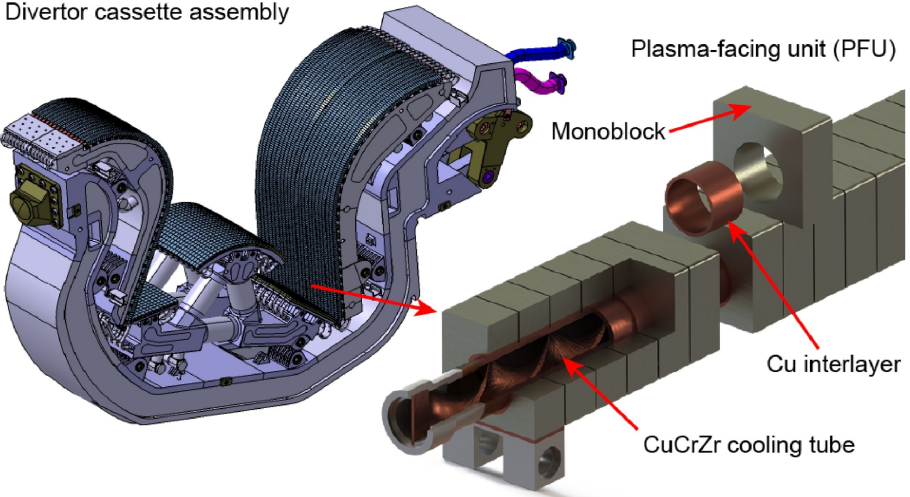


Figure 2.14: Illustrated is the CAD model for one of the 54 divertor cassettes of ITER, with an exploded view of a monoblock chain PFU for the outer vertical target [144].

differ from those in the FW, other PFC designs are used here (see fig. 2.14 for CAD models of an ITER divertor cassette and a monoblock chain plasma-facing unit (PFU)). Divertor PFCs are required to transfer large amounts of thermal energy from the plasma-facing surface through the PFM and the heat sink material to the coolant. Therefore, these materials must have particularly good thermal conductivity in the highly loaded divertor area. For this reason, in the divertor area, tungsten ($\kappa \approx 183 \text{ m}^{-1} \text{ K}^{-1}$ at RT) is used as the PFM and copper ($\kappa \approx 400 \text{ m}^{-1} \text{ K}^{-1}$ at RT), or copper alloy like CuCrZr ($\kappa \approx 360 \text{ m}^{-1} \text{ K}^{-1}$ at RT) respectively, are chosen as the heat sink material [106, 145, 146]. The outer and inner vertical targets of the divertors in ITER and DEMO will use so-called monoblock chains as PFCs [147]. These are tungsten blocks mounted on CuCrZr cooling tubes. CuCrZr is a precipitation hardened copper alloy that, due to its comparatively high strength even at elevated temperatures, is used as heat sink material for thermally highly stressed PFCs [146, 148, 149]. The monoblocks are joined to the cooling tube by hot radial pressing with an oxygen free high conductivity (OFHC)-Cu layer in between (see fig. 2.14) [150, 151]. OFHC-Cu is used as a soft layer between the hard tungsten and CuCrZr to ensure a reliable connection. Due to the geometry of the monoblock enclosing the cooling tube, the tungsten part remains attached to the cooling tube even in the event of a large crack forming at the joint. This inherently excludes the possibility of the entire tungsten armor falling into the plasma, as feared with the flat tile design, due to delamination of the armor from the heat sink.

2.6 Remote maintenance systems

Exchange of PFCs

The current strategy is to replace PFCs when they reach the end of their service life. Since nuclear decay in the neutron-irradiated materials precludes the presence of humans in the vacuum vessel, replacement of PFCs has to be done by specially designed remote controlled robots. After a four-week cool-down period, radiation levels of 500 Gy hr^{-1} are anticipated within the ITER plasma vessel [152]. In DEMO, even higher radiation levels of 2000 Gy hr^{-1} are expected after the four-week cool-down period [153]. Radiation levels in this range can have severely detrimental effects on electronics or polymers. This requires electronics to be either shielded or positioned outside the vacuum vessel.

When the PFCs in the divertor area reach the end of their service life, ITER plans to replace the entire divertor cassette (fig. 2.14). The ITER divertor is currently expected to reach the end of its service life ~ 16 years after the first plasma operation [11]. In previously published works, shorter lifespans of 3 to 4 years were predicted [154]. It needs to be mentioned that the lifetime is highly dependent on the planned operational scenario. The divertor cassettes shall be replaced by means of the in-vessel remote handling (RH) system, as schematically shown in fig. 2.15 [154].

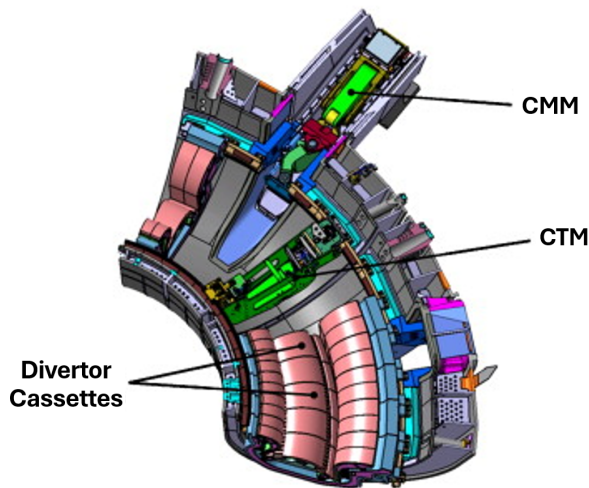


Figure 2.15: The illustration schematically depicts the in-vessel divertor cassette RH for ITER. The divertor cassettes (fig. 2.14) are transported through the RH port by means of the cassette multifunctional mover (CMM). The cassette toroidal movers (CTM) can be used to position the cassettes on rails in the toroidal direction [154].

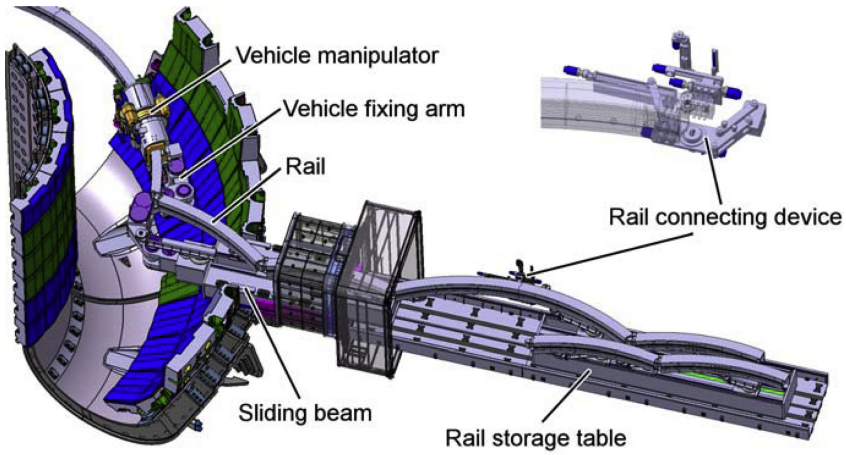


Figure 2.16: CAD model of the ITER blanket RH system (BRHS) during rail assembly and deployment from the equatorial port [155].

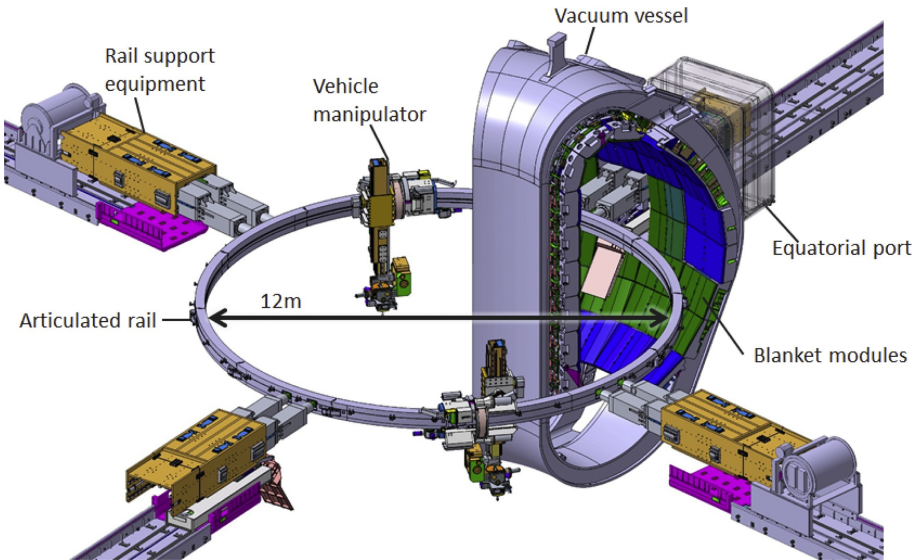


Figure 2.17: CAD model of the fully assembled ITER blanket RH system (BRHS), including the articulated rail-mounted vehicle manipulator [155].

Since each divertor cassette weighs ~ 9 tons, the RH system must be designed to be correspondingly robust [134].

For ITER the system used to exchange entire FW blanket modules is called the

ITER blanket RH system (BRHS) (see fig. 2.16 and fig. 2.17) [155,156]. Each blanket module weighs up to 4.5 tons and features $\sim 1.5 \text{ m}^2$ of plasma-facing surface area [65]. As illustrated in fig. 2.16, a retractable rail, including the vehicle manipulator, is driven out of the equatorial ports. After successful deployment of the BRHS, the vehicle manipulator can move along this rail and pick up the desired blanket module (fig. 2.17). The blanket modules of the FW must, therefore, be designed to allow their replacement by the BRHS after an extended period of operation. In addition to the replacement of failed FW panels and shield blocks, at least one replacement of all 440 FW panels is planned during ITER's lifetime, which is expected to last two years [144]. To deploy the entire BRHS system as displayed in fig. 2.17, four equatorial ports are required. So, the construction of the rail system alone is complex and time-consuming. The BRHS is designed to be sufficiently strong to move the multi-ton blanket modules. However, this massive design is not necessary for in situ repair, as the LMD-w processing head will most likely be comparatively lightweight.

Multipurpose deployers

For more delicate remote maintenance (RM) tasks, such as inspection, measurements, small maintenance, dust monitoring, and removal or rescue operations, dedicated multipurpose deployers (MPDs) are used (ITER MPD fig. 2.18 and current DEMO MPD design iteration fig. 2.19), that can be deployed faster and more easily than the BRHS. Depending on the particular application, MPDs can be configured with different end-effectors, making them highly flexible. This makes MPDs the RM system most likely to be used for in situ LMD-w repair. Ensuring the necessary precision and limiting oscillations as much as possible despite the large dimensions and at times high payloads poses major challenges for MPD systems. In addition, the systems need to be designed to reliably withstand the specific conditions in the vacuum vessel (high radiation exposure, elevated temperatures, possibly operation under vacuum). RH booms have already been successfully utilized in the JET fusion reactor together with the MASCOT end-effector for regular RM tasks [43]. Furthermore parts of the carbon PFCs of JET were changed using RH to beryllium and tungsten.

The ITER MPD (fig. 2.18) is a cask-based system and therefore remains inside the hot cell building during fusion reactor operation [157]. When it is inside the cask, it is folded on the vertical plane, which is possible due to the high degrees of freedom of its joints. Its structure consists of rotational joints and fewer initial planar joints. The main part of the MPD is the articulated transporter (AT), which is deployed "elbow-first" from the cask into the vacuum vessel where it unfolds [158]. The AT is moved from the cask into the vacuum vessel using the mounting equipment (ME). The ME and the AT together form the MPD transporter. To ensure the necessary rigidity, this MPD transporter is solidly mounted to the vacuum vessel

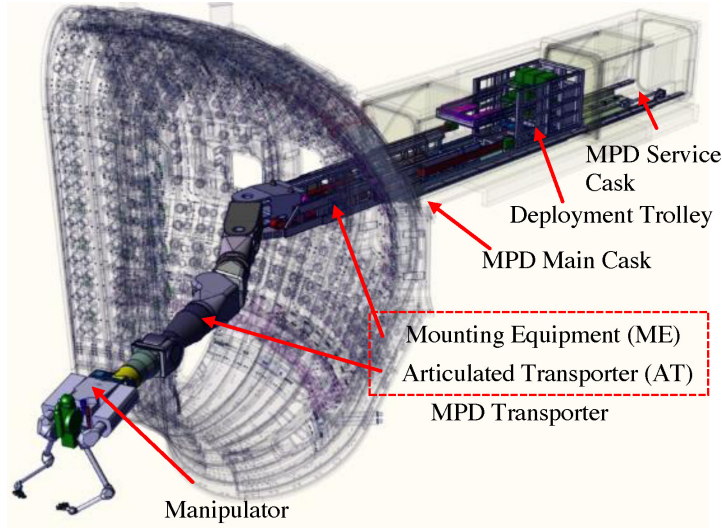


Figure 2.18: The ITER MPD deployed inside the vacuum vessel [157].

port. The modular design of the ITER MPD allows it to be configured either as a long-reach light payload or a short-reach high payload variant with a maximum payload capacity of 2 tons. By using 4 equatorial RH ports, the AT with its 9 degrees of freedom is able to reach the entire inner surface of the vacuum vessel. However, to reach the divertor, the long-reach low payload configuration is necessary. Certain tasks can be performed through collaborative operation of multiple MPDs. The deployment trolley transports tools and components from the service cask to the main cask, allowing the AT to be supplied with these items without the need to leave the vacuum vessel [157]. The last module of the MPD is the application-specific end-effector. Fig. 2.18 and fig. 2.19 displays the MPDs with so-called manipulators as end-effectors. For in situ regeneration of the PFM, a dedicated LMD-w processing end-effector would need to be developed that can be mounted on the AT.

Since the final requirements and constraints for the DEMO MPD still need to be defined, there is no final DEMO MPD design yet. Instead, there is only the current design iteration, which is displayed in fig. 2.19. In this illustration, MASCOT, which was utilized in JET, was taken as a placeholder for the end-effector manipulator [158]. Compared to the foldable ITER MPD, the current design iteration of the DEMO MPD foresees a boom-type deployer. Since these boom-type MPDs have already been successfully deployed in previous reactors, this design option is currently at least at technology readiness level (TRL) 6. In comparison, the foldable ITER MPD design is currently still at TRL 4. Compared to the foldable design, boom-type deployers offer higher payload capacity, controllability, lower complexity and thus easier port deployment, maintenance and recoverability. However, the dis-

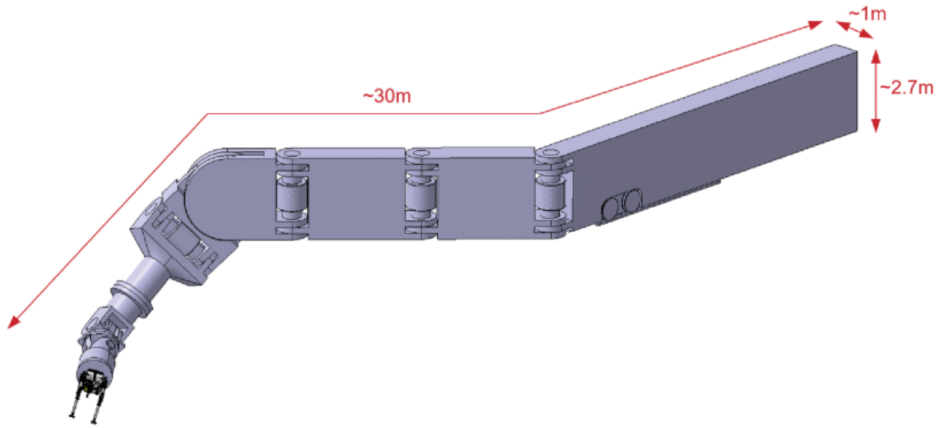


Figure 2.19: The current design iteration of the DEMO MPD, a boom-type deployer with fixed link and roller port connection [158].

advantage of boom-type deployers compared to the foldable variant is that they take up more volume and usually require a separate port for storage transfer. For the current design iteration of the DEMO MPD, a target payload capacity of 1000 kg is mentioned [158]. Since this is only a target payload capacity for the current design iteration and as it is not clear whether this is a long reach variant, that can operate within the divertor area, this value should be treated with caution. However, since it can be equipped with a MASCOT end-effector, which was also utilized for RM tasks in JET, weighing 250 kg, this can be assumed to be the minimum payload capacity of a DEMO MPD [43, 158]. Consequently, the LMD-w processing end-effector is permitted to weigh at least 250 kg, and most likely even more once reliable data on the MPD payload capacity is published. Since it should be possible to develop an LMD-w processing end-effector that weighs ≤ 250 kg, this is a favorable baseline scenario for the LMD-w repair concept.

3 Theoretical background of additive manufacturing

3.1 Basic characteristics and opportunities of AM

Additive manufacturing processes (AM) continue to gain in importance as development in digitalization and automation advances. By adding material layer by layer, AM enables the generation of complex shaped geometries, including those (nearly) impossibly to produce by classical means. According to the classic standard for manufacturing processes, DIN 8580, where all conventional manufacturing processes are classified into different main groups, AM processes can be assigned to main groups 5 (joining) and 6 (coating) [159]. However, for some time now, the AM processes have been defined and classified in the specially designated DIN EN ISO/ASTM 52900 [160]. In conventional manufacturing, products are usually made by a combination of primary shaping processes and subsequent forming or subtractive processes. In contrast to this, AM processes produce the product directly from the feedstock material. In AM, the components are created by depositing material layer by layer. Dependent on the product, both conventional and AM processes can be followed by joining or coating processes. The characteristic feature of AM processes is that the workpiece shape and the cohesion of the material are created simultaneously in a single production step [161, 162].

Since AM processes are not bound by the usual restrictions of conventional manufacturing methods, they offer the opportunity to manufacture almost unlimitedly complex geometries. In particular, the powder bed processes offer this advantage and allow the manufacturing of e.g. hollow or bionic structures. In the context of plasma-facing components, complex cooling structures can be realized to increase the cooling performance. In addition, new joining techniques between PFM and heat sink can be realized [163]. But these topics will be discussed in detail in the next section. Furthermore, near net-shape component geometries can be produced by AM, which allows material to be saved compared to conventional production [164]. And in powder-based processes, a large proportion of the unmolten powder can be recycled. In general, it can be said that AM processes offer great potential in lightweight construction and the conservation of resources [162, 165].

Nevertheless, purely additive process chains are often not economically viable due to the high process times and thus high costs [161]. However, AM processes offer the potential of efficiency increases, especially as a complement to conventional process sequences or replacement of individual process steps with AM. At the current development level, this applies primarily to small series. Additionally, it should be mentioned that additive manufactured products may require subtractive post-processing, depending on the process, material and designated application [166]. Hence, a combination of conventional and additive manufacturing processes can be a powerful approach to achieve optimized process efficiency. A sophisticated implementation of AM processes can contribute to an increase in efficiency and conservation of resources. In the following sections, various metal-based AM processes are introduced and discussed. It is described how these AM processes can be utilized in the context of robustness and lifetime enhancement measures for the PFC of nuclear fusion reactors.

The direct energy deposition (DED) processes (see section 3.4), a subcategory of the AM group, allow the localized deposition of material on preexisting structures or semi-finished products [167,168]. By means of DED, components can be repaired, wear layers applied or functional elements added. The service life of the products can be increased by these measures and the use of special materials with expensive alloying elements can be reduced. Within the context of fusion technology, the targeted use of DED processes in the vacuum vessel could repair damaged PFCs in situ and deposit new material to compensate for erosion losses [169–171]. The exploration of this topic is the subject of this work.

3.2 Fundamentals of additive manufacturing

Due to the unique challenges faced by PFCs of nuclear fusion reactors and the critical importance of preventing their failure, the utilization of all available options offered by modern technology is of interest for this application. These options include the utilisation of AM processes, which can unfold a range of their advantageous characteristics in this context. In the context of fusion technology, different AM processes offer a variety of distinct advantages. For the fabrication of innovative PFCs, powder bed fusion (PBF) processes were used for research purposes, mainly due to the freedom of geometry and precision that they provide. The laser-based powder bed fusion process for metallic materials is officially referred to as PBF-LB/M [160]. The same process is also referred to in certain sources as selective laser sintering (SLS) or melting (SLM) [172]. The electron beam-based powder bed fusion process for metallic materials is officially referred to as PBF-EB/M [160]. Depending on the publication, the abbreviations SEBM for selective electron beam melting or simply EBM are also used [173,174]. However, these refer to the same electron beam-based

powder bed fusion process. In the following, these two PBF processes and their use in building advanced PFCs is discussed. Fig. 3.1 shows a schematic illustration of the PBF process. In the PBF process, metallic powder is distributed in a thin layer by a coater (raker or roller) on the build platform and melted locally by electron (PBF-EB/M) or laser (PBF-LB/M) irradiation. To ensure sufficient material

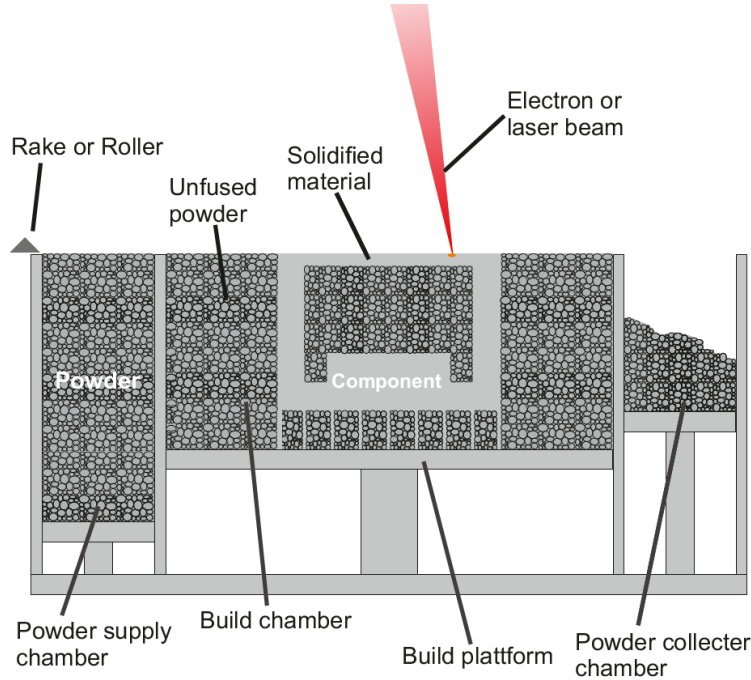


Figure 3.1: Illustrated is a schematic representation of the powder bed fusion (PBF) process. The energy required to selectively melt the powder can be applied by electron or laser beam.

deposition, more material than necessary is provided by raising the powder supply platform. The building platform is lowered slightly to allow a thin layer of powder to be applied. The coater ensures that the powder is evenly distributed on the building platform. Typically, the height of the layers is between 20 and 100 μm . The selected layer height depends on the material, powder particle size, component geometry and machine capability. Excess powder gets pushed into the powder collector chamber. The coater then returns to its initial position and further iterations of this process are carried out until the component is fully built. Depending on the particular process, there may be several intermediate steps where the material is preheated by fast scanning of the electron or laser beam. Since the magnetic lenses in the electron beam process allow rapid deflection, this kind of pre-heating is mainly available for electron beam systems. In the post process, the component is

cleaned of the unfused metal powder and, if necessary, supporting structures are removed. The excess and unfused powder can usually be reused after sieving processes.

The selected energy input has a significant influence on the end result. To establish comparability, the volumetric energy density E_v [J mm^{-3}] is often used, which incorporates the key process parameters for PBF processes [175]:

$$E_v = \frac{P}{v_s D_l h_d} [\text{J mm}^{-3}] \quad (3.1)$$

Where P is the power applied by the beam [W], v_s the scanning velocity of the beam [mm min^{-1}], D_l the height of the powder layer to be melted [m] and h_d the distance between the hatch lines [m]. Alternatively, the line energy input E_l [J mm^{-1}] is also used [175]:

$$E_l = \frac{P}{v_s} [\text{J mm}^{-1}] \quad (3.2)$$

Due to their simplicity, the volumetric energy density and the line energy input have become commonly used in scientific publications to quantify the energy input of different parameter combinations [176–181]. However, these approaches neglect parameters such as the focus diameter, absorption, thermal conductivity, heat capacity as well as melting and evaporation enthalpy. Accordingly, comparability between different materials and systems is not or only to a limited extent possible [176]. The normalised equivalent energy density E_0^* [J mm^{-3}] is better suited for the creation of standardised process parameter maps that cover multiple systems and materials [182]:

$$E_0^* = \frac{aP}{2v_s D_l h_d \rho c_p (T_m - T_0)} [\text{J mm}^{-3}] \quad (3.3)$$

Where a is the material specific absorption coefficient, ρ its density [kg m^{-3}], c_p its specific heat capacity [$\text{J kg}^{-1} \text{K}^{-1}$], T_m its melting point [K] and T_0 the initial temperature [K] of the material. Furthermore, the selected beam scanning strategy, which involves outlining the area to be melted, melting within this area using a specific hatch with defined spacing of the hatch lines and a varying orientation of the hatch lines per layer, has a significant influence on the result.

At low energy densities, the process is in the conduction mode, which is characterized by thermal conduction. In the case of low energy input, the process is prone to lack of fusion defects (LOF), in which the material melts only partially. The conduction-mode is characterized by $W_m > H_m$, where W_m is the width of the melt pool [m] and H_m is the depth of the melt pool [m] (see fig. 3.2). At high energy densities, the process enters the keyhole mode, which is characterized by a deeper melt pool and often features a gas void in the center. This regime is called keyhole mode because in a cross-section view the melt pool appears reminiscent to a

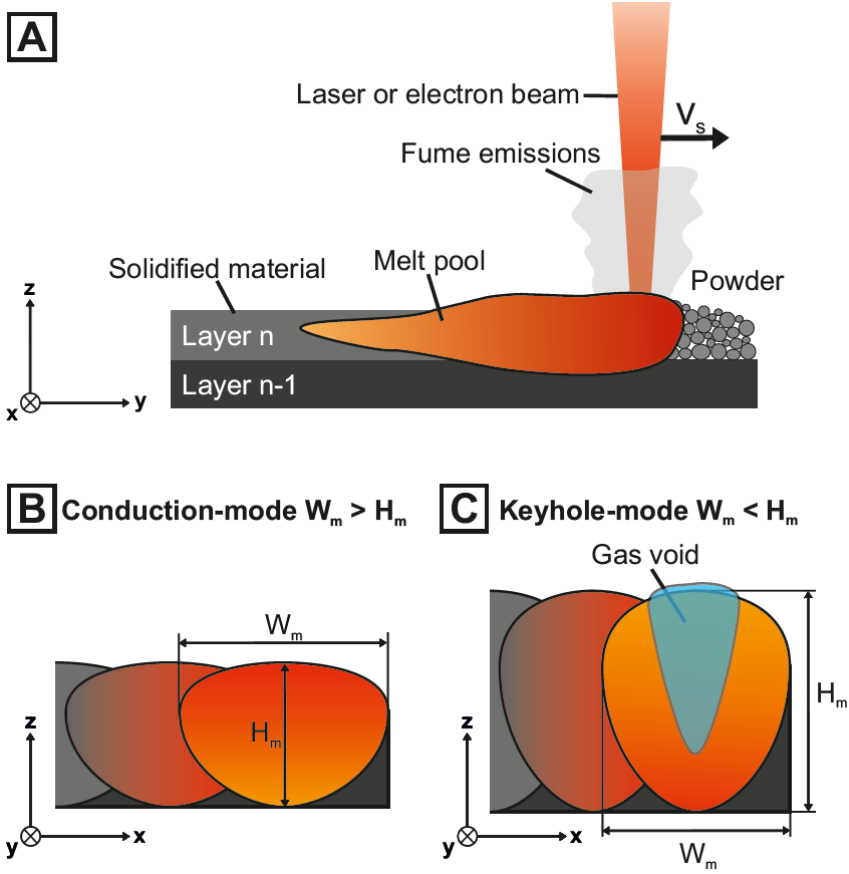


Figure 3.2: The figures schematically illustrate cross-sections of the PBF process with an electron or laser beam. A) depicts the process on the xz -plane. B) visualizes the process in the conduction mode on the yz -plane and C) at higher beam energy input in the keyhole mode [183].

keyhole. Due to reflections in the gas void, higher energy absorption is achieved in the keyhole mode. If gas bubbles detach from the bottom of the gas void, enter the melt pool, and the material solidifies before the gas can escape from the melt pool, a keyhole defect has formed. The keyhole-mode is characterized by $W_m < H_m$ (see fig. 3.2). Depending on the material, a suitable process window that avoids both of these defect types can be found somewhere in between.

The thermo capillary or surface tension and temperature driven Marangoni convection has a dominant influence on the fluid dynamics in the melt pool. This Marangoni convection driven fluid dynamics can lead to an increase of the melt pool depth. The currents induced by this force for the conduction and keyhole

welding modes are schematically visualized in fig. 3.3. The currents are therefore dependent on the surface tension gradient $\frac{d\gamma}{dT}$ [N m⁻¹ K⁻¹] and flow from areas of low surface tension to areas of high surface tension. The strength of these thermo capillary currents can be characterized by the dimensionless Marangoni number Ma [184, 185]:

$$Ma = \frac{d\gamma}{dT} \frac{dT}{dx} \frac{l^2}{\mu_d \alpha_{td}} \quad (3.4)$$

Where $\frac{d\gamma}{dT}$ is the surface tension gradient [N m⁻¹ K⁻¹], $\frac{dT}{dx}$ the temperature gradient [K m⁻¹], l the characteristic length [m] (in this case the width of the melt pool), μ_d the dynamic viscosity [kg m⁻¹ s⁻¹] and α_{td} the thermal diffusivity [m² s⁻¹].

For the fluid in the melt pool during welding processes, the dynamic viscosity μ_d has to be used [186–189]. The dynamic viscosity is defined as follows [190]:

$$\mu_d = \gamma \frac{16}{15} \sqrt{\frac{m}{k_B T}} [\text{kg m}^{-1} \text{s}^{-1}] \quad (3.5)$$

Where γ is the surface tension [N m⁻¹], m the atomic mass, k_B the Boltzmann constant [J K⁻¹] and T the temperature [K]. An increased temperature reduces the viscosity and improves the flowability of the fluid, as well as improving the wetability of the solid material. Besides, an increased viscosity can help to avoid the balling effect [188]. However, the Marangoni current also increases with increasing temperature, which leads to a higher probability of gas being drawn into the melt pool [189]. The temperature and, consequently, the selected energy density E_a [J m⁻²] required to achieve high quality process results should therefore be selected with precision.

The morphology of the melt tracks is mainly characterized by buoyancy forces, gas pressures and convection flows [183]. The flows in the melt pool are mainly dominated by Marangoni convection, which describes the thermocapillary flow in fluids from areas of lower surface tension to areas of higher surface tension [191]. The high temperature gradients associated with AM processes cause high gradients in surface tension $\frac{d\gamma}{dT}$. Where γ [N m⁻¹] is the surface tension and T [K] is the temperature. Exemplary schematic illustrations of melt pool flows are provided in fig. 3.3. If the surface tension decreases with increasing temperature $\frac{d\gamma}{dT} < 0$ so that the surface tension is lower at the hot center and higher at the cooler outside of the melt pool, the surface near melt flow from the center is directed radially outwards (see fig. 3.3 A)) [192]. If on the other hand, the surface tension increases with increasing temperature $\frac{d\gamma}{dT} > 0$ so that the surface tension is higher at the hot center and lower at the cooler outside of the melt pool, the surface near melt flow is directed radially inwards to the center (see fig. 3.3 B)) [192]. In addition to the chemical composition of the particular material, which determines the $\frac{d\gamma}{dT}$,

the exposure parameters also have an influence on the melt flow. Since the vapor capillary in keyhole mode can influence the surface tension gradient and moves the hottest point towards the bottom of the melt pool, which also influences the melt flow, the case $\frac{d\gamma}{dT} > 0$ in fig. 3.3 B) is visualized for the keyhole mode.

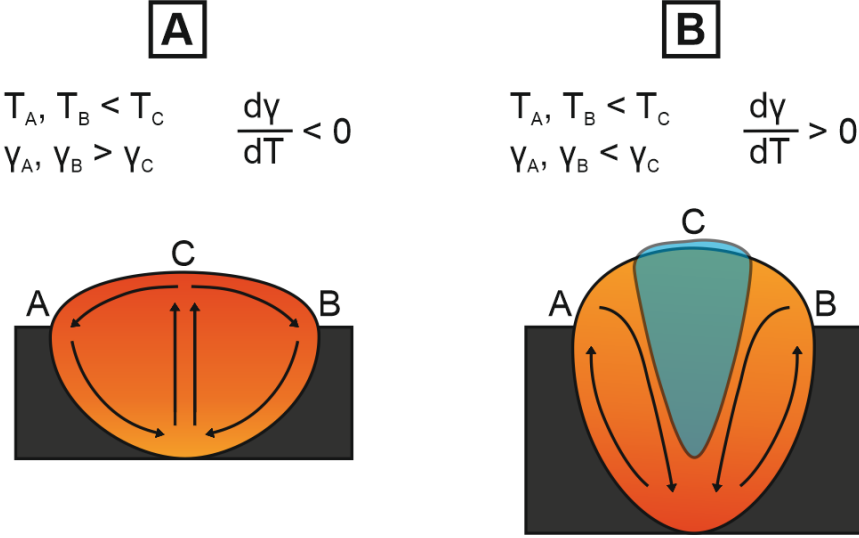


Figure 3.3: A schematic representation of the Marangoni convection that dominates in the melt pool and depends on the surface near surface tension gradient $\frac{d\gamma}{dT}$. A) shows a melt pool with $\frac{d\gamma}{dT} < 0$ in the conduction mode and B) a melt pool with $\frac{d\gamma}{dT} > 0$ in the keyhole mode [183, 184, 192].

3.3 Additive manufacturing of plasma-facing components

The flexibility and opportunities offered by AM technology allow the realization of advanced PFCs. AM can be used to better adapt the plasma-facing armor, the cooling structure and the joint between these parts to their specific requirements. The PBF processes discussed in the previous section are particularly suitable for the production of PFCs due to their precision and geometrical design freedom. Since tungsten will most likely be used as the PFM of future FPPs while RAFM steel is used as the cooling structure material in the first wall and CuCrZr in the divertor region, the AM processing of these materials needed to be researched initially in order to ensure sufficient product quality. Manufacturing of steel components using

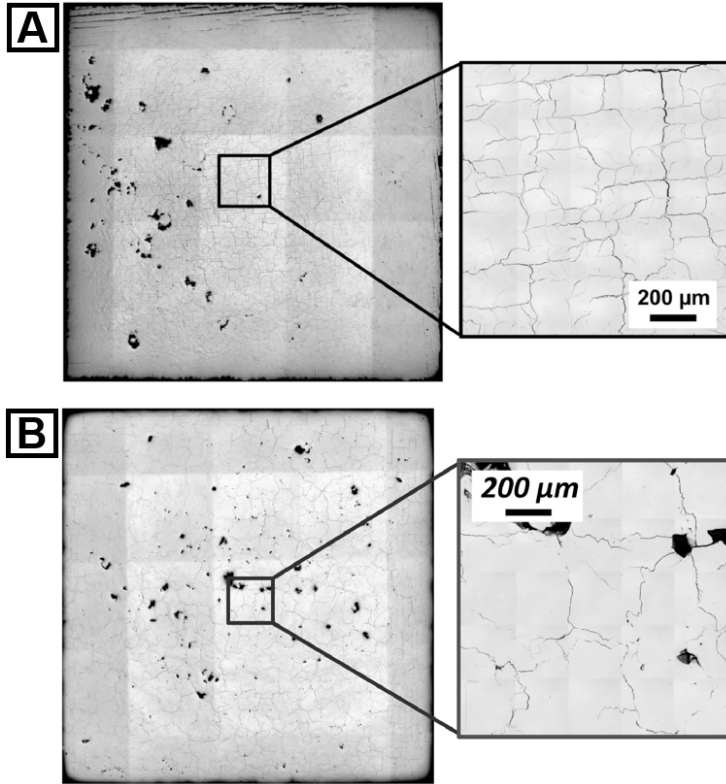


Figure 3.4: Cross-sectional micrographs of tungsten samples manufactured by PBF-LB/M with substrate heating. A) was manufactured at 200 °C substrate heating ($P = 400 \text{ W}$; $v_s = 500 \text{ mm s}^{-1}$) and B) at 1000 °C ($P = 400 \text{ W}$; $v_s = 510 \text{ mm s}^{-1}$) [193].

PBF poses the least problems, due to the fact that steel is relatively easy to process and many scientific studies were already published on additive manufacturing of steel [194].

In comparison, the production of CuCrZr components by PBF poses a significantly more challenging task. In the widely used PBF-LB/M process, infrared lasers are mostly utilized to melt the material. The low absorptivity of CuCrZr for near-infrared light thus presents a problem. To achieve sufficient bulk material density despite the relatively low absorptivity of near-IR radiation, this low absorption needs to be compensated by increased energy density. To increase the absorptivity, powder with a modified surface or laser light in the blue or green range with shorter wavelengths can be used [195]. For CuCrZr, the electron beam based PBF-EB/M process is better suited due to increased absorption in comparison to

the laser process.

However, the most difficult material in this context is the processing of tungsten by PBF. As a brittle refractory metal with an exceptionally high melting point, tungsten comes with several challenges for AM processing. A high amount of energy is needed to melt tungsten and therefore LOF defects can occur if energy densities are too low, resulting in insufficient melting of the tungsten powder (visible in fig. 3.4). Moreover, cracks can occur in the brittle tungsten due to the high temperature gradients during the process, which lead to residual stresses in the material during solidification (visible in fig. 3.4) [193, 196–201]. Substrate preheating to 1000 °C for PBF-LB/M helped to reduce the number of cracks, but as evident in fig. 3.4, they could not be prevented entirely.

Even though the production of dense bulk tungsten using PBF-EB/M is problematic, thin-walled tungsten structures of high quality can be produced, which are suitable for use in the production of composites between the PFM and the structural material [193, 202]. Tungsten copper composites are of interest in the context of PFCs, as they can be used to tailor the mechanical and thermophysical properties of the material. The discrepancy in the coefficient of thermal expansion α_{te} between tungsten ($4.5 \times 10^{-6} \text{ K}^{-1}$) and copper ($16.5 \times 10^{-6} \text{ K}^{-1}$) can cause thermally induced stress at the joints. By manufacturing tungsten copper composites in the form of additively manufactured tungsten structures that are infiltrated with copper, the desired coefficient of thermal expansion can be created and customized to the requirements of the application [193, 200]. In fig. 3.5, such a composite with an additively manufactured tungsten honeycomb structure that was subsequently infiltrated with copper is shown. Research is also being conducted on the corresponding simulations that generate optimized AM structures for stress reduction [202]. Using this approach, AM can be utilized to improve the durability and longevity of the joint between the PFM and the underlying cooling structure.

Although it was not demonstrated for the PBF/LB-M process to achieve defect-free bulk tungsten, crack-free microstructures and high densities of 99.8 % have been achieved using the the electron beam based PBF-EB/M process [163]. It was demonstrated that these cracks can be prevented by preheating the powder bed in the PBF-EB/M process and thus lowering the temperature gradients during the cooling and thus the resulting residual stresses. This ensures that the tungsten is processed at a temperature above the ductile-brittle transition threshold. In addition, the high temperatures lead to significantly more mobile dislocations, which is also beneficial [203]. However, this required the entire powder bed to be heated to 1000 °C [163, 186, 203–205].

AM technology has the potential to be used for near-net shape manufacturing of

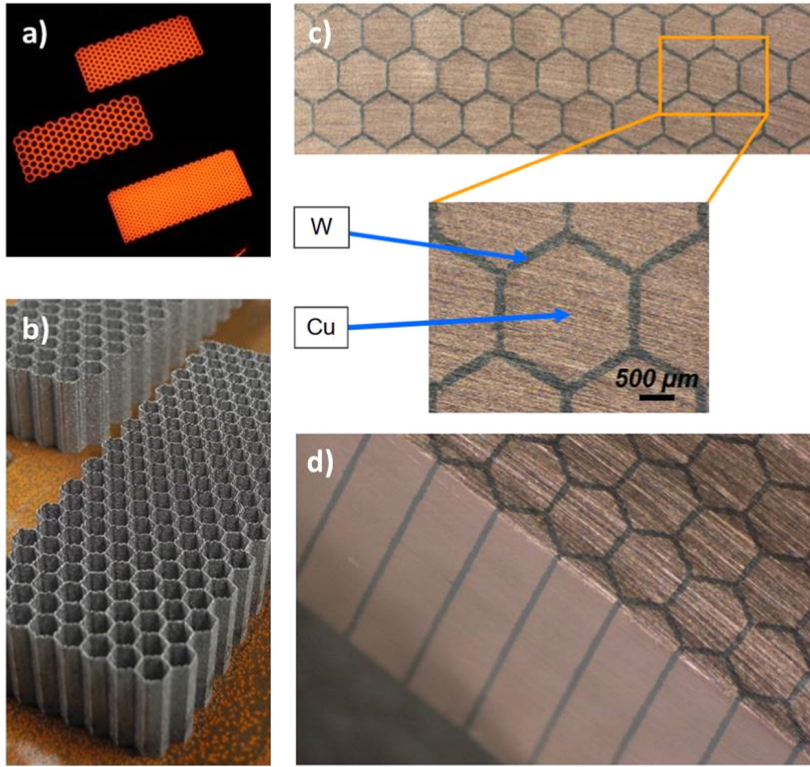


Figure 3.5: A) shows tungsten components with honeycomb structures of different dimensions during the PBF-LB/M process at substrate heating of up to 1000 °C. B) shows the finished honeycomb structure components. In C) and D), various cross-sections of the copper-infiltrated tungsten honeycomb structure are depicted [193,200].

components, thereby saving material and reducing costs. In fig. 3.6, an additive manufactured monoblock using the PBF-EB/M process with over 99 % density and hardness of recrystallized tungsten is imaged. Tungsten manufactured using the PBF-EB/M process was tested under ITER-like steady state (10 MW m^{-2}), combined with transient (10^5 pulses with 0.14 GW m^{-2}) heat loads and showed no macroscopic failure even after this exposure [205].

Moreover, the geometric freedom offered by AM can be used to manufacture advanced cooling structures that cannot or only with great difficulty be realized using conventional manufacturing processes. Improved versions of the hypervapotron that use fins to create turbulence in the cooling water, which could lead to an increase in cooling performance, can be manufactured using additive manufacturing processes [163,206,207]. Furthermore, AM can be used to produce heat sink structures

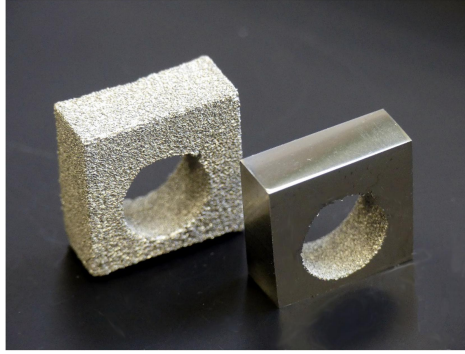


Figure 3.6: Depicted here are two ITER-type tungsten monoblocks ($28 \times 28 \times 12 \text{ mm}^3$) manufactured using the PBF-EB/M process. The one on the left is as fabricated and the one on the right is after grinding and polishing. This is an example of near net shape production of tungsten components using additive manufacturing [205].

with cooling channels that are specially adapted to the thermal conditions of the PFCs with organic forms that are able to reduce the thermally induced residual stresses in the component. Further, AM could be utilized to create a geometric gradation between the PFM and the heat sink material, thereby ensuring a smooth transition of the coefficient of thermal expansion in the joint. Thus, there are a variety of possible applications for AM technology to be used to manufacture advanced PFCs. However, it remains to be investigated how these additively generated materials and components behave when exposed to neutron irradiation [204].

3.4 Direct energy deposition

In addition to the PBF processes, the direct energy deposition (DED) processes are classified as AM technologies [160]. There are further AM processes for polymers, yet this material category is entirely unsuitable for applications involving high temperatures, such as the plasma-facing components of fusion reactors, and will thus not be discussed here. DED processes enable the local deposition of material on even or uneven substrate and the creation of arbitrary structures through line by line deposition of metallic material [208]. Since the size of components manufactured using DED is not limited by the restricted building space as it is with PBF processes, larger components can be manufactured using this method. However, it should be noted that the structures that can be realized using DED processes are less complex than those achievable with PBF processes.

DED can be used, for instance, to create near-surface cooling channels to extend

the service life of metal-forming tools [208]. DED coatings of functional materials can further increase the service life of components. Furthermore, it is an excellent method for repairing damaged components. By deposition of new material, erosion or wear damage can be compensated, and thus the service life of the components extended. In the past, such repairs were done by conventional welding processes, which, however, have the disadvantage of a larger heat affected zone (HAZ). DED has already proven its repair and remanufacturing potential on automotive and aerospace components [209–212]. With regard to material savings, near-net-shape titanium components are already being commercially manufactured by DED for use in the aerospace industry [213]. PFCs of nuclear fusion reactors also fall into the category of highly stressed components that, dependent on their position in the reactor, suffer unavoidable erosion damage of varying severity [7, 9, 43, 120, 214].

In the following, four DED processes are described and their suitability for the in situ repair of tungsten armored PFCs is discussed. Table 3.1 compares various characteristics of the electron beam additive manufacturing (EBAM) process, the powder-based laser metal deposition (LMD-p) process, the wire-based laser metal deposition (LMD-w) process and the wire arc additive manufacturing (WAAM) process.

Table 3.1: Different qualitative and quantitative characteristics of direct energy deposition (DED) processes are compared here [215].

DED-process	Deposition rate [kg h ⁻¹]	Material efficiency	Precision	Process stability	Thermal influence	Process system complexity
EBAM (wire)	5-18	High	High	High	Low	High (Vakuum)
LMD-p (powder)	1-2	Medium	High	High	Low	Medium
LMD-w (wire)	0.5-1	High	High	Low	Low	Medium
WAAM (wire)	9-10	High	Low	High	High	Low

In the LMD-p process, the energy is applied to the substrate by a focused laser beam to create a melt pool. At the same time, the feeder material (in the case of LMD-p, powder) is sprayed into the laser beam and onto the melt pool and melted. This is how material is deposited and welding beads are created by movement of the deposition unit relative to the substrate [216, 217]. The deposition rate of the

sprayed powder depends on several parameters, such as the powder feed rate, the carrier gas flow rate, the diameter distribution of the powder particles, the beam profile of the laser, the scanning velocity of the laser and others [208, 218]. The highest reported deposition rate or material efficiency achieved is almost 90 %, but is significantly lower for most processes [208, 213, 218]. Since contamination by high Z elements such as tungsten must be avoided as they strongly cool the plasma, powder-based processes such as LMD-p can be ruled out for in-vessel repairs [10]. Therefore, the wire-based processes with almost perfect material efficiency are the better choice for in vessel repairs.

For wire-based processes the wire diameter typically ranges between 0.1 and 1.2 mm [219–222]. Small wire diameters enable precise deposition of material and the development of filigree structures, while larger wire diameters offer the advantage of higher deposition rates. For wire-based DED processes, the mass flow rate $\dot{m}$ [kg h⁻¹] can be calculated as follows [222–224]:

$$\dot{m} = v_s \pi r_{wire}^2 \rho [\text{kg h}^{-1}] \quad (3.6)$$

Here, v_s is the scanning velocity [m s⁻¹] of the deposition unit, r_{wire} the wire radius [mm], and ρ the density of the deposited wire material [kg m³]. The ratio of wire feed velocity to scanning velocity can be specified as the dimensionless K value [222–228].

$$K = \frac{v_w}{v_s} \quad (3.7)$$

At the same energy density, the height of the K value can influence the density of defects, such as porosity, and is therefore a relevant metric in the context of LMD-w investigations [229].

In the wire arc additive manufacturing (WAAM) process, wire material gets molten by an electric arc and deposited on the substrate. The effective energy input P_e for the WAAM process, excluding losses due to heat conduction, radiation and convection, can be described by the following equation [230]:

$$P_e = \eta_{arc} U I_a [\text{W}] \quad (3.8)$$

Where η_{arc} is the arc efficiency, U_a the arc voltage [V] and I_a the arc current [A] [231]. The WAAM process is characterized by high deposition rates and low equipment costs, as well as high material efficiency [167, 232]. Published works have already demonstrated that the WAAM process can be used to deposit tungsten of comparatively high quality, particularly in the front feeding configuration [233, 234]. The deposited tungsten showed a high density of 99.4 %, few grain boundaries and high purity, which led to high thermal conductivity [235]. However, since higher deposition accuracies can be achieved with the other two wire-based DED processes (EBAM and LMD-w), the choice of the most suitable process for repair of PFCs

does not favor the WAAM process. Furthermore, the HAZ is larger in the WAAM process compared to EBAM and LMD-w.

In the wire-based electron beam additive manufacturing (EBAM) process, a high-energy electron beam is used as the energy source. By feeding wire into the melt pool and by movement of the deposition unit relative to the substrate, material is deposited and welding beads are generated. The EBAM process is characterized by high deposition rates, a very high material efficiency inherent to wire-based processes, as well as high process stability and precision [236–238]. The necessity for the EBAM process to take place in a vacuum increases the complexity of the process, but eliminates the influence of atmospheric contamination. The processing of tungsten with the EBAM process has also been successfully demonstrated, with overall high quality, high densities of over 99 %, and no macrocracks [239]. The EBAM process is therefore a viable candidate for the in situ repair of PFCs, assuming that the vessel remains under vacuum during maintenance and that there is sufficient space inside the vessel for the electron gun. In addition, the multipurpose deployer (MPD) needs to be able to support the weight of a electron beam gun end-effector. Therefore, the electron beam gun would have to be as small and light as possible. Since the fulfillment of these requirements cannot currently be ensured for future FPPs, the LMD-w process, which can operate without vacuum and in smaller space, was chosen as the most suitable process for in situ repairs.

The wire-based laser metal deposition (LMD-w) process uses a focused laser beam as a heat source to melt the substrate and the added wire material. By movement of the deposition unit relative to the substrate, welding beads are created and, by arrangement of these welding beads in a planned way, complex geometries can be generated. The LMD-w process is characterized by an, for wire-based processes, inherently high material efficiency, high precision and comparatively smooth surface [23, 240, 241]. Laser-based processes offer the advantage to locate the laser source outside of the vacuum vessel and to guide the electromagnetic radiation to the manipulator and the point of operation via glass fibers. As a result, the deposition unit can be used in more narrow regions of the vacuum vessel and the MPD does not have to fulfill such demanding requirements in terms of lifting capacity due to the reduced weight of the LMD-w end-effector. Furthermore, in contrast to the EBAM process, this laser-based process does not require vacuum, just the localized use of an inert gas flow. This would enable the PFCs to be repaired without a vacuum. These are the reasons why the LMD-w process was selected as the most suitable process for the in situ repair of PFCs. However, compared to the previously described DED processes, LMD-w is the least stable process, which is partly due to the requirement for precise positioning. The following section will discuss the LMD-w process and its inherent characteristics in more detail.

3.5 Wire-based laser metal deposition

This section is dedicated to the peculiarities of the LMD-w process for lateral wire feeding and to provide a deeper understanding of the process. A major challenge in the LMD-w process is process stability, which is highly dependent on the positioning of the wire in relation to the melt pool. To avoid process disruptions, the wire should always be fed precisely towards the center of the melt pool. Small unavoidable deformations of the wire that can be traced back to the production of the wire can cause lateral oscillations (see fig. 3.8) of the wire and disrupt the correct melting of the wire [242,243]. For lateral wire feeding, it is possible to choose between the front and rear feeding configurations as schematically illustrated in fig. 3.7. The position of the wire at the start of the process in relation to the focus point of the laser on the substrate also has a major influence on the process results. In the front feeding configuration, favorable process results can usually be obtained by positioning the wire at the leading edge. In this way, the wire melts by conduction from the energy dissipated from the melt pool. In the rear feeding configuration, usually superior results are obtained when the wire is positioned at the trailing edge. Thus, the wire is fed in line with the movement of the melt pool and away from the location where solidification begins [244]. Inadequate process configuration can lead to the errors described in table 3.2. These errors do not directly cause a process to terminate, but they do lead to unacceptable results. The errors listed in table 3.3, on the other hand, require the process to be terminated immediately.

An established method for comparing the energy input in metal-based direct energy deposition (DED) and additive manufacturing (AM) processes in general is to specify the applied energy density. A commonly used and reasonable approximation is the projected area energy density E_a [J mm⁻²] [222,246]:

$$E_a = \frac{P}{d_s v_s} [\text{J mm}^{-2}] \quad (3.9)$$

Where P is the laser power [W], d_s the diameter of the laser spot [mm] and v_s the scanning velocity [mm min⁻¹]. For improved comparability between studies with different spot geometries and sizes, the energy density E_n [J mm⁻²] is calculated as follows [247]:

$$E_n = \frac{P}{v_s \frac{A_s}{d}} [\text{J mm}^{-2}] \quad (3.10)$$

Where A_s is the area of the laser spot [mm²] and $d_{||}$ is the spot diameter [mm] parallel to the scanning direction.

Prior to conducting experiments involving a material with melting points as high as tungsten, it is useful to calculate analytically the minimum amount of energy

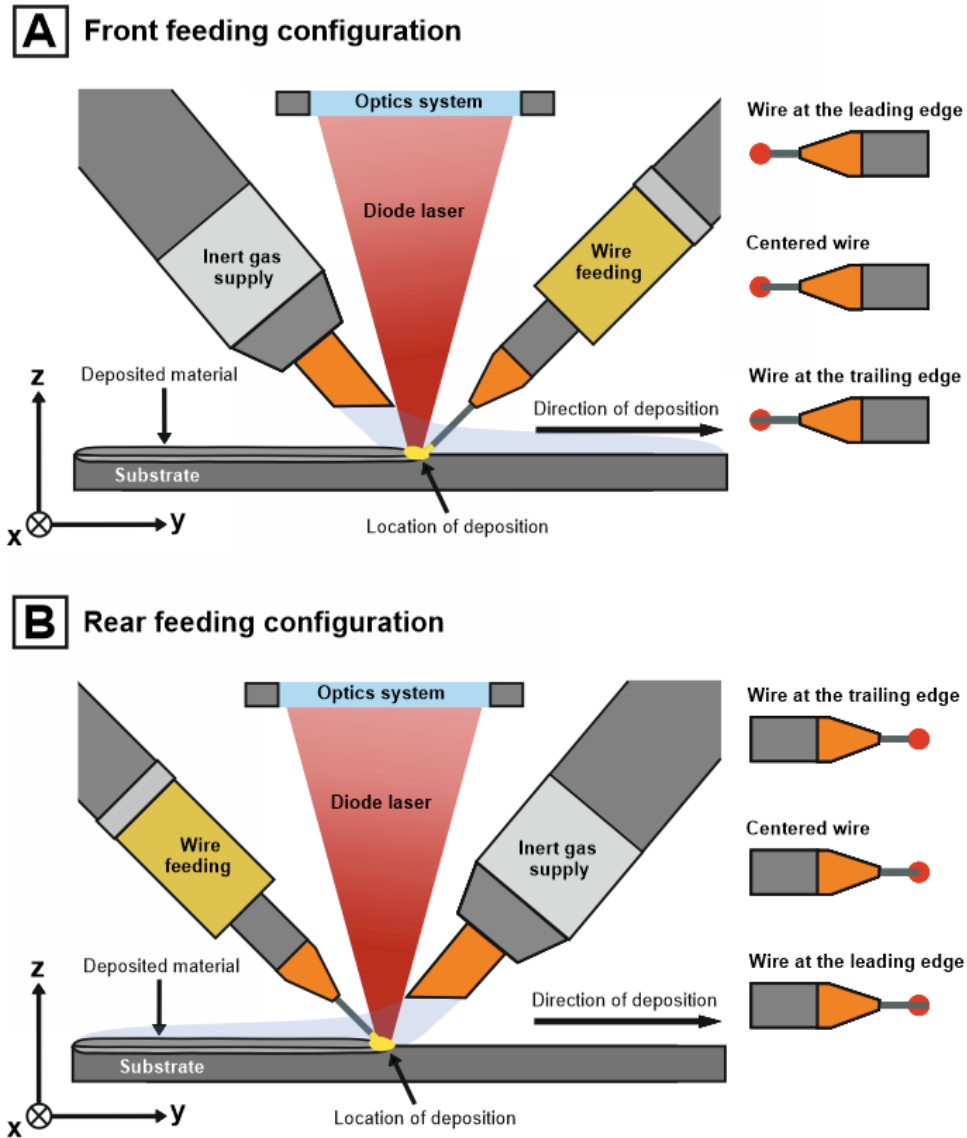


Figure 3.7: Illustrated are schematic representations of the wire-based laser metal deposition (LMD-w) process A) in the front feeding configuration and B) in the rear feeding configuration. The position of the wire in relation to the laser beam and the processing plane strongly influences the process results. On the right, the possible positions of the wire relative to the focus spot of the laser on the substrate are shown for both configurations.

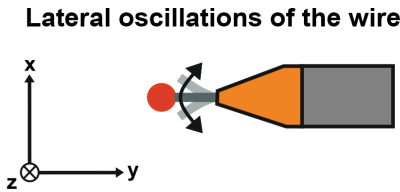


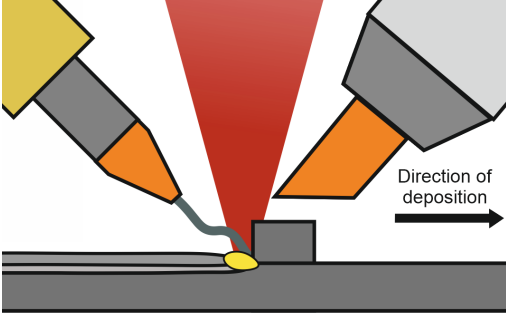
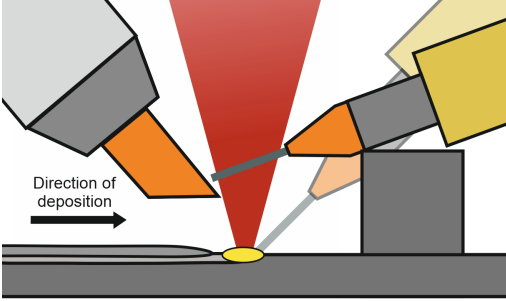
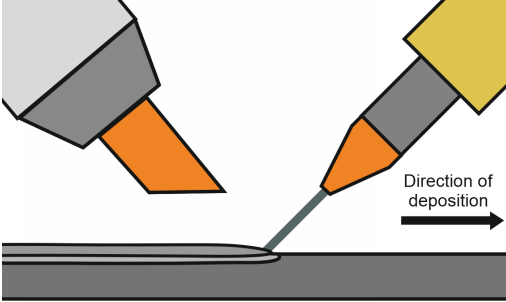
Figure 3.8: Lateral oscillations of the wire at the wire nozzle can be caused by inherent straightness irregularities in the wire material. These deflections can slide the wire outward out of the laser spot center, resulting in process instabilities caused by uneven wire melting.

Table 3.2: Process errors that lead to inadequate process results [243, 245].

Schematic illustration	Description
	Balling defects arise due to an discontinuous transition from wire material to substrate. These defects occur when the geometric alignment of wire, laser beam and substrate is inadequate.
	Excessive wire feed in relation to the melting rate results in contact between the still solid wire and solid substrate. In this case, the wire scratches through the melt pool. Elastic bending of the wire, which can lead to lateral deflection, causes process instabilities. The result is a bumpy welding bead surface.

required to melt the material. However, this is not the energy required for a stable process and should only be seen as an estimate of the minimum intensity required for melting. Using the following equation, the minimum intensity required for a

Table 3.3: Process errors that lead to abortion of the process [243, 245].

Schematic illustration	Description
	The collision of the wire with a structure causes it to bend and be deflected out of the laser beam. If the wire feed continues, the wire will buckle or break. As a result, the deposition unit can be damaged. This defect occurs due to faulty movement, inadequate wire alignment or insufficient melting rate.
	Collision of the wire feeding nozzle with a structure results in deflection of the wire and abortion of the process. The deposition unit may suffer damage in such an event.
	If the wire leaves the laser beam, it is possible that it remains connected to the welding bead as the material solidifies. This can occur in particular due to lateral deflection or insufficient separation of the wire from the weld pool at the end of the weld bead. Further movement can cause damage to the deposition unit.

laser beam to melt specific materials can be approximated [248, 249]:

$$I_{th_{min}} = \frac{\kappa(T)(T_{melt} - T_0)}{a(\lambda_w, T)d_s J_{max}} [\text{W mm}^{-2}] \quad (3.11)$$

Where $\kappa(T)$ is the temperature dependent thermal conductivity of the material [$\text{W m}^{-1} \text{K}^{-1}$], T_0 its initial temperature and T_{melt} the melting temperature [K]. $a(\lambda, T)$ is the wavelength and temperature dependent absorption coefficient of electromagnetic radiation for the specific material. J_{max} is a dimensionless parameter

that represents a measure of the distribution of thermal energy in the material at a certain scanning velocity and spot area. For small beam focus diameters at low scan velocities a J_{max} value of 0.52 can be assumed [248, 249].

For intensities between $I_{th_{min}}$ and $I_{th_{max}}$, the welding process is in conduction mode [247]. At intensities higher than $I_{th_{max}}$, the vaporization of the material causes the transition to keyhole mode [249]. The conduction mode and the keyhole mode were previously discussed in section 3.2 (see figure 3.2 B) and C)). In contrast to the conduction mode, the keyhole mode is characterized by a deeper melt pool and continuous homogeneous evaporation of material in the melt pool. Therefore, the evaporation temperature is set as the target temperature in the formula for $I_{th_{max}}$. For long dwelling times, the intensity $I_{th_{max}}$ is calculated as follows [250]:

$$I_{th_{max}} = 4\sqrt{\frac{2}{\pi}} \frac{k(T)T_v}{a(\lambda, T)d_s} [\text{W mm}^{-2}] \quad (3.12)$$

Where $\kappa(T)$ is the temperature dependent thermal conductivity of the material [$\text{W m}^{-1} \text{K}^{-1}$] and T_v its vaporization temperature [K]. $a(\lambda, T)$ is the wavelength and temperature dependent absorption coefficient of electromagnetic radiation for the specific material and d_s the diameter of the laser spot [mm].

4 Test facilities and methods

4.1 LMD-w facility and diode laser

The following describes the facility setup that was utilized for the LMD-w experiments. The electromagnetic radiation was generated by a continuous wave (CW) diode laser from Laserline GmbH (model LDF 5000-40). The laser diodes deliver a combined power of up to 5 kW and emit radiation in the infrared range with wavelengths of 910, 940, 980 and 1030 nm. The light was guided from the source to the optics of the deposition unit using glass fibers. The optics of the deposition unit are also a model made by Laserline GmbH and are operated with a water cooling system, as the high CW laser power would otherwise cause problems for the optical system. This optic focuses the IR radiation in a conical beam and projects a circular focal spot with a diameter of 2 mm on the building plane. A beam caustic measurement determined the focal length from the optics to the building plane to be 139 mm. The schematic illustration of the LMD-w process in fig. 3.7 is based

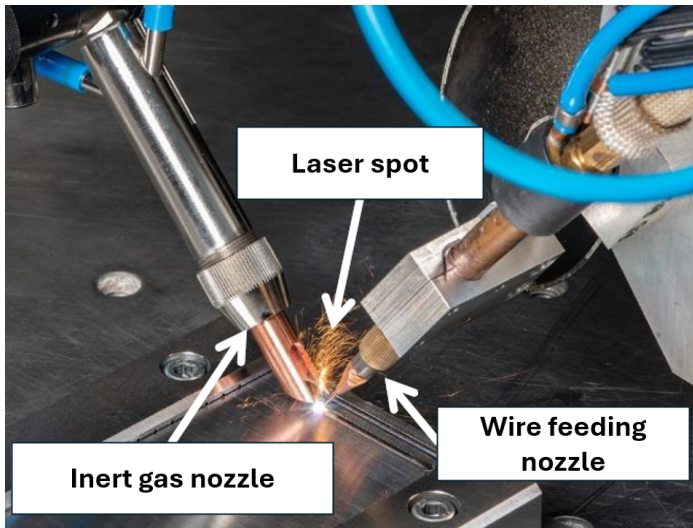


Figure 4.1: Image of the LMD-w facility [215].

on the setup that was used. Fig. 4.1 displays the LMD-w facility used in this work. To avoid uncontrolled temperature fluctuations due to heat accumulation, the substrate was cooled to 19 °C by a water cooling system. The wire feeding system by the manufacturer Abicor Binzel enables the precise supply of welding wire, with the compensation of wire curvature through a special wire straightening unit. This is important to minimize lateral oscillations of the wire, which would reduce the stability of the process (see fig. 3.8). An inert gas nozzle, which is visible in fig. 4.1, is directed at the melt pool and blows argon onto it. Thus, a constant gas flow protects the melt pool from undesired contact with the atmosphere and oxidation. The wire feeding and inert gas nozzles are water cooled to prevent undesired heat accumulation in these components. The movements of the building platform in the x- and y-directions, as well as the movement of the deposition unit in the z-direction, is performed by a computerized numerical controlled (CNC) 3-axis system with a Rexroth control unit.

4.2 Electron beam facility JUDITH 2

The electron beam facility JUDITH 2 was used for the high heat flux experiments. JUDITH stands for “JUelich DIvertor Test facility in the Hot cells“. Nuclear activated samples can be handled in hot cells. The decommissioned JUDITH 1 facility was placed in a hot cell and JUDITH 3 will be placed in a hot cell. JUDITH 2, however, is not located in a hot cell and is used for tests on non-activated samples. The JUDITH facilities are used for tests and qualification of material samples, mockups or semi-prototypes of PFCs.

Fig. 4.2 shows a schematic representation of this facility on the left-hand side and a photo on the right. JUDITH 2 consists of an electron beam gun (EH800V by Von Ardenne Anlagentechnik GmbH) that emits up to 200 kW of power and a cylindrical vacuum chamber (by Trinos Vakuum-Systeme GmbH). The resulting electron beam has a Gaussian shape and is characterized by the following intensity on the xy-plane perpendicular to the beam direction [251]:

$$I(x, y) = I_0 e^{-\frac{x^2 + y^2}{2\sigma_S^2}} [\text{W mm}^{-2}] \quad (4.1)$$

where I_0 is the intensity at the center of the beam:

$$I_0 = \frac{P}{2\pi\sigma_S^2} [\text{W mm}^{-2}] \quad (4.2)$$

σ_S is the standard deviation (here the characteristic beam width) and P the power of the electron beam gun. The electron beam diameter (at full width half maximum) varies between 3 and 15 mm depending on various parameters, but in particular the applied power. It is important to consider that only a fraction of the electrons is

absorbed, as some are reflected as backscattered electrons. The reflection factor is strongly dependent on the material, whereby $r_W = 0.45$ can be assumed for tungsten. This value comes from Monte Carlo simulations, but it is consistent with literature values of 0.38 to 0.5 determined by measurements and also with experiments in the JUDITH facilities [81, 252, 253].

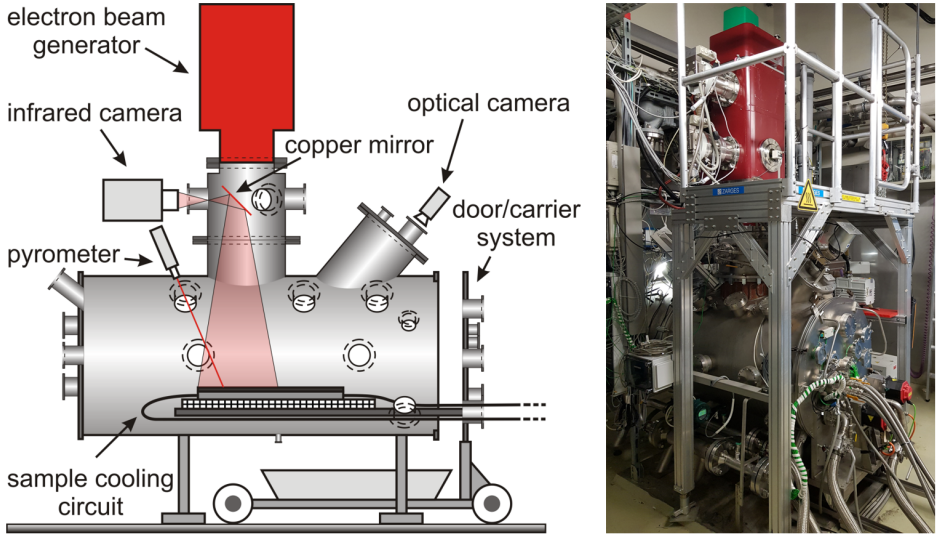


Figure 4.2: The electron beam facility JUDITH 2 is illustrated schematically on the left-hand side, with a photograph of the facility displayed on the right-hand side [80].

The penetration depth of the electrons depends on the acceleration voltage and the exposed material. At an acceleration voltage of 50 kV, which is a characteristic value for JUDITH 2, the penetration depth of the electrons in tungsten is $\sim 5 \mu\text{m}$. Reaching this depth, 95 % of the electrons are absorbed [252, 254].

The electron beam is guided by two magnetic focusing lenses and one deflecting lens that are controlled by a program following x- and y-coordinates of a command file (standard ASCII file) [255]. Whereby the dwell time per point is between $5 \mu\text{s}$ and 1 s [251]. The electron beam can be moved at high velocities along the defined path. To simulate steady state heat loads in the kW m^{-2} or MW m^{-2} range, the beam is moved in cyclically repeated grid patterns over the area to be loaded. Since the electron gun delivers a continuous beam during operation, the excess energy usually has to be placed on specially designed beam dumps. These beam dumps

are actively cooled copper structures that partially absorb the thermal energy and transfer it to the cooling water. Even transient heat loads in the GW m^{-2} range can be simulated in JUDITH 2 if the electron beam is directed at certain spots in circular patterns to achieve high power densities. Here, the beam usually targets eight points along a circle. This circle is traversed 12 times, with each orbit taking 40 μs . The targeted thermal energy is deposited in the loaded area over the total pulse duration Δt of 0.48 ms [80, 256]. In this context, information about the beam profile is necessary, which is regularly determined by designated measurements.

To measure the power acting on the tested components in JUDITH 2 during the loading process, water calorimetry is used [251]:

$$I = \frac{\Delta T \dot{V} \rho c_p}{A_s} [\text{W mm}^{-2}] \quad (4.3)$$

Where I is the applied intensity [W m^{-2}], ΔT the difference between the inlet and outlet temperature [K] of the cooling water of the exposed component, $\dot{V}$ the coolant flow [$\text{m}^3 \text{s}^{-1}$], ρ the coolant density [kg m^{-3}], c_p the coolant specific heat [$\text{J kg}^{-1} \text{K}^{-1}$] and A_s the loaded area [m^2].

The facility includes a water-based cooling system and the coolant can be operated at temperatures from room temperature (RT) to $\sim 120^\circ\text{C}$, flow rates of 200 l min^{-1} and pressures of up to 40 bar (4 MPa) [80, 257]. In this way, 150 kW can be dissipated from the exposed components. In addition, the beam dumps achieve a cooling performance of up to 150 kW and the vacuum chamber reaches 80 kW. The vacuum pressure in the main chamber is adjustable and can achieve up to 10^{-5} mbar.

4.3 Temperature measurements

The temperature measurements performed in the context of this work are mainly carried out using radiation measurements. The following is a brief overview of some important and fundamental information on radiation temperature measurements. Above 0 K, every body emits thermally excited electromagnetic radiation. For opaque solids, the radiant flux is largely determined by the surface temperature. The radiant flux Φ is defined as [258]:

$$\Phi = \frac{dQ}{dt} [\text{W}] \quad (4.4)$$

Where Q is the radiation energy [J]. The spectral specific radiation $M_{\lambda s}$ of a black body is described as follows [183, 259]:

$$M_{\lambda s} = \frac{d\Phi}{dA} = \frac{\int Q_\lambda d\lambda}{dA} \frac{d}{dt} [\text{W m}^{-2}] \quad (4.5)$$

Where Q_λ is the wavelength λ dependent radiated energy, related to the time t and the emitting area A . By integration of the spectral specific radiation $M_{\lambda s}$ of a black body over λ from 0 to ∞ , the Stefan-Boltzmann law is obtained [258]:

$$M_{ts} = \sigma_{SB} T^4 [\text{W m}^{-2}] \quad (4.6)$$

Here σ_{SB} is the Stefan-Boltzmann constant ($\sigma_{SB} = 5.67037 \cdot 10^{-8} \text{ W m}^{-2} \text{ K}^{-4}$) and T is the temperature. For non-black bodies, the following radiated power P is emitted, taking into account the temperature dependent specific emissivity $\epsilon(T)$ and the radiating surface A [258]:

$$P = \epsilon(T) \sigma_{SB} A T^4 [\text{W}] \quad (4.7)$$

Since emissivity depends on a variety of different parameters, including, for instance, the specific material of the body, its temperature and surface structure, accurate temperature measurements by radiation are challenging. A changing surface morphology (e.g. roughness, cracking, oxidation) and temperature during the measurement further complicates accurate measurements. The following equation describes an approximation of the relationship between temperature uncertainty and the uncertainty of the emission coefficient [258]:

$$\frac{\Delta T}{T} \approx - \frac{\lambda}{5\lambda_{max}} \frac{\Delta \epsilon(T)}{\epsilon(T)} \quad (4.8)$$

λ is the used spectral wavelength and λ_{max} is the peak wavelength. According to Wien's displacement law, the peak wavelength is related to the temperature of a black body as follows [260]:

$$\lambda_{max} = \frac{c_3}{T} [\text{m}] \quad (4.9)$$

The Wien's constant is $c_3 = 2.8977685 \cdot 10^{-3} \text{ m K}$.

Pyrometers were used to measure temperatures during the exposure of components in the electron beam facility JUDITH 2. On the one hand, the two-color pyrometers QKTR 1075 manufactured by Maurer GmbH were used to measure the surface temperature. For different temperature ranges, specific two-color pyrometers were used (pyrometer 1: 300-900 °C; pyrometer 2: 600-1600 °C; pyrometer 3: 1000-3300 °C). The two-wavelength method of these pyrometers ensures that the emissivity is less relevant as long as the two emissivities of the two wavelengths are close to each other (grey body assumption). Normally this cannot be assumed, but since the wavelength ranges from 1.4 to 1.75 and 1.6 to 1.75 are close to each other, the error is minimized. These pyrometers achieve time resolutions of 20 ms [80].

Additionally, a high-speed pyrometer was used to achieve a higher time resolution, particularly to measure the transient heat flux scenarios appropriately. This pyrometer is the model KMG 740-USB from the manufacturer Kleiber, which achieves

a time resolution of 10 μs in the temperature range from 400 to 3500 $^{\circ}\text{C}$. This pyrometer utilizes the wavelength range from 2 to 2.5 μm . The two-color pyrometer, which is nearly independent of emissivity, was used to set the high-speed pyrometer to the correct emissivity [80].

In addition, infrared (IR) cameras were used to monitor and visualize temperature over larger areas. The infrared (IR) camera VarioCAM® HDx from the manufacturer InfraTec was used. It achieves a resolution of 0.5 mm pixel⁻¹ with 640 $\times$ 512 pixels and operates in the wavelength range of 0.2-0.6 μm . This particular camera model was used for IR thermography at the LMD-w and electron beam facilities.

4.4 Microstructural investigation

For the microstructural examination of the specimens, different imaging techniques were used depending on what was intended to be observed. The techniques used include light microscopy (LM) and scanning electron microscopy (SEM), including the detection of secondary electrons (SE), backscattered electrons (BSE), electron backscatter diffraction (EBSD) and energy dispersive X-ray spectroscopy (EDX). In the context of this work, these techniques were used mainly for the analysis of the cross-sectional micrographs of various specimen and component types.

Preparation of the micrographs

Detailed microstructural analysis of the specimens using the mentioned imaging techniques requires appropriate preparation. The samples were separated at the section plane of interest using a precision cutting machine with water cooling. These samples were subsequently embedded under vacuum in liquid, room temperature self-curing, and non-conductive epoxy resin. The embedded samples were then ground under running water with SiC paper of the grades P240 to P2500 and successively polished with diamond suspension with particle sizes of 9, 3 and 1 μm . For analysis methods requiring particularly smooth surfaces, such as EBSD, the samples were subsequently treated with agglomerate-free oxide polishing suspension (SiO_2) with a particle size of 0.05 μm and subsequently cleaned thoroughly using water. Vibration polishing was also used in some cases to achieve exceptionally smooth surfaces. To improve the visibility of the grain structure in the micrographs, some samples were also etched. The etching agent used for tungsten is a mixture of 2.5 % potassium hydroxide (KOH), 2.5 % potassium hexacyanoferrate(III) ($\text{K}_3[\text{Fe}(\text{CN})_6]$), 25 % water (H_2O) and oxide polishing suspension (70 %). To ensure the necessary current flow from the samples during SEM analysis, an approximately 2 nm thick gold or platinum layer was coated onto the embedded samples using magnetron sputtering.

Light microscopy

Light microscopy (LM) has a lower maximum resolution than selective electron mi-

croscopy (SEM). The useful magnification that can be achieved by LM is limited not only by the characteristics of the specific device but also by the wavelength of the light. The maximum resolution d_R that is achievable in a perfect optical system can be described using Abbe's equation [261]:

$$d_R = \frac{0.612\lambda}{n_r \sin \theta} [\text{m}] \quad (4.10)$$

Where λ is the wavelength of the imaging radiation, n_r refractive index of the medium between the point source and lens, relative to the free space. And θ is half of the angle of the maximum cone of light collected by the lens. $n_r \sin \theta$ is often called the numerical aperture (NA). LM was selected primarily for rather large-area images and for higher magnifications, SEM was selected. A Nikon Eclipse LV100D microscope was used for the LM observations.

Detection of secondary electrons

For high-resolution imaging, the scanning electron microscopy (SEM) was used. For SEM, an electron beam is directed at the sample and, in the case of SE, the secondary electrons are detected. When the beam of primary electrons strikes the atoms of the sample, the atoms emit low-energy secondary electrons. These secondary electrons can be collected and detected. Since the secondary electrons have a low energy, only the electrons from the top few nanometers of the sample can leave the surface [262]. Thus, in the case of SE, it is mainly the upper few nanometers that are represented. The Zeiss Crossbeam 540 field emission scanning electron microscope (SE, BSE, EBSD, EDX) and the Zeiss LEO DSM 982 (SE, BSE, EDX) were used for SEM imaging.

Detection of backscattered electrons

In backscattered electron (BSE) detection, the primary electrons of the beam are elastically scattered by atoms of the sample. The elastically scattered primary electrons, which have retained most of their original energy are considered backscattered electrons and are able to escape from deeper regions of the sample and be detected by the sensor placed above the sample. Thus, BSE provides information from deeper lying regions compared to SE detection. As a result, BSE is less effected by thin surface contaminations. Furthermore, BSE is characterized by a lower spatial resolution, since electrons are detected from larger volumes. BSE has the advantage of sensitivity to the atomic number of the impacted atoms and can resolve different phases or chemical compositions [263].

Electron backscatter diffraction

In electron backscatter diffraction (EBSD), the sample is tilted so that the electron beam hits the sample at an angle. The backscattered electrons are scattered at slightly different angles depending on the crystal alignment, and this can be detected. The resulting electron backscatter diffraction patterns are analyzed and

provide information about the distance between crystal planes and the crystal orientation [263].

Energy dispersive X-ray spectroscopy

Energy dispersive X-ray spectroscopy (EDX) can be used to perform elemental analysis or chemical characterization of a sample. The utilized SEM microscopes have an integrated EDX system. For EDX investigations, an acceleration voltage of 15 kV was used for the primary electron beam. If the primary electron beam hits the sample surface, X-rays with characteristic wavelengths dependent on the specific material are emitted. These X-rays can be detected with an EDX detector. This allows local resolution of the elemental and chemical composition of the microstructure [264, 265].

4.5 Profile measurements

Laser profilometry was used to characterize the surface of the specimens. A profilometer from OPM GmbH equipped with a KF3 confocal sensor was used for the profilometric measurements performed in this work. This device achieves a lateral resolution of 2000 points mm^{-1} with a vertical resolution of 0.02 μm . Here, profilometry was used to characterize the surface of tungsten layers generated by LMD-w. In addition, the surface structure of the samples was compared before and after exposure to high heat fluxes. Surface changes, such as cracks, can be detected in this way. Furthermore, 3D images of the surface can be created.

To quantify the roughness or waviness of samples, different parameters are used depending on which aspect should be analyzed. The arithmetic average height parameter R_a is usually used to quantify the surface roughness [266]:

$$R_a = \frac{1}{l} \int_0^l |y(x)| dx [\mu\text{m}] \quad (4.11)$$

Where l represents the length over which the roughness is measured and $y(x)$ represents the deviation of the surface profile from the mean line of the surface profile.

In comparison to R_a , the R_q value is more sensitive to large individual deviations from the profile mean value such as cracks. The R_q value is described as follows [266]:

$$R_q = \sqrt{\frac{1}{l} \int_0^l \{y(x)\}^2 dx} [\mu\text{m}] \quad (4.12)$$

The W_t parameter can be used to quantify the waviness of surfaces. In the context of this work, the W_t parameter is used to quantify the waviness of layers produced

by LMD-w. This parameter is calculated from the difference between the highest and lowest point of the low-pass filtered profile within the length l . In this case, the filtered profile is used to minimize the influence of roughness on the waviness value.

These roughness and waviness parameters can be measured not only for individual lines but also for surfaces and are in this case referred to as sR_a , sR_q and sW_t .

4.6 Hardness measurements

To quantify the hardness of the material samples, HV1 Vickers hardness tests were carried out in accordance with the DIN EN ISO 6507-1 standard [267]. The hardness tests were performed using a Qness 10/60 A+ from the manufacturer ATM Qness GmbH. The hardness of a material is determined by pressing a pyramid-shaped diamond indenter onto the sample and measuring the resulting indent. The diamond pyramid is pressed onto the material for a specified period of time (10-15 s) with a specified force. For HV1 hardness tests, the test force of 9.807 N is applied. The angle of the opposite faces at the vertex of the pyramid-shaped indenter is 136° . The diagonals d_1 and d_2 of the indent are measured and d_v is determined:

$$d_v = \frac{d_1 + d_2}{2} [\text{m}] \quad (4.13)$$

The HV value is determined from a constant times the ratio of the test force F and the surface area of the indent:

$$HV = \frac{0.102F2 \sin(\frac{136^\circ}{2})}{d_v^2} \quad (4.14)$$

To ensure that in the case of multiple indents on the same sample, the individual measurement indents do not influence each other, measurements are usually taken at a minimum distance equal to three times the diagonal length of the indents.

4.7 Finite element temperature calculations

The majority of physical phenomena can be expressed in the form of partial differential equations with appropriate boundary and initial conditions [268]. Finite element method (FEM) is used here to solve heat transfer with differential equations in three-dimensional space. The governing equation for heat conduction through a solid in a cartesian coordinate system is the general heat equation [269]:

$$\kappa \left(\frac{\delta^2 T}{\delta x^2} + \frac{\delta^2 T}{\delta y^2} + \frac{\delta^2 T}{\delta z^2} \right) + q = \rho c_p \frac{\delta T}{\delta t} \quad (4.15)$$

or:

$$\kappa \nabla^2 T + q = \rho c_p \frac{\delta T}{\delta t} \quad (4.16)$$

Where κ [$\text{W m}^{-1} \text{K}^{-1}$] is the thermal conductivity, T [K] the temperature, q [W] the rate of heat flux/convection/radiation/internal heat generation inside the volume, ρ [kg m^{-3}] the density of the material, c_p [$\text{J kg}^{-1} \text{K}^{-1}$] the specific heat capacity and t [s] the time. The temperature dependence of material-specific data such as thermal conductivity was taken into account in these simulations.

FEM simulations were performed to investigate the thermal influence of the moving laser beam on the tungsten substrate. To perform the simulations, Ansys 2020 R1 software with the moving heat source V4.1 extension was used. The power loss due to radiation, as described in the Stefan-Boltzmann law (equation 4.7), was taken into account. In addition, heat loss due to conduction at the points where the tungsten plate is attached to the cooling plate has been taken into account. Temperature profiles were generated in the tungsten as it is irradiated by the moving beam. The dimensions of the resulting melt pool were of particular interest.

In the context of the high heat flux experiments in the electron beam facility JUDITH 2, thermal stationary FEM simulations were carried out. In this way, crucial information was obtained in advance to enable the design of the correct thermal loading conditions, taking into account the water cooling parameters. For this purpose, the stationary thermal module of the ANSYS software was used. The heat flux when using certain parameters of the water cooling system was determined through a Nukiyama code developed by ITER. In this code, the convective heat transfer coefficient h_{cv} [$\text{W m}^{-2} \text{K}^{-1}$] is calculated as follows [270]:

$$h_{cv} = Nu \frac{\kappa}{D_h} [\text{W m}^{-2} \text{K}^{-1}] \quad (4.17)$$

Where Nu is the dimensionless Nusselt number, κ is the thermal conductivity [$\text{W m}^{-1} \text{K}^{-1}$] and D_h is the hydraulic diameter $D_h = 4 A p_w^{-1}$. Here, A is the cross sectional area [m^2] of the duct and p_w is the wetted perimeter [m]. The Nusselt number Nu describes the ratio of convective and conductive heat transfer. According to the Sieder-Tate equation, which the code uses to calculate the Nusselt number, Nu can be approximated as follows [271]:

$$Nu = 0.027 Re^{\frac{4}{5}} Pr^{\frac{1}{3}} \left(\frac{\mu_{db}}{\mu_{dw}} \right)^{0.14} \quad (4.18)$$

Where μ_{db} is the dynamic viscosity [$\text{kg m}^{-1} \text{s}^{-1}$] of fluid at bulk temperatures and μ_{dw} at wall temperatures. Re is the dimensionless Reynolds number. Re is ratio of inertial and viscous forces [270]:

$$Re = \frac{v_m D_h}{\mu_k} \quad (4.19)$$

Here, v_m is the mean fluid velocity [m s^{-1}] and μ_k is the kinematic viscosity [$\text{m}^2 \text{s}^{-1}$] ($\mu_k = \mu_d \rho^{-1}$). With μ_d being the dynamic viscosity [$\text{kg m}^{-1} \text{s}^{-1}$] and ρ the density

$[\text{kg m}^{-3}]$ of the fluid. The dimensionless Prantl number Pr is the ratio of momentum and thermal diffusivity [270]:

$$Pr = \frac{\mu_k}{\alpha_{td}} = \frac{\mu_d c_p}{\kappa} \quad (4.20)$$

Where μ_k is the kinematic viscosity $[\text{m}^2 \text{s}^{-1}]$, α_{td} the thermal diffusivity $[\text{m}^2 \text{s}^{-1}]$, μ_d the dynamic viscosity $[\text{kg m}^{-1} \text{s}^{-1}]$, c_p the specific heat capacity $[\text{J kg}^{-1} \text{K}^{-1}]$ and κ the thermal conductivity $[\text{W m}^{-1} \text{K}^{-1}]$.

4.8 ERO2.0 code

The ERO2.0 Monte-Carlo code was used, in the context of this work, to investigate the influence of macroscopic surface waviness of PFCs on their erosion, the deposition of impurities and the resulting heat fluxes. This code is able to perform simulations of PWI on the entire inner wall of fusion reactors and simulate global impurity (e.g. eroded wall material) transport. The general workflow of ERO2.0 code is illustrated in fig. 4.3 [272]. These global simulations are enabled by massive

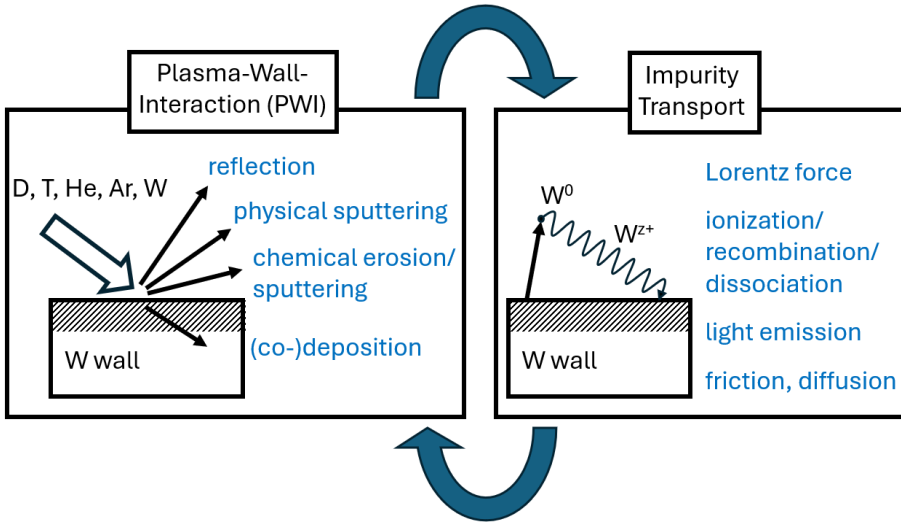


Figure 4.3: The general workflow of the ERO2.0 code based on the example of a tungsten surface exposed to DT plasma.

parallelization of the code, which allows for the efficient use of high-performance supercomputers. The code can be executed on a global, local and micro scale. For the investigations carried out in the context of this work, the scale local was chosen so that the use of powerful supercomputers was not required.

The plasma parameters of the specific fusion reactor, obtained by means of codes such as SOLPS or by measurements, are required as input. For the simulations performed here, plasma background parameters generated by SOLPS for the divertor area of the European DEMO FPP at the outer vertical target were used. These plasma parameters usually include the electron density, the electron temperature, the ion temperature, the magnetic field and the angle of incidence of the plasma on the wall. In addition, the program needs to be provided with the geometry of the inner wall. For the simulations carried out as part of this work, as the wall, a simple 3D geometry was chosen to represent a single welding bead. This geometry was duplicated to create the wavy topology of an unsmoothed LMD-w layer. This allows the effect of the leading edges to be simulated and studied. But more details are given in the designated section 5.2. For reasons of complexity, the physically correct simulation of the gyro orbits of the charged particles was discarded in these simulations. Instead, the particles move strictly along the magnetic field lines.

5 Adaptation of the LMD-w process for tungsten

5.1 Process development for tungsten

To enable the compensation of eroded PFM and repair of damaged armor of PFCs, the LMD-w process must be adapted to process tungsten. On the one hand, tungsten is currently the most promising candidate for the PFM of future FPPs due to its characteristic properties as a refractory metal. In this context, tungsten is particularly suitable due to its high thermal conductivity ($\kappa \approx 183 \text{ W m}^{-1} \text{ K}^{-1}$ at RT), high melting point ($T_{\text{melt}} = 3422 \text{ }^{\circ}\text{C}$), low tritium inventory and comparatively low erosion rate under plasma exposure [8,9,17,81,106]. However, on the other hand, its high melting point and brittle nature, especially at low temperatures, make tungsten one of the most challenging materials to process by LMD-w. The following describes the process development for the deposition of tungsten on tungsten substrate using LMD-w. Here, the aim is to be able to deposit entire layers on the tungsten substrate by lining up the welding beads.

Experimental

In the LMD-w process development studies for tungsten, the setup described in detail in section 4.1 was used. This LMD-w setup depicted in fig. 4.1 and schematically illustrated in fig. 3.7.

The deposition experiments were performed using commercially available tungsten wire with a diameter of 1 mm from Osram Licht AG. For the in situ repair of PFCs, the substrate material will be the armor in the form of the PFM. Therefore, rolled tungsten plates were used in the deposition experiments to represent the PFM. These tungsten plates were machined from larger tungsten blanks using electrical discharge machining (EDM). To remove surface contamination and Cu residues of the EDM process from the tungsten plates, these were etched and subsequently cleaned.

The aim of adapting the LMD-w process to tungsten is the successful deposition and thereby the identification of suitable process parameters. However, tungsten presents some obstacles compared to the materials conventionally processed by

LMD-w. Due to the high melting point of tungsten, the energy required to successfully weld the material is particularly high. In the case of a thermally insulated substrate plate, the high energy input required to weld tungsten would cause the substrate plate to heat up and build up of temperatures during the process. Since the oxidation protection in the form of the inert gas stream ensures only local protection in the region of the melt pool and not for the entire substrate plate, these unprotected areas would suffer from oxidation (see fig. 5.1 A)) [106,273]. To keep

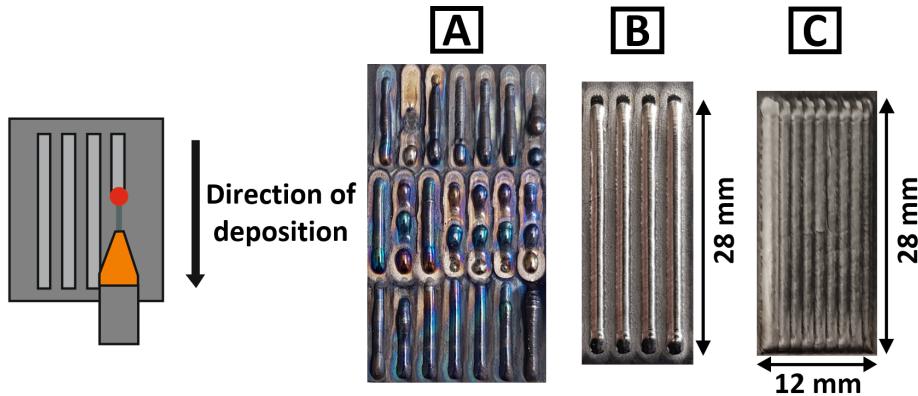


Figure 5.1: Depicted are samples demonstrating the evolution of the process development for tungsten in the form of welding beads and a layer generated by LMD-w. A) depicts welding beads from preliminary studies, which contain a variety of defects. B) shows welding beads from a more advanced study to create higher quality welding beads. In C), several welding beads were arranged next to each other to coat an area of $28 \times 12 \text{ mm}^2$ with tungsten [169].

the substrate temperature reasonably constant and to avoid undesired oxidation, the tungsten substrate plate was fixed by two screw fastened copper plates to a water cooled (19°C) copper heat sink plate.

The LMD-w process is influenced by a variety of factors. On the one hand, the positioning of the wire in relation to the laser beam and the processing plane, as discussed in section 3.5, plays an essential role in the stability of the process. On the other hand, factors such as laser power, scanning velocity, wire feed velocity, wire position, wire feeding angle, laser spot diameter, substrate temperature, ambient temperature, inert gas flow rate, deposition direction, etc. play a role. The interdependence of these parameters further increases the inherent process complexity. The results of the first preliminary studies on the deposition of individual tungsten welding beads on a tungsten substrate (see fig. 5.1 A)), which were defective in many respects, show that the development of a stable, high-quality process is by no means trivial. Due to the high number of influencing parameters, an overarching cause-

and-effect study would require a unreasonably high number of trials. Therefore, the frame of eligible parameters was determined using calculations and preliminary studies. In this identified range, more extensive studies were subsequently conducted to further optimize the process. Thereby, only the parameters with the major influence on the process result, such as laser power, scanning velocity, wire feed velocity, hatch distance and wire position, were varied to examine their respective influence. The other relevant process factors were set to empirically determined reasonable settings.

In preliminary studies, the tungsten process in the front feeding (i.e., dragging) configuration (see fig. 3.7) showed slightly superior results in direct comparison with the rear feeding (i.e., piercing) configuration. Therefore, the following experiments were performed in the front feeding configuration. Since the process performed well at 45 degrees wire feeding angle, this configuration was used for the further experiments. The experiments were conducted under standard atmospheric conditions, with a local flow of inert gas directed at the melt pool. Argon was used as the inert gas to protect the melt pool from contamination during the process. The gas flow rate should be selected so that a laminar flow of the inert gas is achieved without much turbulence. A gas flow rate of 15 l min⁻¹ showed favorable results and was therefore set as standard. Between the deposition of the individual welding beads, the program included a waiting period of 10 seconds, allowing the substrate plate to reach the starting temperature and thus ensuring the same initial temperature for all welding beads. In order to achieve the lower threshold for the required energy input, the minimum intensity $I_{th_{min}}$ to melt the material was determined according to equation 3.11. For the melting of tungsten with the specified setup, $I_{th_{min}}$ is determined with $\Delta T = 3403$ K, an assumed absorption coefficient of 0.42, and J_{max} value for small focal point diameters of 0.52 [106, 248]. Thus, according to equation 3.11, in this context the minimum intensity required to melt tungsten is 1.25 kW mm⁻². For the laser focus diameter of 2 mm, the necessary power to melt tungsten is 3.9 kW, which is in line with the empirically determined lower threshold of 4 kW (see table 5.1). Assuming a scanning velocity of 800 mm min⁻¹, the laser irradiates one location in its path for 0.15 s. This duration multiplied with the intensity $I_{th_{min}}$ of 1.25 kW corresponds to a required minimum energy density of 191 J mm⁻² necessary for the laser-induced melting of tungsten.

Fig. 5.1 B) pictures the results of the more advanced experiments on individual welding beads. Table 5.1 (welding beads) delivers details of the range of process parameters that were used for individual welding beads. To allow comparability of the laser energy input for the different parameter sets, the energy density E_n , according to equation 3.10, is used.

Table 5.1: Overview of the covered range of parameters in the experiments for individual welding beads, layers and the remelting (smoothing) of layers [169].

Experiment	Symbol	Welding beads	Layers	Layer remelting	Unit
Energy density	E_n	191-458	191-477	477-955	J mm ⁻²
Laser power	P	4-4.8	4-5	5	kW
Scanning velocity	v_s	400-800	400-800	200-400	mm min ⁻¹
Wire feed velocity	v_w	600-1200	500-2000	-	mm min ⁻¹
Deposition length	l	28	14-28	14	mm
Hatch distance	h_d	-	1.12-1.51	1.51	mm

To regenerate the armor of PFCs, it is necessary to be able to generate layers and not just individual welding beads. Therefore, after the successful deposition of individual welding beads, the next step involved studies on the generation of LMD-w layers. The layers are deposited by side by side arrangement of individual welding beads at specific distances (i.e., hatch distances) from one another. The parameter range listed in Table table 5.1 was used to generate these layers. The successful deposition of an entire layer is shown in fig. 5.1 C), with the area of 28×12 mm² representing the dimension of the plasma wetted surface of a typical divertor monoblock. In addition, experiments were carried out with layers stacked on top of each other, with the deposition direction rotating by 90° for each additional layer.

To reduce the surface waviness that is inherent for layers of welding beads that are deposited next to each other, experiments to smooth these layers were carried out. In this context, after deposition, the layers were remelted with the laser, with the intend of smoothing. The used process parameters are specified in table 5.1. These remelting attempts were carried out with the scan direction oriented perpendicular or parallel to the deposition direction of the layer. The surface waviness values sW_t of normal layers and various remelted layers was measured by laser profilometry and compared. More detailed analysis of the results and discussion of information obtained from cross-sectional micrographs and laser profilometry is given in the following results part of this section and the discussion section 5.3 of this chapter. For these investigations, several cross-sectional micrographs were prepared for SEM, EDX and EBSD analysis by cutting, grinding, polishing and partly also etching as detailed in section 4.4.

Results

The improvement in the quality of the welding beads, which is visible in the direct comparison in fig. 5.1 A) and B), was achieved by introducing several changes to the process. The balling defects which occurred in the preliminary studies (see fig. 5.1

A)) and are described in table 3.2 were successfully avoided by proper adjustment of the wire bending system and correct alignment of the wire, laser beam and substrate. For this purpose, the wire was positioned in the front feeding configuration at the leading edge, as illustrated schematically in fig. 3.7. Precise positioning of the wire is essential to achieve the desired surface finish, geometry control and defect free deposition of the welding beads [244]. The oxidation was prevented by substrate plate cooling and by selecting a favorable value for the inert gas flow (15 l min^{-1}). In the context of comprehensive parameter studies in the range described in table 5.1, welding beads of 28 mm in length depicted in fig. 5.1 B) were deposited. This process development and parameter optimization for individual welding beads has enabled the reliable deposition of tungsten with minimal oxidation, good bonding to the substrate and without balling defects. The surface of the deposited material appears as a shiny metallic film, indicating low or negligible contamination. Since the boiling point of tungsten oxides is significantly lower than the melting point of tungsten, which is exceeded during the manufacturing process, the presence of tungsten oxides in the deposited material is unlikely [106]. EDX measurements (as detailed in section 4.4) were performed for the LMD-w tungsten and for the substrate tungsten as the reference. These measurements revealed no difference in oxygen content between the deposited and reference tungsten. In both cases, the oxygen content was well below 1 %. The deposition of high-quality tungsten in the form of individual welding beads was thus demonstrated.

However, the generation of layers by the deposition of welding beads next to each other poses new challenges. Fig. 5.2 shows the etched cross-section of an LMD-w layer cut perpendicular to the deposition direction, as imaged using SE detection. The depicted layer was produced with an energy density of 210 J mm^{-2} . Fig. 5.2 highlights crack, pore, wavy surface, gas entrapment between the welding beads and failed connection to the substrate defects.

The high temperature gradients in the material during the process between the melt and the substrate result in residual stresses within the material during subsequent cooling and solidification. As the material cools down, it shrinks, which results in high tensile stresses. In areas subject to particularly high loads, these tensile stresses can discharge at weak points in the material in the form of the cracking observed here. Cracking during or directly after the solidification phase is called hot cracking [274]. These cracks originate from the surface of the deposited material and tend to run vertically down into the substrate material, reaching depths of up to several millimeters. Further information on how to deal with these cracks is provided in the discussion section 5.3.

The pores and gas entrapment defects, as visible in the micrograph shown in fig. 5.2, can have several causes. Pores can remain in the solidified material when gas is

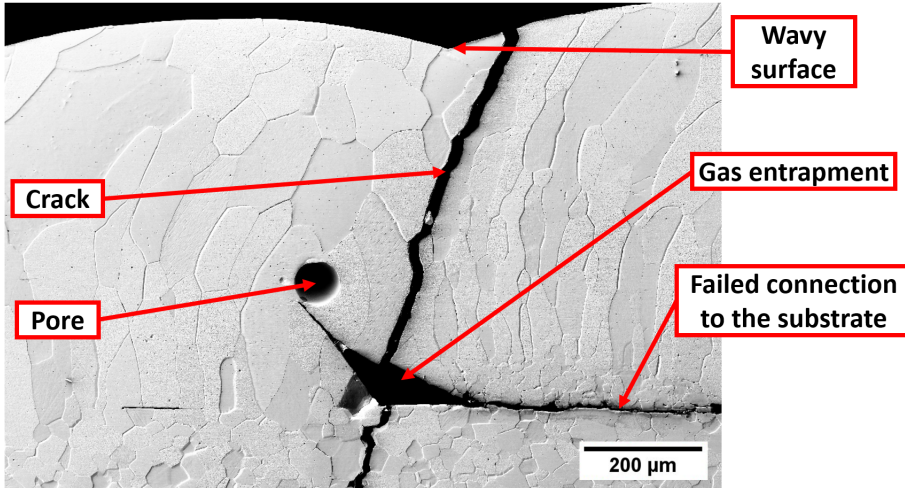


Figure 5.2: The figure displays the etched cross-sectional micrograph of an LMD-w layer cut perpendicular to the direction of deposition. Various visible defects are highlighted. This layer was generated with an energy density of 210 J mm^{-2} [169].

trapped during the melting process due to insufficient liquefaction or unfavorable flow transport in the material. The fluid dynamics in the melt pool are strongly influenced by the dynamic viscosity μ_d , as defined in equation 3.5 [186–189]. For tungsten, a surface tension of $\gamma = 2.48 - 0.31 \times 10^{-3} (T - T_m) \text{ N m}^{-1}$ ($T \geq 3695 \text{ K}$) can be assumed [275,276]. Increasing the temperature during the process reduces the viscosity and surface tension. This helps to minimize defects resulting from gas inclusions. In addition, the molten metal can wet the substrate more effectively, which helps to prevent the formation of failed connection to the substrate defects. However, the energy density should not be too high, as this can lead to keyhole defects, which are particularly likely to occur in keyhole mode (fig. 3.2). In the case of keyhole defects, gas transfers from the gas void (see fig. 3.3) into the melt pool due to unfavorable currents. If the material solidifies before the gas bubble can leave the melt pool, a spherical pore results, which is called a keyhole defect. Here, the Marangoni convection (eq. 3.4 and fig. 3.3) can prevent the gas bubble from leaving the liquid metal and thereby contribute to the formation of keyhole defects [277]. The energy density used and, consequently, the temperature in the melt pool, therefore should not be too low or too high in order to achieve satisfactory process results. Parameter studies were carried out with the aim of generating layers with as few defects as possible in the range specified in table 5.1. Area density analyses were carried out on the cross-section micrographs of the deposited layers, which are cut perpendicular to the deposition direction. The program ImageJ was used to determine the area density based on the cross-section micrographs. When comparing the

areal density achieved in the first and last development step for generating layers, an increase in the area density from 94.8 % to more than 99 % was achieved by raising the energy density E_n and hatch distance h_d between the welding beads. Increasing the hatch distance was particularly effective in preventing gas entrapment defects. The energy density of 477 J mm⁻² and hatch distance of 1.51 mm yielded the best results. Here, the maximum laser power of 5 kW was utilized and the lowest tolerable scan velocity selected. In addition, a wire feed velocity that could achieve the desired stable process was selected. The so called K value (equation 3.7) is the ratio between the wire feed velocity and the scanning velocity. For LMD-w, typically a K value ≥ 1 is selected, to ensure uniform, uninterrupted melting of the wire [222, 223, 226–229, 278, 279]. The most suitable parameters for the deposition of LMD-w tungsten layers on tungsten substrate based on this study and this specific facility setup are summarized in table 5.2.

Table 5.2: According to the results of the performed studies, the following process parameters are most suitable for the deposition of tungsten on a tungsten substrate when using this specific facility setup (see section 4.1) in combination with the described process settings [169].

Process parameter	Symbol	Welding beads	Unit
Energy density	E_n	477	J mm ⁻²
Projected area energy density	E_a	375	J mm ⁻²
Line energy input	E_t	750	J mm ⁻¹
Laser power	P	5	kW
Scanning velocity	v_s	400	mm min ⁻¹
Wire feed velocity	v_w	500	mm min ⁻¹
Hatch distance	h_d	1.51	mm

Various studies were carried out on the waviness of the surfaces of the LMD-w layers. The waviness of single layers and layers stacked on top of each other was measured. In addition, attempts were made to smooth the layers by remelting the deposited layer with the laser. The measured line profiles of different layers, remelted layers and stacked layers are plotted in fig. 5.3. Here, the y-axis was stretched in relation to the x-axis in order to more clearly display the waviness. The layers were stacked with an offset of 90° in the deposition direction. Among other aspects, this experiment showed that it is more advantageous to stack the layers without offset in the same deposition direction. Laser profilometry was used to obtain the surface profiles and to determine the waviness values sW_t of the LMD-w layer surfaces, which are listed in table 5.3. For the deposition of all layers listed in table 5.3 (with the exception of “1 layer smoothed perpendicular 955 J mm⁻²”), a laser power of 5 kW, scanning velocity of 400 mm min⁻¹, wire feed velocity of 500 mm min⁻¹ and hatch distance

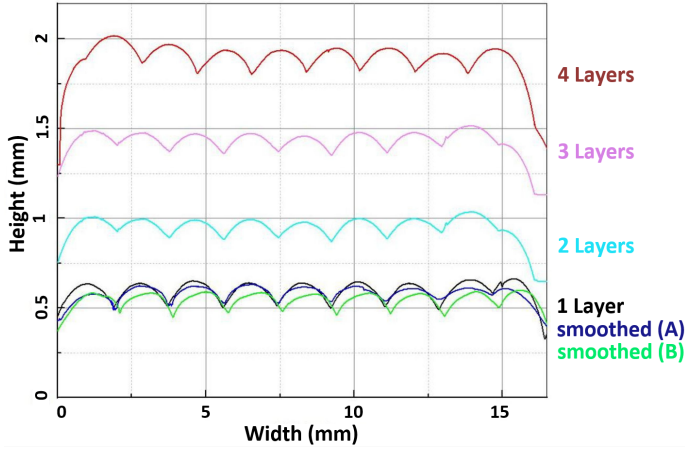


Figure 5.3: This figure illustrates the line profiles of various LMD-w layers measured by laser profilometry. The profile lines have not been shifted in the y-direction and represent the total coating thickness. The energy densities specified in table 5.3 were used to generate these layers. For the slightly smoothed layers (A) and (B), the energy densities of 477 and 955 J mm^{-2} were applied in the laser remelting process, which was performed perpendicular to the direction of deposition [169].

of 1.51 mm was used. Subsequently, an energy density of 477 J mm^{-2} was used for these layers. Only during the remelting process of the layer “1 layer smoothed perpendicular 955 J mm^{-2} ” the scanning velocity was decreased to 200 mm min^{-1} to increase the energy density. The energy density E_n according to 3.10 and the initial position on the z-axis before application of the last layer is listed in table 5.3. The surface waviness value sW_t was determined as specified in 4.5. For 2 stacked layers, the surface waviness is higher than for a single layer (see table 5.3). However, the waviness of 3 and 4 stacked layers is comparable to that of 2 stacked layers. The surface waviness sW_t seems to be saturated at a value of $\sim 240 \text{ }\mu\text{m}$. As shown in fig. 5.4 and table 5.3, the surface can be smoothed by remelting and thus reducing the waviness value. The direct comparison of a normal layer, a smoothed layer with perpendicular remelting and one with parallel remelting with the corresponding waviness value sW_t measured by laser profilometry is shown in fig. 5.4. The measured profile data are illustrated in fig. 5.4 in isometric view. Remelting with a higher energy density of 955 J mm^{-2} did indeed reduce the waviness further, but keyhole defects appeared in the cross-sectional micrograph, which are an indication of excessive energy density and temperature.

Table 5.3: The surface waviness values sW_t of various layers, stacked layers and smoothed layers measured by laser profilometry are listed here. For the deposition of the LMD-w coatings, the process parameters specified in table 5.2 were used.

	Symbol [Unit]	1 Layer	2 Layers	3 Layers	4 Layers
Initial position on the z-axis	d_z [mm]	0	+0.65	+1	+1.5
Surface waviness	sW_t [μm]	159	226	244	238
		1 Layer smoothed perpen- dicular	1 Layer smoothed perpen- dicular	1 Layer smoothed parallel	
Energy density (remelting)	E_n [J mm^{-2}]	477	955	477	
Initial position on the z-axis	d_z [mm]	0	0	0	
Surface waviness	sW_t [μm]	128	95	66	

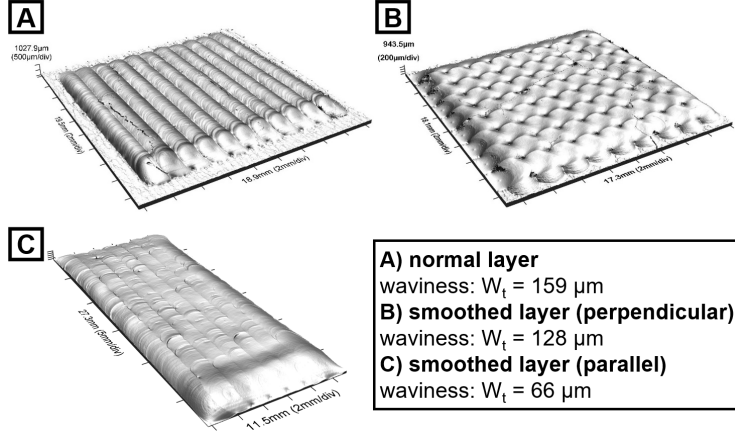


Figure 5.4: The surface topology of these layers, which were deposited by LMD-w, was measured using laser profilometry and is shown here in an isometric view. A) represents a normal LMD-w layer. B) and C) are LMD-w layers that were smoothed by laser remelting. In B) the laser remelting was performed perpendicular to the deposition direction and in C) parallel to the deposition direction. For all layers shown here, the energy density used in the deposition and subsequent smoothing processes was 477 J mm^{-2} [169].

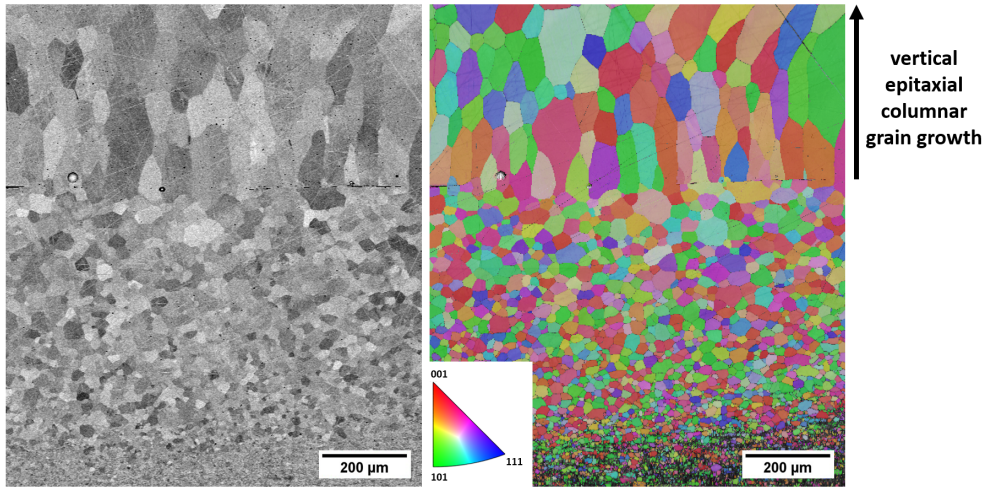


Figure 5.5: SEM (left) and EBSD (right) cross-sectional micrograph of an LMD-w layer and its substrate. The large grains in the upper area represent the deposited material, which is characterized by vertical epitaxial columnar crystal growth. The center area with the grains decreasing in size towards the bottom is the heat-affected zone of the LMD-w process. The original grain structure of the rolled substrate material is visible in the lower part of the images with the small grains. The LMD-w layer was deposited with an energy density of 210 J mm^{-2} [169].

The surface waviness can be reduced by remelting for both scan directions, but remelting parallel to the deposition direction of the LMD-w layer results in the lowest waviness value of $sW_t = 66 \text{ } \mu\text{m}$. Remelting reduces the maximum layer height of $650 \text{ } \mu\text{m}$ by about $50 \text{ } \mu\text{m}$, which is visible in fig. 5.3. The measured maximum height of 2, 3 and 4 stacked layers corresponds to about 1 mm, 1.5 mm and 2 mm, respectively. Overall, the surface waviness increases slightly for stacked layers before reaching saturation and can be significantly reduced by remelting [169].

Fig. 5.5 shows the SEM image on the left and the EBSD image on the right of an LMD-w layer deposited at 210 J mm^{-2} , cut perpendicular to the deposition direction. In the upper area, LMD-w deposited material with large columnar grains is visible. Partial epitaxial crystal growth vertically upwards from the substrate is visible in fig. 5.5. This is evident from the grains in the cross-sectional micrograph, which are located both in the substrate and in the deposited material and have the same grain orientation (see EBSD image in fig. 5.5). In this case, the tungsten grains grew during solidification according to the grain orientation of the substrate

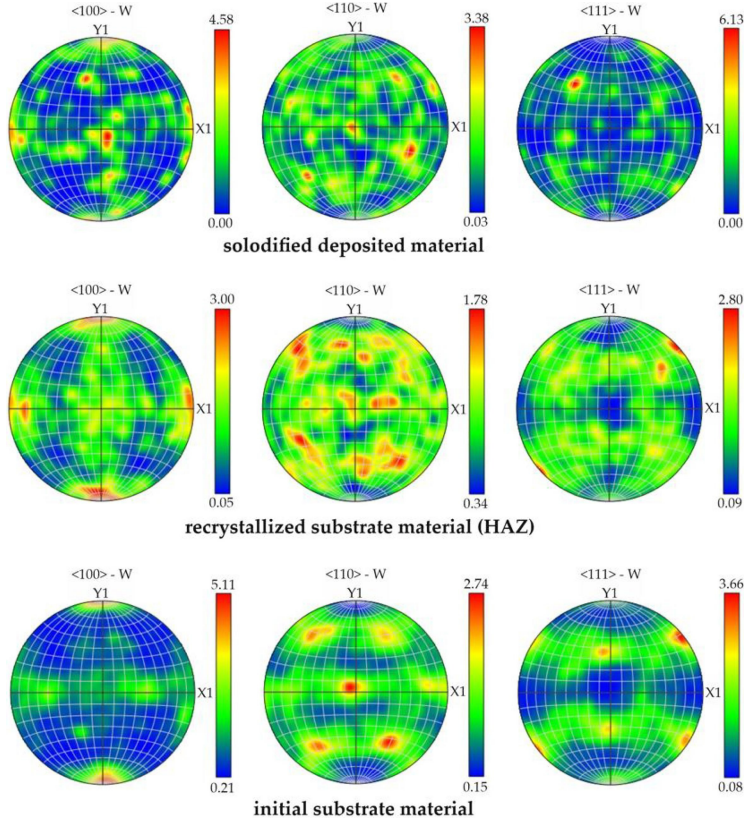


Figure 5.6: Depicted are the stereographic pole figures of $\langle 100 \rangle$, $\langle 110 \rangle$ and $\langle 111 \rangle$ for the solidified deposited material, the recrystallized substrate material in the HAZ of the substrate and the rolled initial substrate material determined from the EBSD data [169].

grains. This vertical grain growth from the substrate along the temperature gradient is also the reason for the elongation of the grains in the building direction, i.e. in the z-direction. The grains in this area had solidified from the melt and exhibit an average surface area of $3240 \mu\text{m}^2$ with an aspect ratio of 2.3. In the middle part of fig. 5.5, grains of medium size with an average surface area of $500 \mu\text{m}^2$ and aspect ratio close to 1 are visible. These grains belong to the heat affected zone (HAZ) and are partly solidified from the melt and partly recrystallized. Here, the HAZ is about $640 \mu\text{m}$ deep and the size of the grains within it decreases with increasing depth. At the bottom of fig. 5.5, the original grain structure of the tungsten substrate material resulting from forging is visible. These grains have the smallest surface area of $73 \mu\text{m}^2$ compared to the other areas, and are elongated in the xy-plane with an aspect ratio of 3.1. Fig. 5.6 illustrates the stereographic pole figures for the three just

described areas within fig. 5.5. The solidified deposited material and recrystallized substrate material grains in the HAZ show no preferred crystal orientation, i.e. no texture. However, the grains in the initial substrate material show a preference for the (110) orientation.

5.2 Influence of wavy topology on heat flux, erosion and redeposition

In order to investigate the influence of wavy topology of the plasma wetted surface of PFCs on the PWI, a set of ERO2.0 (see section 4.8) simulations were performed in the local scale. Here, a flat surface ($1.8 \times 1.8 \text{ mm}^2$) was selected as the reference geometry and two wavy topologies with waviness values of $W_t = 130 \text{ }\mu\text{m}$ and $260 \text{ }\mu\text{m}$ and the same dimensions in the x- and y-direction as the flat surface were used. The wavy surface geometries are represented in the code by a hill that is duplicated in x-direction to take into account the shadowing factor for wavy surfaces. The individual hills are represented by a half sine curve in the x-direction. The height of the curve from $z = 0 \text{ }\mu\text{m}$ to its apex corresponds to the target waviness value.

For the ERO2.0 simulations, data from a plasma background generated by SOLPS at the outer divertor target was used. The plasma data at a position 0.3 m distant from the strike line were used, with the intention to select data in an area of high tungsten gross erosion (see fig. 5.7). The net erosion value is lower than the gross erosion value since the net erosion takes into account the redeposition. When considering the erosion of wall material during reactor operation, it is important to note that a large proportion of the eroded tungsten particles are promptly ionized and redeposited within one Larmor radius. This significantly reduces the net erosion of material. The fact that the area of highest erosion is not exactly at the strike line, but has shifted along the vertical target is, due to the detachment. In this case, argon was assumed as the detachment gas.

The influence of wavy topologies on heat flux, gross erosion, redeposition and net erosion was investigated within the context of the ERO2.0 simulations described here. These aspects were calculated for the three geometries $W_t = 0 \text{ }\mu\text{m}$, $130 \text{ }\mu\text{m}$ and $260 \text{ }\mu\text{m}$ for scenarios where the plasma approaches from different incidence directions (ϕ) and incidence angles (α). Here, ϕ is the direction from which the plasma is incident on the geometry in relation to the deposition direction, during the deposition of the LMD-w layer. Scenarios were simulated with the plasma incident direction parallel to the deposition direction ($\phi = 0^\circ$), perpendicular to the deposition direction ($\phi = 90^\circ$) and diagonally ($\phi = 45^\circ$) for different incidence angles α . For the incidence angle in relation to the xy-plane, $\alpha = 0^\circ$ means parallel to the surface and 90° perpendicular to the surface. For α , the scenarios 1° , 5° and 10°

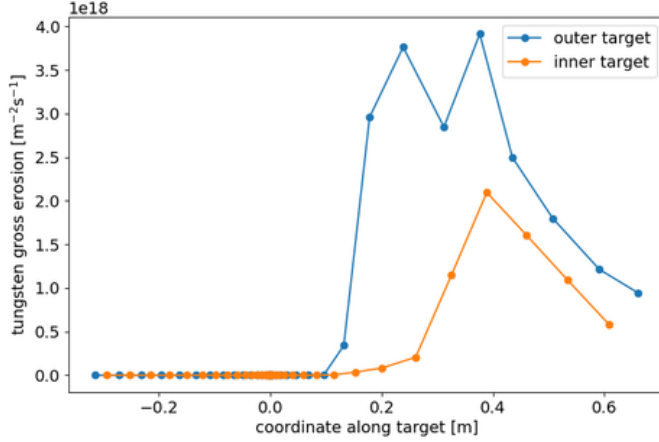


Figure 5.7: The simulated gross erosion along the inner and outer divertor targets of the European DEMO FPP for flat PFC surfaces. The coordinate 0 along the target represents the center of the strike line. The area of the highest gross erosion is at a distance of 0.2-0.4 m from the strike line, which is a effect of the detachment.

were simulated, for all three surface geometries and incidence directions. Fig. 5.8 illustrates the heat fluxes for the scenario $\alpha = 5^\circ$ and $\phi = 90^\circ$ and the three discussed geometries. Fig. 5.9 illustrates the heat fluxes on the $W_t = 130 \mu\text{m}$ geometry at a plasma incidence angle of $\alpha = 5^\circ$ for the two scenarios $\phi = 0^\circ$ and $\phi = 45^\circ$. Fig. 5.10 illustrates the net erosion (red) and redeposition (blue) for the scenario $\alpha = 5^\circ$ and $\phi = 90^\circ$ and the three discussed geometries. Fig. 5.11 illustrates the net erosion on the $W_t = 130 \mu\text{m}$ geometry at a plasma incidence angle of $\alpha = 5^\circ$ for the two scenarios $\phi = 0^\circ$ and $\phi = 45^\circ$. The resulting density of tungsten particles per volume in the plasma near the surface is visualized in fig. 5.12 for the $W_t = 130 \mu\text{m}$ and $W_t = 260 \mu\text{m}$ geometries and the scenario $\alpha = 5^\circ$ and $\phi = 90^\circ$. It is evident that the highest density of tungsten particles, $\sim 10^{16.5} \text{ m}^{-3}$, is found in the volume with the highest erosion around the leading edge (see fig. 5.10 and fig. 5.12). Table 5.4 lists the calculated peak heat fluxes, gross erosion and deposition rates for the three geometries, incidence directions and incidence angles. In addition, the total heat fluxes, gross erosion and deposition rates over areas of $1.8 \times 1.8 \text{ mm}^2$ for the three sample geometries and the respective plasma scenarios are listed in table 5.4. It is important to note that in this context, higher peak values do not necessarily mean higher total heat fluxes/erosion/deposition, but rather a stronger focus of these effects within certain areas. The values calculated in the context of this study should not be interpreted as quantitatively reliable values, due to various assumptions that had to be made in the context of these simulations, but rather as a qualitative comparison of the various scenarios.

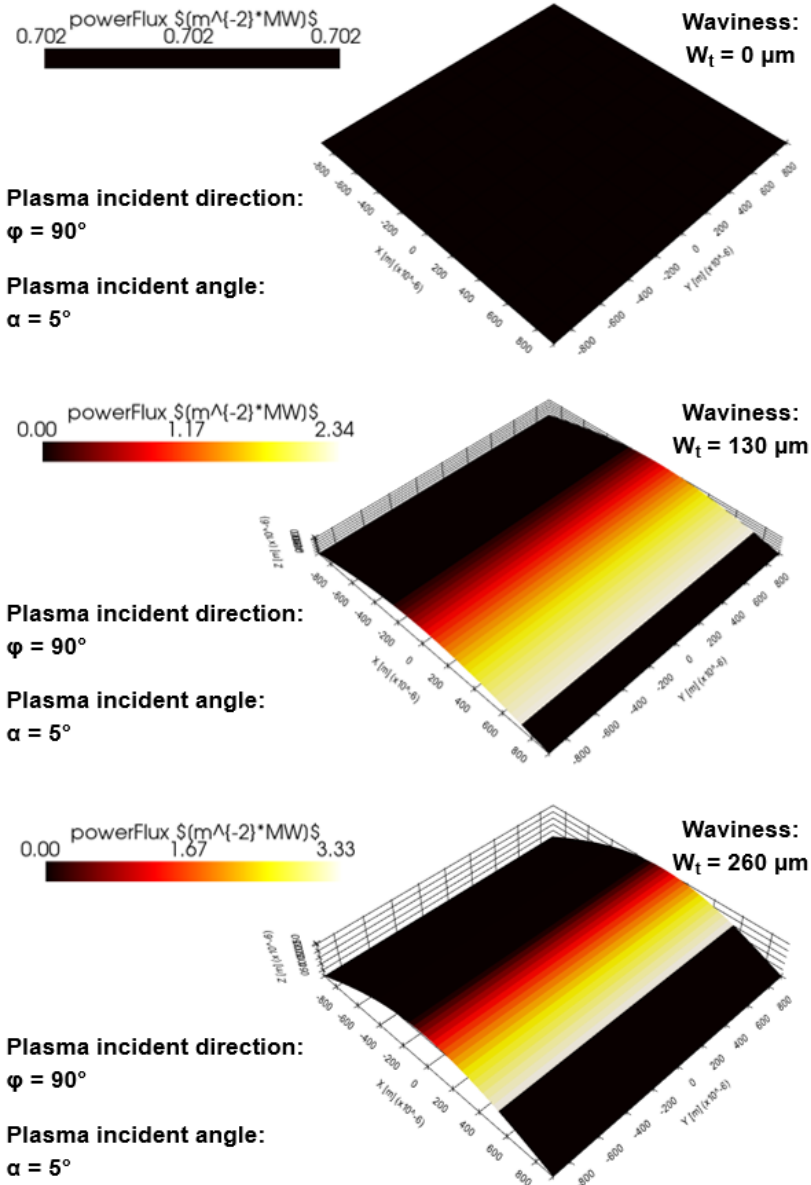


Figure 5.8: The calculated heat fluxes in the scenario where the plasma enters from the incidence direction $\phi = 90^\circ$, i.e. perpendicular to the deposition direction, and hits the three geometries shown with an angle of incidence to the xy-plane of $\alpha = 5^\circ$. The shadowing factor is achieved by duplication of the geometries in the x- and y-direction.

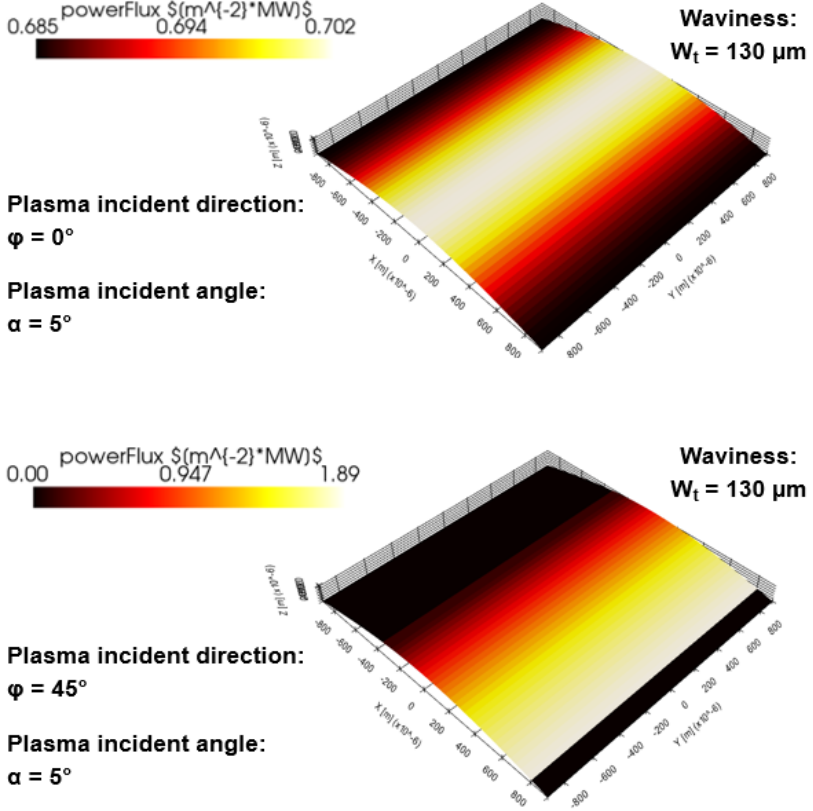


Figure 5.9: The calculated heat fluxes for scenarios with a plasma incidence direction parallel to the deposition direction of the LMD-w layer ($\phi = 0^\circ$) and diagonal ($\phi = 45^\circ$) at a plasma incidence angle of 5° for the $W_t = 130 \mu m$ geometry.

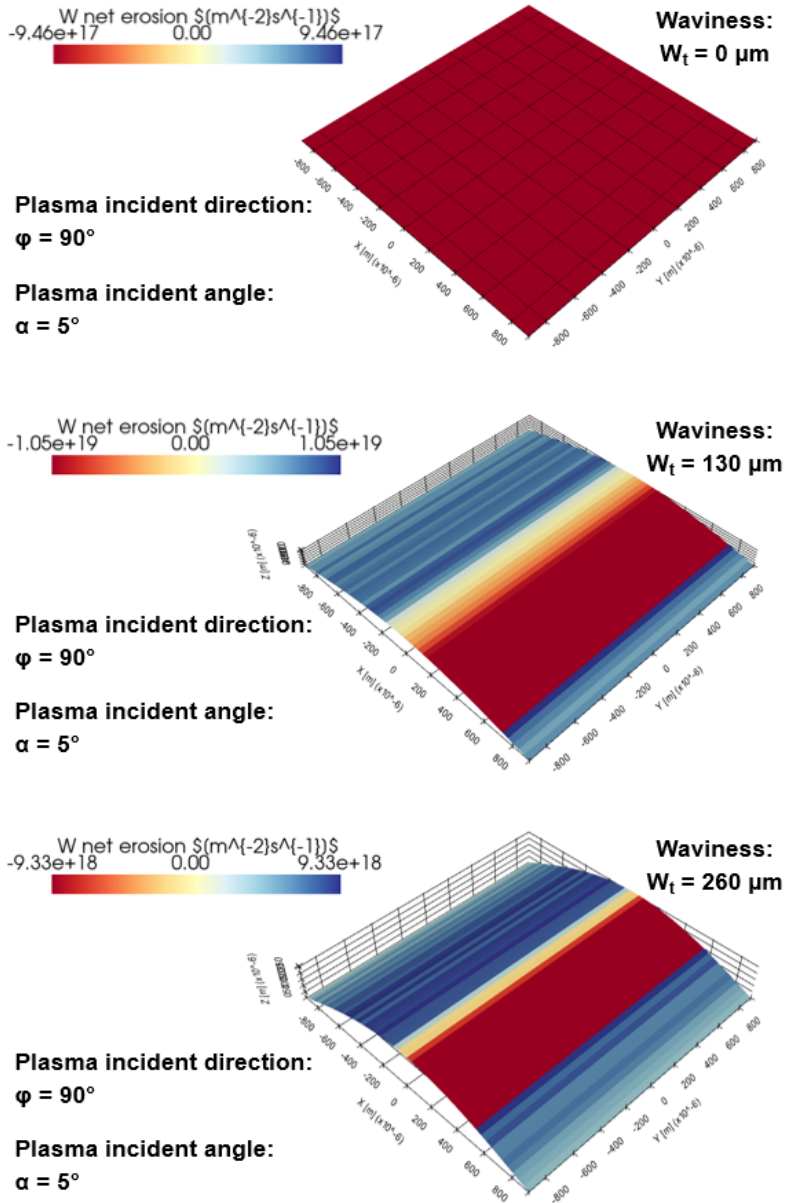


Figure 5.10: The calculated tungsten net erosion in the scenario where the plasma enters from the incidence direction $\phi = 90^\circ$, i.e. perpendicular to the deposition direction, and hits the three geometries shown with an angle of incidence to the xy-plane of $\alpha = 5^\circ$. In the red areas, material is eroded and in the blue areas, redeposited. The shadowing factor is achieved by duplication of the geometries in the x- and y-direction.

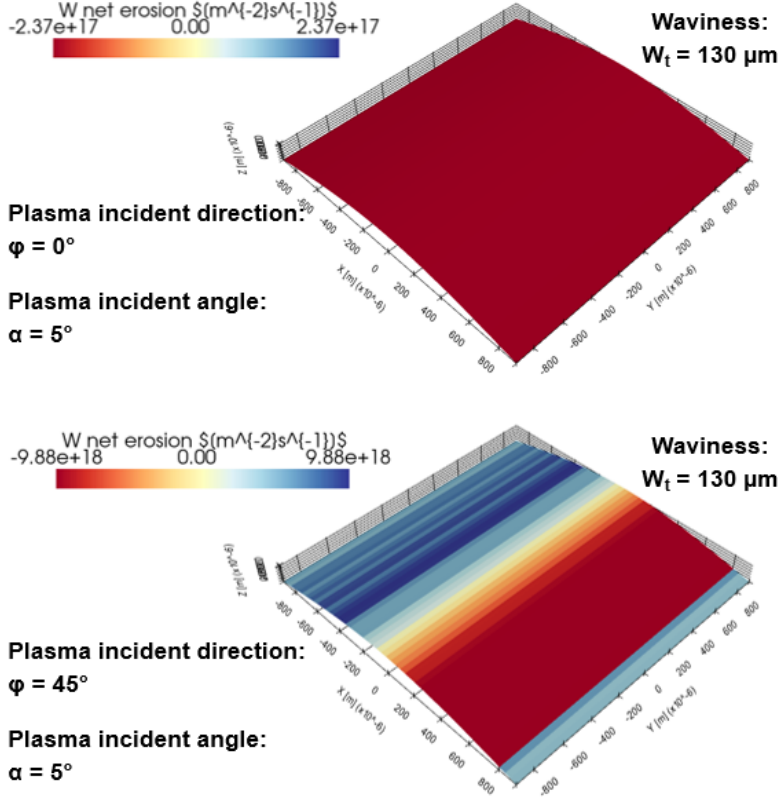


Figure 5.11: The calculated tungsten net erosion for scenarios with a plasma incidence direction parallel to the deposition direction of the LMD-w layer ($\phi = 0^\circ$) and diagonal ($\phi = 45^\circ$) at a plasma incidence angle of 5° for the $W_t = 130 \mu\text{m}$ geometry. In the red areas, material is eroded and in the blue areas, redeposited.

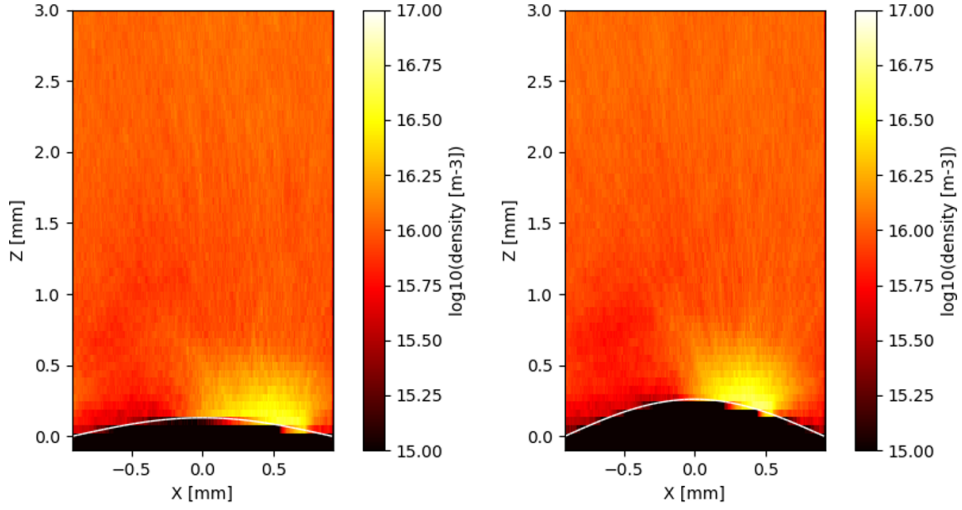


Figure 5.12: The calculated tungsten density in the plasma volume is encoded in color. In the area around the leading edges, the tungsten density is highest. From this view, the plasma impinges on the test geometries from the right at an incidence angle of $\alpha = 5^\circ$ and an incidence direction of $\phi = 90^\circ$. The left test geometry features a waviness of $W_t = 130 \mu\text{m}$ and the right one a waviness of $W_t = 260 \mu\text{m}$.

Table 5.4: The calculated peak heat flux, gross erosion and deposition values are listed in this table for all considered cases, with the three geometries ($W_t = 0 \text{ } \mu\text{m}$; $W_t = 130 \text{ } \mu\text{m}$; $W_t = 260 \text{ } \mu\text{m}$), the incidence direction ϕ (0° ; 45° ; 90°) and incidence angle α (1° ; 5° ; 10°). The total heat flux, gross erosion, and deposition values for the $1.8 \times 1.8 \text{ mm}^2$ test surfaces are also listed.

		$W_t \text{ } [\mu\text{m}]$			$W_t \text{ } [\mu\text{m}]$		
		0	130	260	0	130	260
$\phi \text{ } [^\circ]$	$\alpha \text{ } [^\circ]$	Peak heat flux $[\text{MW m}^{-2}]$			Total heat flux $[\text{W}]$		
0	1	0.14	0.14	0.14	0.47	0.48	0.49
0	5	0.70	0.70	0.70	2.35	2.35	2.35
0	10	1.40	1.40	1.40	4.69	4.69	4.69
45	1	0.14	0.97	1.40	0.47	0.50	0.52
45	5	0.70	1.89	2.69	2.35	2.35	2.30
45	10	1.40	2.59	3.52	4.69	4.69	4.66
90	1	0.14	1.14	1.54	0.47	0.47	0.44
90	5	0.70	2.34	3.33	2.35	2.41	2.35
90	10	1.40	3.09	4.36	4.69	4.66	4.78
$\phi \text{ } [^\circ]$	$\alpha \text{ } [^\circ]$	Peak gross erosion $[\text{m}^{-2} \text{ s}^{-1}]$			Total gross erosion $[\text{s}^{-1}]$		
0	1	2.6e+18	2.6e+18	2.6e+18	8.6e+12	8.7e+12	9.0e+12
0	5	1.3e+19	1.3e+19	1.3e+19	4.2e+13	4.2e+13	4.2e+13
0	10	2.6e+19	2.5e+19	2.5e+19	8.3e+13	8.3e+13	8.3e+13
45	1	2.6e+18	1.7e+19	2.4e+19	8.6e+12	9.1e+12	9.4e+12
45	5	1.3e+19	3.3e+19	4.6e+19	4.2e+13	4.2e+13	4.1e+13
45	10	2.5e+19	4.5e+19	6.0e+19	8.3e+13	8.3e+13	8.3e+13
90	1	2.6e+18	2.0e+19	2.6e+19	8.6e+12	8.6e+12	8.0e+12
90	5	1.3e+19	4.0e+19	5.7e+19	4.2e+13	4.3e+13	4.2e+13
90	10	2.5e+19	5.4e+19	7.5e+19	8.3e+13	8.3e+13	8.5e+13
$\phi \text{ } [^\circ]$	$\alpha \text{ } [^\circ]$	Peak deposition $[\text{m}^{-2} \text{ s}^{-1}]$			Total deposition $[\text{s}^{-1}]$		
0	1	2.4e+18	2.7e+18	2.8e+18	7.0e+12	7.1e+12	7.3e+12
0	5	1.1e+19	1.2e+19	1.3e+19	3.3e+13	3.3e+13	3.4e+13
0	10	2.3e+19	2.3e+19	2.3e+19	6.3e+13	6.3e+13	6.4e+13
45	1	2.5e+18	3.7e+18	4.1e+18	7.0e+12	7.3e+12	7.5e+12
45	5	1.1e+19	1.3e+18	1.4e+19	3.3e+13	3.3e+13	3.2e+13
45	10	2.1e+19	2.2e+19	2.4e+19	6.3e+13	6.2e+13	6.1e+13
90	1	2.4e+18	3.4e+18	3.7e+18	6.9e+12	6.9e+12	6.4e+12
90	5	1.2e+19	1.4e+19	1.5e+19	3.3e+13	3.3e+13	3.2e+13
90	10	2.3e+19	2.2e+19	2.3e+19	6.2e+13	6.1e+13	6.1e+13

5.3 Discussion

Process development for tungsten

The LMD-w deposition experiments described above clearly highlighted the fact that the positioning of the wire, the laser and the build plane in relation to each other has a significant influence on the process stability and thus on the process performance. In this context, the positioning of the tungsten wire tip at the leading edge of the melt pool in the front feeding configuration (see fig. 3.7 A)) has proven suitable for the used LMD-w facility. If the positioning is not aligned correctly, even with the other parameters set to their respective optimal values, a good process result cannot be achieved. It was shown that, through precise positioning, the occurrence of balling defects, can be largely prevented. In addition, the wire bending system should be properly adjusted to compensate for the deformation of the spooled wire and to prevent lateral oscillations (see fig. 3.8) that can compromise process stability.

The increase of thermal energy in the melt pool was able to largely eliminate the occurrence of failed connection to the substrate defects. This is due to the resulting reduction in the dynamic viscosity and surface tension that dominate the melt pool, enabling better wetting of the substrate. In addition, this favors the constant transition of molten wire material into the melt pool. However, the energy density used was kept low enough to avoid the increased occurrence of keyhole defects, which would be a major problem if the applied energy density was excessively high. The improved flowing properties of the tungsten achieved by higher temperatures also helped to reduce the gas entrapment and the resulting pore defects, which occurred when layers are formed by placing the welding beads next to each other. However, gas entrapment defects between the welding beads primarily could be prevented by a slight increase in the hatch distance between the beads.

AM of metallic materials under high deposition rates is usually characterized by high process temperatures and comparatively slow cooling. This usually results in large columnar grains with elongation in the build direction (in this case z-direction) [205, 249]. This columnar grain structure with elongation of the grains in the build direction is also observed in the material deposited by LMD-w. This behavior can be attributed to the partial epitaxial crystal growth from the substrate material in the z-direction. Since cracks in materials with grains elongated perpendicular to the surface also tend to run in this direction, this type of grain structure is advantageous to prevent the breaking out of grains that would pass into the plasma and thus cool it [113]. In most AM processes with clearly oriented temperature gradients, the nucleation of crystals that grow into grains starts in the underlying solid material, if the different layers have the same chemical composition [22, 205, 249, 280, 281]. The fact that the solidified grains are relatively large indicates a comparatively slow cooling process. And the partially epitaxial grain growth underlines the good bond

with the substrate.

Furthermore, in the LMD-w process, the substrate is also melted to a certain depth. This effect might help to melt and thus heal damages such as cracks in previously used PFCs. In chapter 7, this topic is investigated in detail and it is shown that this form of repair works for damages typical for fusion reactors.

To reduce the probability of cracking, in future studies the substrate material should be heated to at least 600 °C, which raises the tungsten above its ductile to brittle transition temperature. This also reduces temperature gradients in the material between the molten tungsten and the comparatively cold substrate, which lead to undesired residual stresses during the solidification of the metal. And these residual stresses can discharge in the form of macrocracks. To completely avoid cracks, substrate heating of 1000 °C will probably be necessary [193, 201, 205, 282]. With substrate heating, the entire tungsten substrate plate has to be protected from oxidation by either inert gas or vacuum [106]. In the eventual application inside the fusion reactor, the PFCs will be preheated inherently. Since, following a certain period of operation, the PFCs of high-power fusion reactors will be heated by the nuclear decay energy. Precisely tailored pre- and post-heating could also be realized, using an additional laser beam with a larger spot diameter. A configuration with a central laser spot for melting and one or more laser spot rings surrounding it could also be used to achieve more homogeneous temperature profiles and thus flatter temperature gradients in the material. Furthermore, it is possible that only the use of a single wider laser spot with flatter flanks of its energy density profile might be sufficient to prevent the formation of cracks, which has already been proven successful in the deposition of tungsten using the EBAM process [283]. The coaxial LMD-w process, where the wire is feed vertically from above and where the wire can be preheated by the laser itself, could be beneficial for tungsten deposition and should also be investigated in the future. According to the literature hot cracking (cracking during or directly after the solidification phase) tends to occur when the K value (equation 3.7) is below the material specific threshold value [224]. Since increasing the K value at constant v_s is accompanied by an increased mass flow rate $\dot{m}$ (equation 3.6), and the 5 kW laser just manages to melt tungsten sufficiently at the process parameters currently used, an upgrade of the available laser power would be necessary to perform studies with large K values.

The waviness of the welding beads can be greatly reduced by remelting the surface. In this process, remelting with an energy density of 477 J mm⁻² with a scanning direction parallel to the deposition direction of the previously applied coating proved to be most advantageous. Whether the reduction in surface waviness is necessary at all, taking into account the heat fluxes, erosion and deposition occurring on PFCs, was investigated in section 5.2 and is discussed below.

Influence of wavy topology on heat flux, erosion and redeposition

In the ERO2.0 studies conducted here, the peak and total heat flux, erosion, and deposition values increased with increasing plasma incidence angle α for all surface topologies and plasma scenarios. This was expected and this is the reason why divertors are designed in a way that the plasma strikes the PFC's surface at a flat angle of incidence α . Since this distributes the loads over a large surface area.

Furthermore, it turned out that both surface waviness and plasma incidence direction ϕ have virtually no influence on the calculated total heat flux, erosion, and deposition values (see table 5.4). However, it also turned out that wavy topologies of the surfaces of PFCs influence the distribution of heat flux, erosion, and deposition and thus their peak values during plasma exposure. In contrast to the smooth surface, where the peak values for heat fluxes and erosion remained constant for the various plasma incidence directions ϕ , these values increased for a rising surface waviness W_t with a rising plasma incidence direction angle ϕ (for $0^\circ < \phi \leq 90^\circ$). The peak was reached at a plasma incidence direction perpendicular to the deposition direction ($\phi = 90^\circ$; see fig. 5.8 and fig. 5.10). Thus, the layers should be generated with the deposition direction parallel to the subsequent plasma incidence direction ($\phi = 0^\circ$), to distribute the loads during fusion reactor operation as evenly as possible.

It should be noted that the peak heat flux increase on wavy surfaces occurred due to local concentration and that the total heat flux did not increase. The peak erosion values also increased slightly with increasing waviness and concentrated locally, even though the total erosion remained constant. The deposition was distributed relatively evenly and the peak deposition values did not increase significantly with increasing waviness. The deposition occurred over the entire area of the surfaces, with the highest deposition mainly in the regions of high erosion. This can be attributed to prompt redeposition of eroded material within one Larmor radius.

The results of these simulations are intended merely to provide a qualitative comparison of the scenarios under consideration, and the resulting quantitative data should not be seen as exact values for the European DEMO FPP due to the various assumptions that had to be made.

5.4 Conclusion and outlook

It has been successfully demonstrated that deposition of tungsten on tungsten substrate is possible using the AM process of LMD-w. In the first step, the process was adapted to enable the deposition of individual tungsten welding beads. Since the initial positioning has significant influence on the process stability of LMD-w, the first priority was to find suitable alignment for the wire, laser beam and building plane in relation to each other. It was found that good results can be achieved by

positioning the wire tip at the leading edge of the weld pool in the front feeding configuration (see fig. 3.7). Consequently, this configuration was used for the further experiments. In addition, the correct setting of the wire bending device was essential to compensate for deformations in the coiled wire that could lead to lateral oscillations, which also affect process stability (see fig. 3.8). After comprehensive parameter studies in which the laser power, scanning velocity and wire feed velocity were varied, individual welding beads have been deposited with consistently high quality (see fig. 5.1 B)).

The welding beads are arranged next to each other in order to coat larger areas as evenly as possible. However, defects appeared in the cross-section of these layers. By optimizing the parameters, the occurrence of gas entrapment and failed connection to the substrate defects were eliminated. The probability of pore defect occurrence was decreased significantly. This was achieved by increasing the applied energy density. In the selection of suitable parameters, it was necessary to ensure that the thermal energy is sufficiently high to achieve good flowability and wetting behavior. However, the energy density is not allowed to be too high, otherwise the probability of keyhole defect occurrence increases, which has to be avoided. Based on this study, the most suitable process parameters for LMD-w deposition of tungsten layers at the described facility are listed in table 5.2. It was demonstrated that stacking the layers one on top of each other is a suitable option to increase the coating thickness. Alternatively, the K value could be adjusted or wire with different diameters utilized to achieve customized layer heights.

The macro cracks that occur during the generation of LMD-w tungsten layers, which usually start at the surface of the deposited layer and run perpendicular down through the layer into the substrate, are unavoidable with the current setup. The occurrence of these cracks due to the high temperature gradients during the solidification and cool down of tungsten is a known issue. Studies on the AM processing of tungsten (PBF-LB/M and PBF-EB/M) have shown that these cracks can be reduced with substrate temperatures of at least 600 °C and be completely avoided by increasing the substrate temperature to 1000 °C [193, 201, 205, 282]. The PFCs of high-power fusion reactors are radioactively activated after certain periods of operation with neutron exposure. Accordingly, these PFCs release significant amounts of nuclear decay energy. This means that the regeneration process of PFCs armor by LMD-w within a high-power fusion reactor would already feature inherent substrate heating. However, as it is currently unclear to which temperatures the PFCs will be moderated by the cooling system during maintenance phases, alternative options to achieve precise control of the substrate temperature would be advantageous. This could be achieved by adding a second laser spot, which would provide pre- and post-heating. This could be achieved, with the coaxial process, by a second ring-shaped laser spot (for pre- and post-heating) that surrounds the pri-

mary laser spot (for melting). An increase in the laser spot diameter alone can also help to reduce or eliminate the occurrence of macrocracks due to homogenization of the temperature profile [283]. The undesired steep temperature gradients in the material, caused by the beams energy density profile, can be reduced by customized beam shaping. Ring-shaped laser spots and beam energy density profiles with less steep flanks can help to distribute the energy more evenly and thus help to prevent cracks. In future endeavors, trials with substrate heating and the aforementioned options to homogenize the temperature profile in the material, should be performed, to investigate measures for crack prevention during LMD-w tungsten processing.

With regard to the waviness of the layers resulting from the side-by-side arrangement of the welding beads, experiments were conducted with the intention of reducing the waviness. For smoothing the surface, the layers were once again treated with laser irradiation. The best results were achieved with the scan direction parallel to the deposition direction of the layer. In this way, the measured layer surface waviness sW_t was halved. However, the question of whether this reduction in surface waviness is necessary at all was examined using the ERO2.0 simulations outlined in section 5.2 and discussed in section 5.3.

To gain knowledge about the influence of wavy topology of the plasma-facing surface of PFCs on heat fluxes, erosion and deposition, ERO2.0 simulations were performed. For this purpose, three different geometries were exposed to plasma. One with a flat surface, and the others with surface waviness values of $W_t = 130 \mu\text{m}$ and $W_t = 260 \mu\text{m}$ were exposed to plasma. Data generated by SOLPS at the outer vertical target of the future European DEMO FPP was used as the plasma background. Various scenarios with different plasma incidence directions (ϕ) and plasma incidence angles (α) were run. The peak and total heat flux, erosion, and deposition values increased with increasing plasma incidence angle α for all surface topologies and plasma scenarios. This was expected and this is the reason why divertors are designed in a way that the plasma meets the plasma-facing surface at a flat angle. Since this distributes the loads over a larger area. Furthermore, it was found that the total heat flux, erosion, and deposition are not influenced by surface waviness or plasma incidence direction ϕ . The peak heat flux and erosion values hardly differ from those of a smooth surface when the LMD-w deposition direction is chosen parallel to the plasma incidence direction ($\phi = 0^\circ$). But heat fluxes and erosion at wavy topologies for $\phi > 0^\circ$ are concentrated in certain areas (as visible in fig. 5.8 and fig. 5.10), which leads to increased local peak heat flux and erosion values. It was observed that with increasing ϕ and increasing surface waviness (for $0^\circ < \phi \leq 90^\circ$), the peak heat flux and erosion values increase. Overall, the deposition is relatively evenly distributed across the surface, but slightly increased in the regions of high erosion. The tungsten deposition occurs primarily in the form of prompt redeposition of eroded tungsten within one Larmor radius.

The results of this study should rather be used as a qualitative comparison of the individual scenarios than as quantitative data due to various assumptions that had to be made. However, the data obtained is conclusive enough for a recommendation for the deposition of layers parallel to the plasma incidence direction ($\phi = 0^\circ$). In this case, surface waviness does not result in local concentration of heat flux and erosion.

Overall, the conducted studies demonstrated that it is potentially possible to compensate for erosion losses by deposition of new LMD-w generated coatings on the tungsten armor of PFCs. The next chapter will examine how these LMD-w coatings behave at exposure to fusion relevant steady state, as well as combined steady state and transient thermal loads.

6 High heat flux testing of coated plasma-facing components

6.1 Test component development

In this chapter, high heat flux (HHF) experiments to test the behavior of LMD-w coated PFCs at fusion relevant thermal loads, that are expected in the divertor area of future fusion reactors, are presented, analyzed and discussed. LMD-w coated test components were thermally loaded in the electron beam facility JUDITH 2 (see section 4.2).

Since it is difficult to get a whole divertor monoblock chain (see fig. 2.14) for these early-stage experiments, an alternative test component design was developed. The test components were designed to be able to create conditions for the coatings during the HHF experiments that are comparable to those in the divertor area of future FPPs. To ensure that the test components can be manufactured in-house, the flat tile design was chosen. In contrast to monoblock chains, these test components enable testing of a large quantity of coatings under fusion-relevant thermal conditions more easily, in less time and at lower cost. This test component design is displayed in fig. 6.1 B) in the cross section and isometric view next to a monoblock chain cross section in fig. 6.1 A). The components were designed in such a way that the plasma-facing surface area is $28 \times 12 \text{ mm}^2$, just like that of a monoblock [284] (fig. 6.1). In addition, the test components were specifically designed so that the thermal energy impinging at the center of the tungsten must flow through 8 mm of tungsten and then 2.5 mm copper to reach the coolant. This is comparable to the heat flow within real monoblocks (see the cross sections in fig. 6.1 A.1) and B.1)). Thereby, conditions comparable to the real monoblock design can be created for the LMD-w layers which are located on the plasma-facing surface of the components. Here it should be mentioned that the tungsten double tiles, which should be coated by LMD-w, were manufactured with initial heights of 7 mm instead of 8 mm to simulate the erosion of 1 mm. Subsequently, tungsten was applied to these 7 mm high double tiles by means of LMD-w to achieve about 8 mm height.

On monoblock chains, the individual monoblocks are spaced 0.5 mm from each

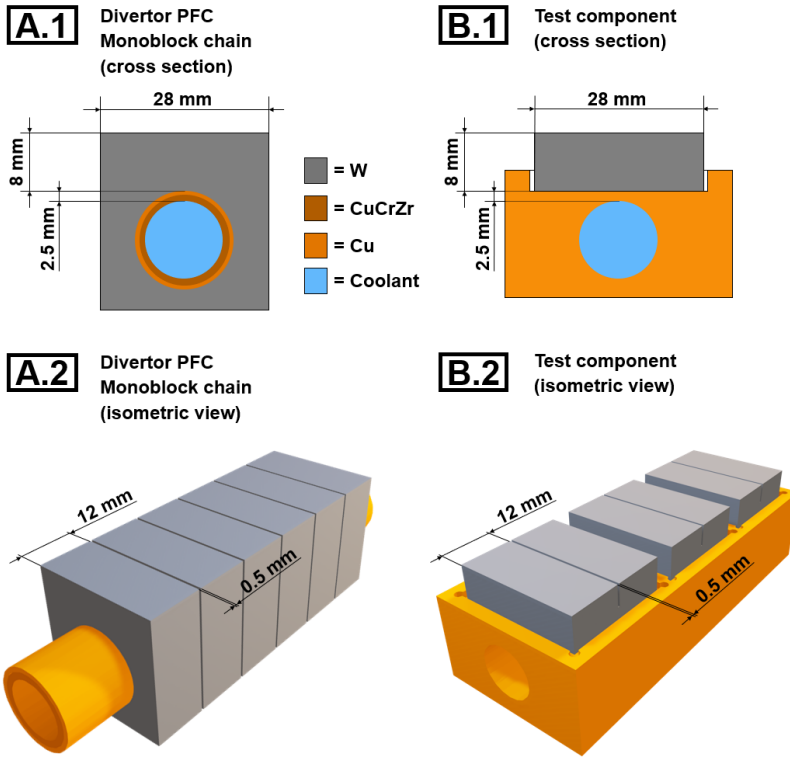


Figure 6.1: A.1) depicts the cross-section and A.2) the isometric view of a CAD model of a divertor monoblock. B.1) depicts the cross-section and B.2) the isometric view of the test geometry used for the HHF experiments. In A) and B), the thermal energy applied to the component surface at the top of this figure must flow through 8 mm W and then 2.5 mm Cu/CuCrZr. The plasma-facing surface of the tungsten components is, in both cases, $28 \times 12 \text{ mm}^2$ [5,171,284].

other. To test whether the LMD-w tungsten deposition is precise enough to maintain this distance, the tungsten tiles were manufactured in pairs with a spacing slot of 0.5 mm (see fig. 6.1). These pairs remain connected at the bottom by 3 mm thick tungsten. This ensured that the tiles are precisely 0.5 mm apart.

The calculated surface temperatures of divertor monoblock PFCs during exposure to steady state heat fluxes are specified in table 6.1. Since the cooling parameters and designs considered varied slightly depending on the respective study, an expected surface temperature range is specified [5,285–287]. The aim of the subsequent HHF experiments is to expose the top surface of the LMD-w coated tungsten double tiles to temperatures equivalent to those reached at the surface of divertor components during fusion reactor operation. To perform tests with sufficiently high heat

Table 6.1: The calculated surface temperature ranges resulting from the exposure to the respective steady state heat fluxes on the surface of the tungsten armor of the monoblocks are listed below [5, 285–287].

Heat flux	Temperature ranges
10 MW m ⁻²	~900-1100 °C
15 MW m ⁻²	~1200-1600 °C
20 MW m ⁻²	~1700-2200 °C

fluxes, the component must be cooled intensely to dissipate large amounts of energy that impinges on its surface. In this way, the temperatures specified in table 6.1 are reached at the surface of the test components. In addition, the active cooling achieves realistic temperature gradients inside the test components. For the HHF studies, the test components were connected to the cooling circuit of the JUDITH 2 electron beam facility via clamping elements. Table 6.2 lists the parameters used during the thermal loading studies to dissipate heat from the test components with purified water. The maximum heat flux, which can be dissipated by convection at

Table 6.2: This table summarizes the most relevant adjustable parameters of the JUDITH 2 sample cooling system. In this case, the coolant is purified water [171].

Cooling parameters	Symbol	Value	Unit
Fluid temperature	T_f	70	°C
Fluid velocity at the input	v_f	7.5	m s ⁻¹
Fluid pressure at the input	p_f	4	MPa

the inner surface of the cooling channel, was calculated using the ITER Nukiyama code. Here, the Nusselt number, which is the ratio of total heat transfer to conductive heat transfer at a boundary in a fluid, is calculated according to the Sieder-Tate equation (see equation 4.18). With the parameters listed in table 6.2 for cooling the component illustrated in fig. 6.1, 300 °C was calculated as the critical temperature at the inner surface of the cooling channel. If this temperature is exceeded, excessive boiling of the coolant causes the formation of gas films on the inside of the cooling channel. Thermal stationary FEM calculations using ANSYS software revealed that loading the component at the surface of the tungsten tiles with 14 MW m⁻² results in a peak temperature of 291 °C on the inner surface of the cooling channel. Hence, for the HHF experiments, 14 MW m⁻² was selected as maximum steady state heat flux to preserve a sufficient safety margin. The resulting surface temperatures are visualized in fig. 6.2 A) and the temperature profile within the component in the cross section view in fig. 6.2 B). The resulting surface temperatures on the inner surface

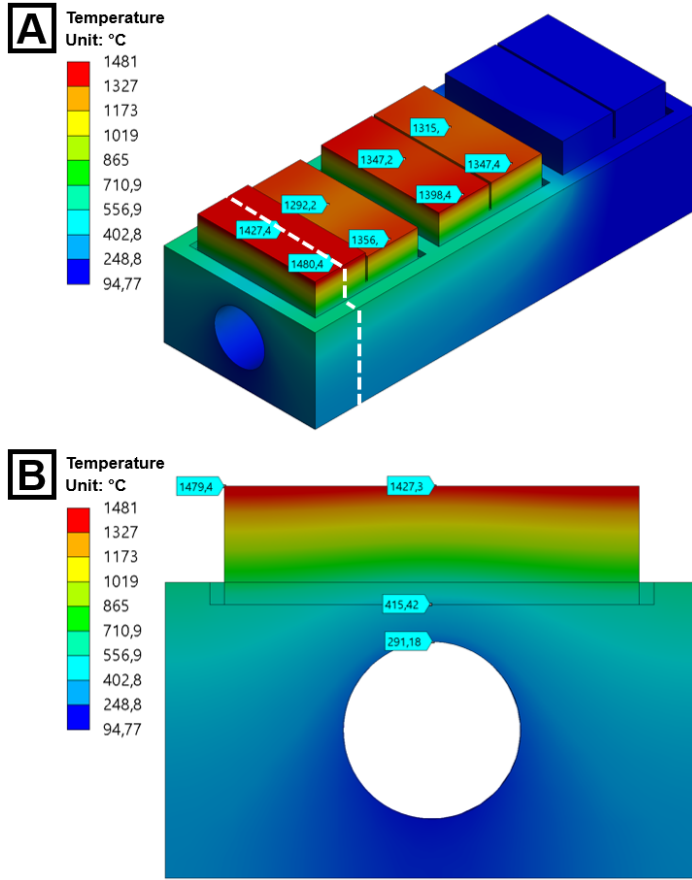


Figure 6.2: These FEM simulation images visualize the temperature distribution in the component during 14 MW m^{-2} loading. Here, only the two left tungsten double tiles are loaded. A) shows the isometric view where the two left tungsten double tiles with their surfaces illustrated in red (with the heights from left to right: 8 mm and 7 mm; 7.5 mm and 7.5 mm) exhibit different temperatures. In the center of the tiles, the temperature is higher than at the edge. This is due to the shorter distance between the surface and the coolant in the center of the tile. B) shows the cross section, cut according to the white dashed line in A), of the 8 mm high tungsten tile [171].

of the cooling channel are visualized in the side cross-section view in fig. 6.3 A). The energy incoming on the top surface of the tungsten double tiles takes the shortest path through the components and focuses at the top cooling channel to heat fluxes of 22 MW m^{-2} . In fig. 6.3 B) the heat fluxes on the upper inner surface of the cooling channel are visualized in the cross-section view from below. In comparison with

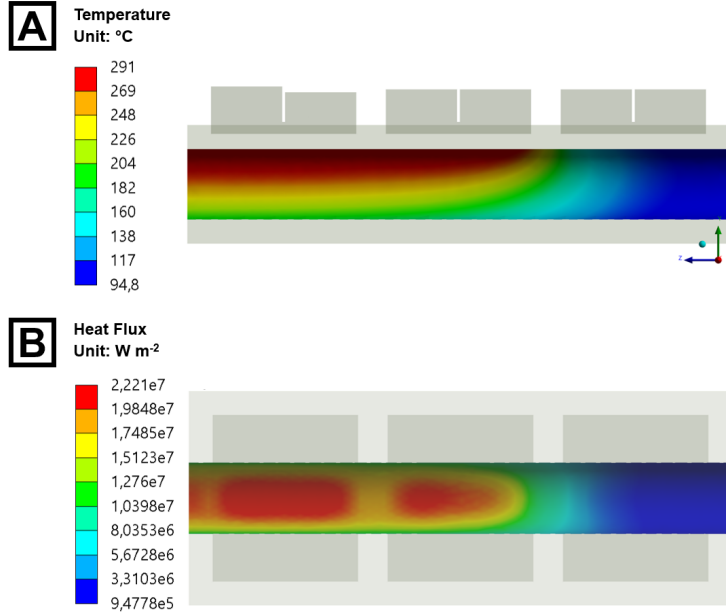


Figure 6.3: Illustrated are images generated by FEM simulation during 14 MW m^{-2} HHF loading (the same scenario as in fig. 6.2). A) shows the resulting surface temperature at the inner surface of the cooling channel, as seen in the cross section of the component cut parallel to the cooling channel. B) shows the heat fluxes resulting at the top of the cooling channel, viewed from below [171].

monoblock chains, the cooling performance of the test components is slightly lower. This is due to the use of swirl tapes in monoblock chains (see fig. 2.14), which leads to stronger turbulence in the cooling water and thus to a higher Reynolds number and cooling performance (see equations 4.17-4.19). In addition, monoblock chains can be operated at slightly higher cooling water pressures, temperatures and flow rates, which also increase the ability to remove thermal energy. As a result, when the test components are loaded with 14 MW m^{-2} heat fluxes, the surface temperatures are equivalent to the surface temperatures reached at 15 MW m^{-2} loading on monoblock chains (see table 6.1). Since the resilience of the LMD-w layers tested in these experiments is primarily relevant in the context of the loading of divertor components, within the next sections, exposures with surface temperatures on the test components equivalent to steady state 10 or 15 MW m^{-2} on monoblock chains are referred to as steady state 10 or 15 MW m^{-2} experiments.

6.2 Manufacturing and LMD-w coating of the test components

Three test components were manufactured, each with three tungsten double tiles, as illustrated in fig. 6.1 B) (with the exception of the two center tiles on component 3, which were not manufactured as a double tile but as single tiles).

The tungsten double tiles were cut by EDM from a double forged pure tungsten (DW) block from the manufacturer Plansee. This forging process caused the tungsten grains to become elongated. To achieve the desired T-orientation, the tungsten tiles were cut from the tungsten block so that the grains are elongated transversely to the plasma-facing surface. This transversal grain orientation minimizes the risk of individual grains detaching and entering the plasma during fusion reactor operation [111]. Therefore, the grain structure with T-orientation is the preferred variant in the context of tungsten as armor material of PFCs.

The top surfaces of the tungsten double tiles were coated using the LMD-w process with the process parameters evaluated as the most suitable in chapter 5 and specified in table 6.3. During the coating process, the tungsten double tiles were

Table 6.3: The following process parameters were used for the LMD-w deposition of layers on the tungsten double tiles [171].

Process parameter	Symbol	Welding beads	Unit
Energy density	E_n	477	J mm^{-2}
Laser power	P	5	kW
Laser spot diameter	d_s	2	mm
Scanning velocity	v_s	400	mm min^{-1}
Wire feed velocity	v_w	500	mm min^{-1}
Hatch distance	h_d	1.51	mm

clamped to a copper plate cooled with room temperature water to keep the substrate temperature as constant as possible and to avoid oxidation of the tile due to heat accumulation. Fig. 6.4 depicts a double tile coated in this way. This double tile, which measured 7 mm in height prior to coating, has a 4 mm deep and 0.5 mm wide gap in the center, dividing it into two areas of $28 \times 12 \text{ mm}^2$. The surface visible at the bottom left of the image was coated with a normal layer. The surface in the top right of the image was coated with a normal layer and then smoothed by remelting the coating via laser. The scanning direction during smoothing was the same as the deposition direction, since this had a greater effect than perpendicular to the scanning direction. The smoothing effect was described in detail in chapter 5 and is illustrated in fig. 5.4. After coating, the tiles of initially 7 mm height, measure

~7.65 mm height. After coating with two layers the double tiles have heights of ~8.2 mm. During laser remelting, the material is evenly distributed and flows from the raised areas into the valleys. Thus, the smoothing process reduces the maximum height of the layers by about 50 μm .



Figure 6.4: LMD-w single layers were deposited on the top surfaces of this tungsten double tile. On the left half one normal layer and on the right one smoothed layer was deposited, using the process parameters listed in table 6.3 [171].

The three copper cooling structures were manufactured using CNC milling. On the upper side of the cooling structures, 1.5 mm deep squared pockets are milled, into which the tungsten tiles are placed and brazed. AgCu28Ge2Co0.3 brazing alloy with a melting point of 780 $^{\circ}\text{C}$ was selected for brazing the tungsten tiles and copper heat sinks. This brazing alloy is only used for the purpose of mock-up fabrication and not suited for the use in real divertor components, due to the long term neutron activation of silver. After cleaning all components by etching, the 0.1 mm thick brazing foil is first placed in the designated pockets in the heat sink, and subsequently the tungsten tiles. The components prepared in this way are then subjected to heat treatment in a vacuum furnace. The components are brazed at a temperature of 850 $^{\circ}\text{C}$ for 15 minutes. The heating and cooling phases of this brazing process are controlled at rates of 13 K min^{-1} .

Fig. 6.5 shows the prepared components, connected by means of clamping elements to the cooling system of the electron beam facility JUDITH 2 shortly before their HHF exposure. The intended loading of the tiles is marked. On component 1 there

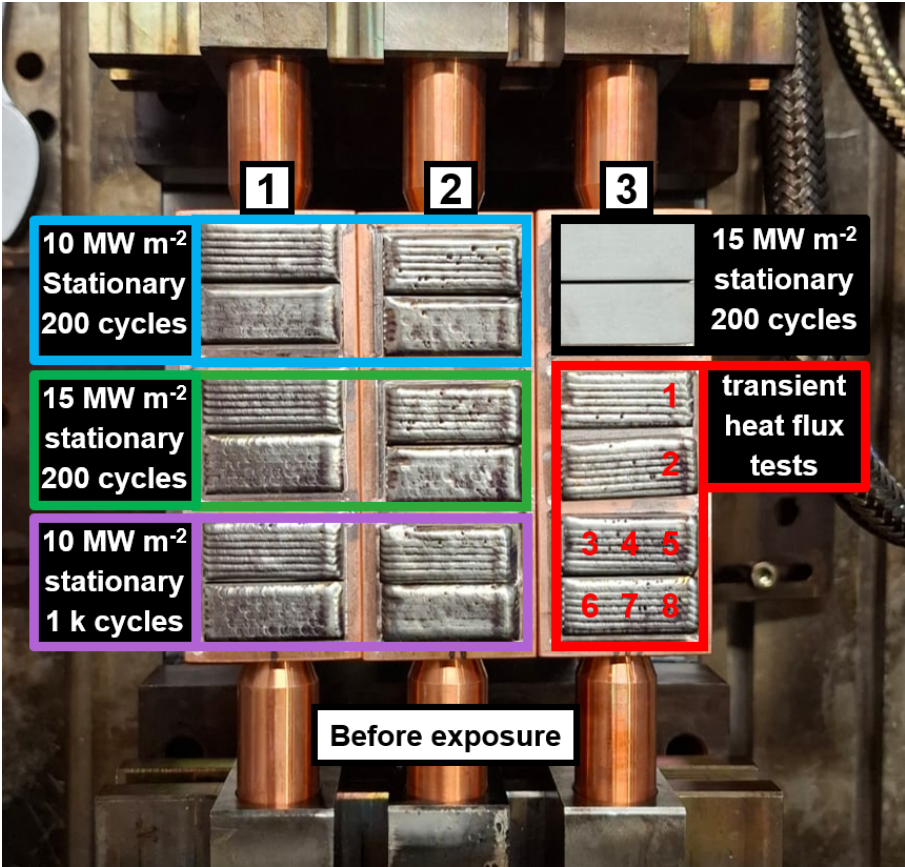


Figure 6.5: Here the test components are depicted prior to the HHF experiments. The three components featuring LMD-w tungsten coatings are clamped to the cooling system of the electron beam facility JUDITH 2. The double tiles of component 1 were each coated with one LMD-w layer and one smoothed LMD-w layer. On the double tiles of component 2, two stacked layers as well as two stacked and smoothed layers were deposited. And component 3 has uncoated reference surfaces at the top (black framed) and four single layer surfaces at the bottom (red framed). The red numbers indicate the positions where the combined steady state and transient thermal loads will act (according to table 6.5) [171].

are three surfaces coated with a normal layer and three with a smoothed layer. On component 2 there are three surfaces with two stacked layers and three with two stacked smoothed layers. On component 3 the black framed 8 mm (upper half) and 7 mm (lower half) high uncoated double tile serves as reference to the coated ones. Below them (red frame) are four surfaces with normal layers. The single tile marked

with a red 2 shifted slightly during the brazing process, but this had no negative consequences for the HHF experiments. Comparing the coatings of components 1 and 2, a higher defect density (pores and irregularities) of the stacked layers is visible. These defects are mostly due to minor process instabilities, and the possible contributing causes are addressed in the discussion section 6.5.

6.3 Steady state thermal load exposure

Experimental

The test components coated using LMD-w and connected to the cooling system of the electron beam facility are exposed to thermal loads in accordance with the exposure plan shown in fig. 6.5. The electron beam moves in a grid pattern (see section 4.2)) developed specifically for these tests over the top surfaces of the tungsten tiles that shall be exposed to thermal stresses.

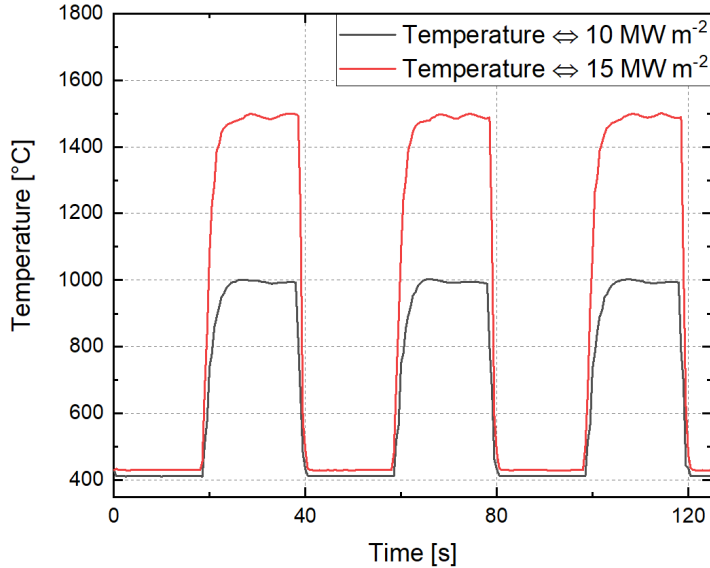


Figure 6.6: These are the temperatures measured by single-color pyrometry on the LMD-w coated (two stacked normal layers) surfaces of two tungsten double tiles of component 2 during cyclic steady state thermal loading [171].

In the steady state HHF experiments, two LMD-w coated double tiles per exposure scenario are loaded. Among these, there is, for each exposure scenario, one normal layer coated surface, one with a smoothed layer, one with two stacked normal layers and one with two stacked smoothed layers. In addition, one uncoated reference

double tile was loaded. In fig. 6.6, temperature data measured by single-color pyrometry at the surface of two coatings during the exposure is plotted. The exposure to steady state thermal loads is performed in cycles. One cycle consists of 20 seconds of loading phase and 20 seconds of no exposure phase. Each of these 40 second cycles is intended to simulate one pulse of reactor operation. The cyclically occurring temperature shifts due to the transitions between the electron beam exposure and no exposure phases are visible in fig. 6.6. During the loading phases, conditions according to table 6.1 are reached on the exposed tungsten surfaces. In the no exposure phases, the components cool down to roughly the temperature of the cooling water, which is 70 °C. In fig. 6.6, a minimum measured temperature of ~ 400 °C is plotted, since this is the lower limit of the measuring range of the used single-color pyrometer.

The two LMD-w coated tungsten double tiles framed in blue in fig. 6.5 are loaded with 200 cycles of 10 MW m^{-2} . The two green framed double tiles are loaded with 200 cycles of 15 MW m^{-2} . The purple framed double tiles are loaded with 1000 cycles of 10 MW m^{-2} . And the uncoated reference, framed in black in fig. 6.5 is subjected to 200 cycles of 10 MW m^{-2} .

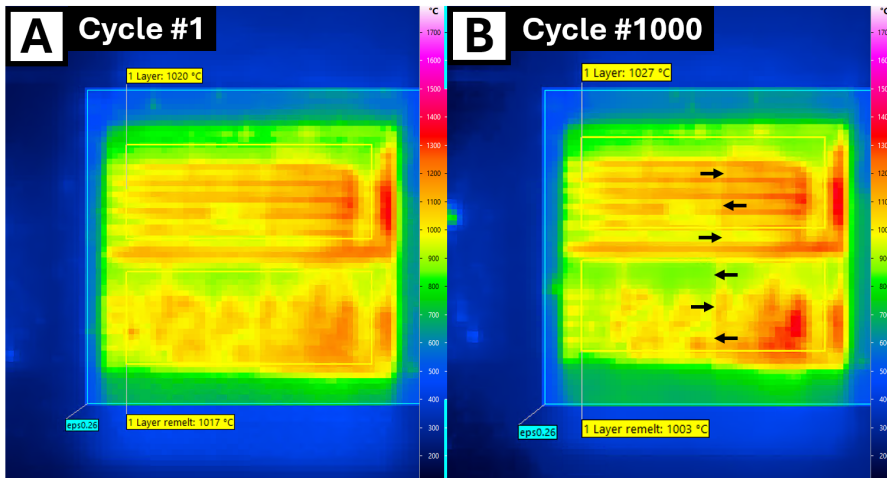


Figure 6.7: IR thermography images of LMD-w coated tungsten tiles at the 1st and 1000th exposure cycle of the steady state 10 MW m^{-2} HHF experiments. The black arrows in B) indicate the position of a crack running parallel to the cooling channel [171].

Fig. 6.7 displays exemplary for all coatings exposed to the 10 MW m^{-2} scenarios IR camera images taken during during the 1st and 1000th exposure cycle. For the 10 MW m^{-2} experiments, the average surface temperature of all LMD-w coated tungsten tiles was measured to be ~ 1000 °C across all cycles. During these experi-

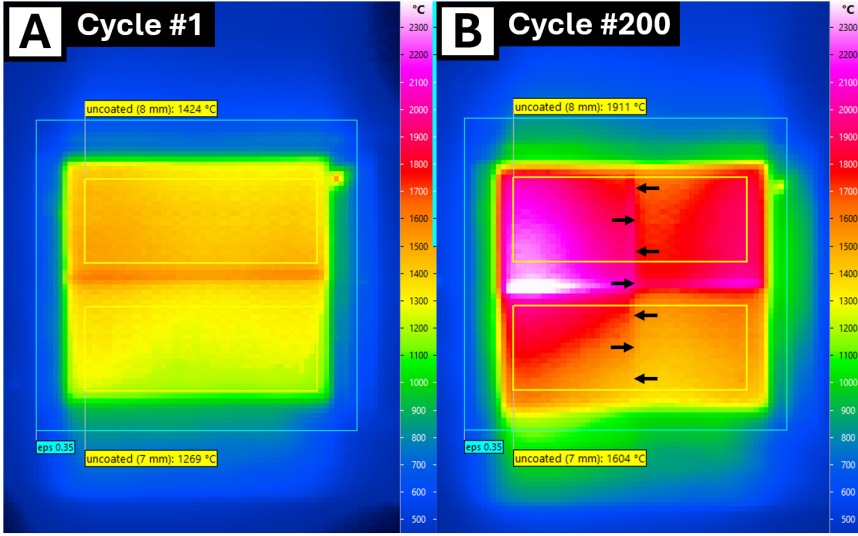


Figure 6.8: IR thermography images of the uncoated reference tiles at the top 8 mm thick tungsten at the bottom 7 mm at the 1st and the 200th exposure cycle of the 15 MW m^{-2} HHF experiments. The black arrows in B) indicate the position of a crack running parallel to the cooling channel [171].

ments, the IR camera was calibrated by comparing the measurements taken with the IR camera and the measurements taken with a two-color pyrometer, both pointed at the exposed area. The emission coefficient used in the IR camera system was then adjusted accordingly. Due to changes in surface structure and temperature, the emission coefficient of the exposed surface may change during the exposure, making further adjustments necessary.

Fig. 6.8 displays IR camera images taken during the exposure of the uncoated reference double tile during the 1st and 200th exposure cycle of the 15 MW m^{-2} scenario. In this picture, the upper part of the tungsten double tile, which is 8 mm high, reaches a higher surface temperature than the lower 7 mm high part. It should be noted that the thermally loaded tungsten surfaces displayed in fig. 6.8 (uncoated) and fig. 6.9 (LMD-w coated) increased their measured average surface temperatures during the exposure with increasing cycle count as plotted in fig. 6.10. On the IR image at cycle 200 of fig. 6.8, a crack can be recognized, which runs centrally through both halves of the double tile parallel to the cooling channel. Furthermore, the left half of the upper surface in the cycle # 200 image shows an increased temperature. These rises in temperature with increasing cycle count is due to local delamination of tungsten and copper at the brazing joint, as visible in fig. 6.14 C).

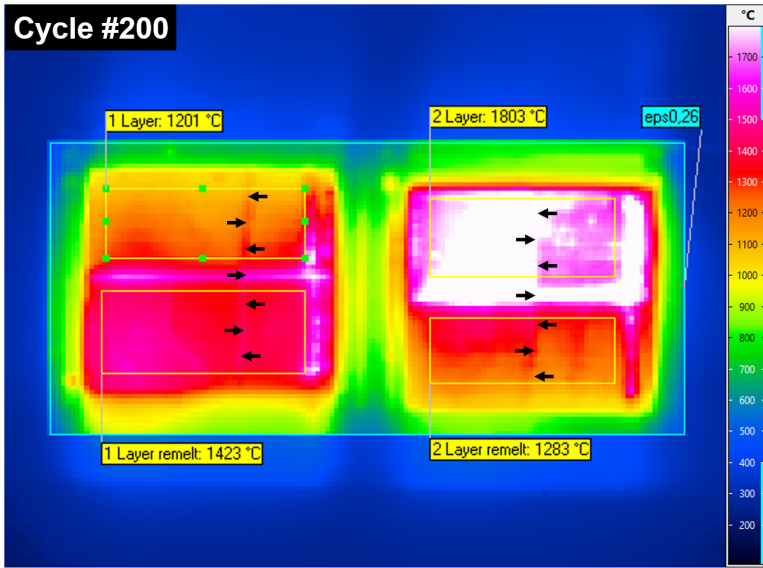


Figure 6.9: Shown here are IR thermography images of the LMD-w coated tungsten tiles at the 200th exposure cycle of the steady state 15 MW m^{-2} HHF experiments. The black arrows indicate the locations of cracks running parallel to the cooling channel. The measured relative increases of the temperatures with increasing load cycles of the various coated and uncoated tungsten surfaces are visualized in fig. 6.10 [171].

Fig. 6.9 depicts IR camera images of two coated double tiles during the 15 MW m^{-2} exposure experiments. The upper left half was coated with 1 normal layer, the lower left with 1 smoothed layer, the upper right with 2 stacked normal layers, and the lower right with 2 stacked smoothed layers. Similar cracks to those on the reference double tile can be observed on the coated ones when subjected to the same thermal loading scenario (compare fig. 6.8 and fig. 6.9). Both double tiles have a central crack running parallel to the cooling channel.

Results

The condition of the three components after the thermal load tests is depicted in fig. 6.11. On the one hand, cracks can be observed on the steady state loaded tungsten tiles that run centrally parallel to the cooling channel. On the other hand, changes in the surface structure are visible. The cracks were examined by microscopy

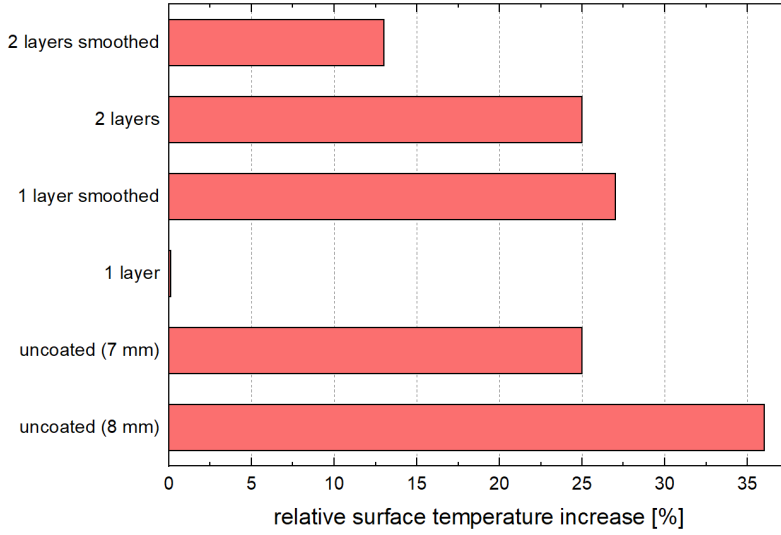


Figure 6.10: Visualized are the relative temperature increases measured on various LMD-w coated and uncoated tungsten surfaces during cyclic steady state HHF experiments. Here, the temperature increases refer to the difference between the measured surface temperatures at the 1st and 200th exposure cycles of the 15 MW m^{-2} HHF studies [171].

of the cross sections. And to examine the change in surface structure, roughness measurements were taken before and after the HHF exposure.

The crack on the reference tiles visible on the IR image at cycle number 200 in fig. 6.8 is also visible optically, as shown in the photograph in fig. 6.11 (black framed tiles). A cross-sectional micrograph perpendicular to the cooling channel was prepared for these tiles. The cross section of the 8 mm high tungsten tile is depicted in fig. 6.12 A). It is clearly visible that the crack runs transversely to the surface of the tungsten all the way to the copper heat sink and ends there. It needs to be mentioned that the crack can be vaguely seen on the IR images after the first ~ 20 cycles of the 15 MW m^{-2} exposure. Due to the further cyclic loading, the crack and its gap then grew with increasing number of cycles. The coated tiles also exhibit cracks similar to those on the uncoated reference tile after the 15 MW m^{-2} load with 200 cycles. Before the exposure, no cracks are visible on the uncoated tiles at first glance in the figure. However, on closer inspection, it was found that after the LMD-w coating, there were few macrocracks as discussed in chapter 5. These cracks are located in the center of the tiles and run parallel to the cooling channel. On the IR images, these cracks are already schematically visible during the first exposure cycles. Similar to the crack in the uncoated reference tile, the crack and

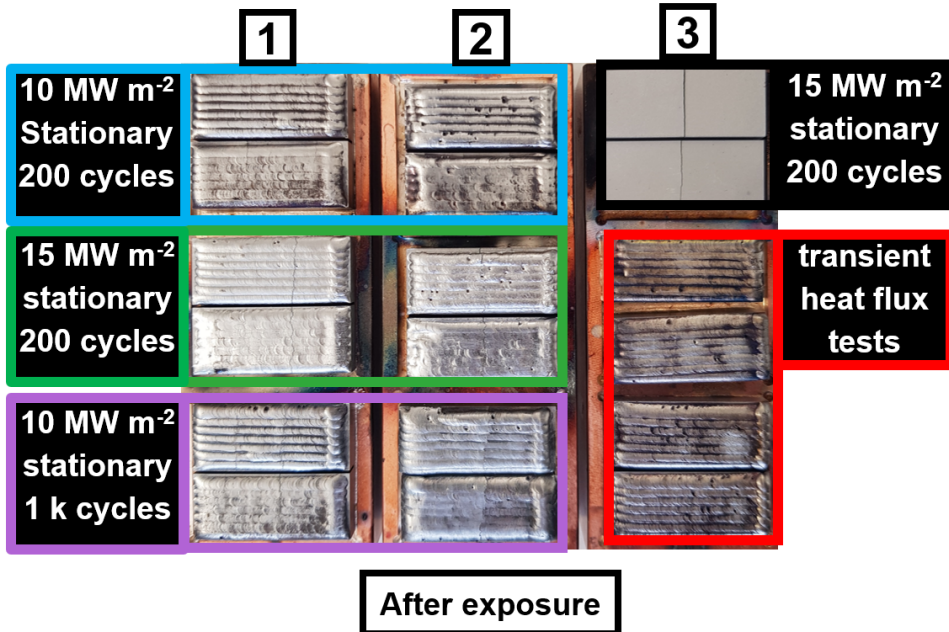


Figure 6.11: The three components are depicted after the HHF experiments. It is marked in color which tungsten tiles were subjected to which stress. The steady state loaded uncoated and coated layers (black, green, purple) show cracks in the middle of the tile parallel to the cooling channel. The blue framed layers show no directly visible changes. The combined steady state and transient loaded tiles highlighted in red exhibit localized surface changes at the stressed positions (according to fig. 6.5 and table 6.5) [171].

the gap then grew with increasing number of cycles. The result is visible in fig. 6.11 on the tiles framed in green. In fig. 6.12 B), the crack is then visible in a tile that has been coated and smoothed. It is evident that the crack in the coated tile also runs transversely to the surface down towards the copper heat sink and ends there. However, the resulting gap is not as wide as the one in the uncoated tile. Furthermore, another small crack is visible in the coated tile in fig. 6.12 B) on the right, but it only reaches 2 mm in depth. The coatings also contain small pores as visible in the cross-sectional micrographs in fig. 6.13. However, the connection between the coating and the substrate appears to be flawless, even after cyclic HHF exposure. Also, the connection between the stacked tungsten layers (see fig. 6.13 C) and D)) is excellent. These small cracks and pores appeared during the LMD-w process as discussed in chapter 5.

In fig. 6.11, very fine cracks are visible in the tiles framed in blue, which were

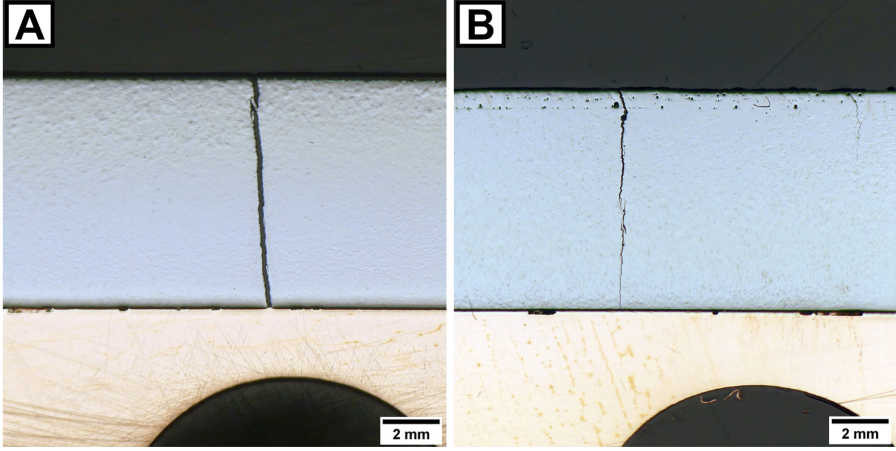


Figure 6.12: The images show the cross section of A) an uncoated reference tungsten tile and B) a coated tile, both after exposure to 200 cycles of 15 MW m^{-2} [171].

subjected to 200 cycles of 10 MW m^{-2} loading. These cracks also run centrally parallel to the cooling channel. These are the cracks that occurred during deposition (as discussed in chapter 5) and have grown slightly during the cyclic HHF experiments. This type of crack is also visible in the tiles framed in purple, however they have grown larger due to the exposure to a higher amount of cycles. These purple framed tiles were subjected to 1000 cycles at 10 MW m^{-2} . The position of such a crack is also indicated with black arrows on the IR camera image of fig. 6.7 B).

The images in fig. 6.14 display the cross section of the components and were taken using SEM. The joint between tungsten and copper brazed with silver alloy is mostly flawless as shown in fig. 6.14 A). In some locations, small pores such as those shown in fig. 6.14 B) are visible within the brazing solder. These pores are local defects that most likely occurred during the brazing process. Another type of defect at the brazing joint is delamination, which occurred locally and is depicted in fig. 6.14 C).

The average surface roughness values sR_a (see equation 4.11) and sR_q (see equation 4.12) were measured on the surface of all tiles by laser profilometry. In addition to the standard roughness value sR_a , the value sR_q was also determined, since this value reacts more sensitively to individual spikes and thus responds more sensitively to defects such as cracks. The roughness values of the uncoated and coated surfaces were measured before and after the HHF experiments and their respective differences, ΔsR_a and ΔsR_q , obtained. Table 6.4 lists the changes of these roughness values due to the steady state HHF experiments for the various surfaces. The average roughness increase of the coated surfaces is with $\Delta sR_a \approx 1 \text{ }\mu\text{m}$ ($\Delta sR_q \approx 3.5 \text{ }\mu\text{m}$)

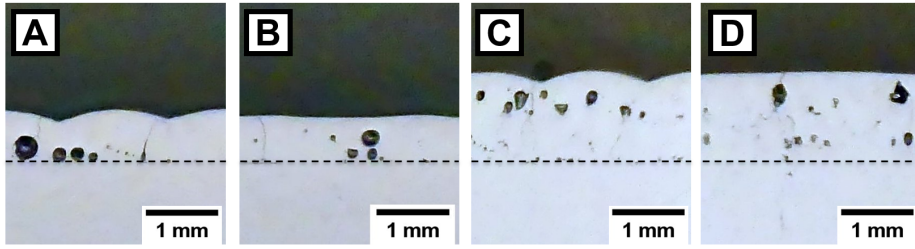


Figure 6.13: These are cross-sectional micrographs of various LMD-w coatings after exposure to 200 cycles of 15 MW m^{-2} . The plane of the cross section lies perpendicular to the deposition direction. These tungsten double tiles are framed in green in fig. 6.11. The areas above the black dotted line represent the LMD-w coating, with the substrate underneath. A) displays a single layer, B) a smoothed single layer, C) two stacked layers, and D) two stacked smoothed single layers [171].

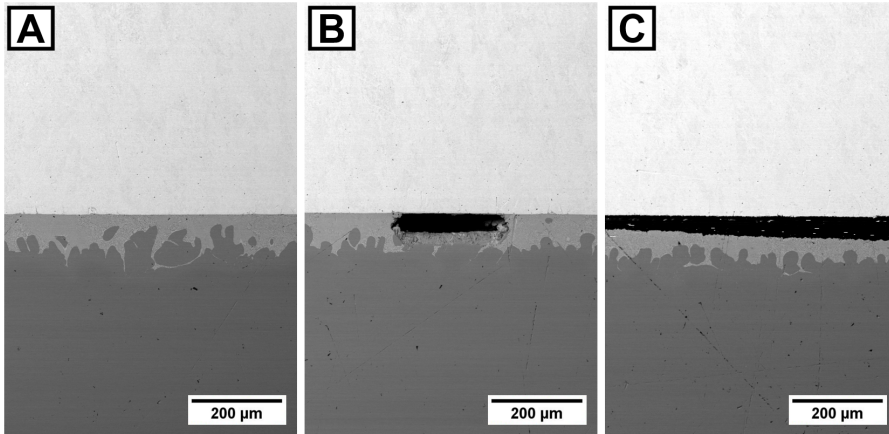


Figure 6.14: Illustrated are selected SEM images of cross sections of the joint brazed using silver brazing alloy (AgCu28Ge2Co0.3) between tungsten at the top and copper at the bottom. The majority of the joints are as shown in A), with an intact joint between tungsten and copper. But in some cases pores formed at the brazed joint, which occur sporadically and are visible in B). C) shows delamination in a small area at the edge of the brazed joint of a double tile after exposure to steady state thermal loading of 200 cycles at 15 MW m^{-2} [171].

after the 200 cycles and 15 MW m^{-2} loading scenario the highest compared to the other loading scenarios. The second largest increase in surface roughness measured at $\Delta sR_a \approx 0.5 \text{ μm}$ ($\Delta sR_q \approx 1 \text{ μm}$) is observed for the 10 MW m^{-2} and 1000 cycles

Table 6.4: Listed are the measured roughness increases ΔsR_a and ΔsR_q of the surfaces after exposure to various steady state heat fluxes and exposure cycles. Here, the average surface roughness measured prior to the HHF exposure scenarios is $sR_a = 1.5 \mu\text{m}$ ($sR_q = 1.9 \mu\text{m}$) for uncoated tungsten and $sR_a = 6.8 \mu\text{m}$ ($sR_q = 11.7 \mu\text{m}$) for the coatings. The surface characteristics of LMD-w coatings prior to HHF exposure are discussed in more detail in chapter 5 [171].

Heat flux [MW m^{-2}]	10		10		15	
Number of cycles	200		1000		200	
Surface roughness [μm]	ΔsR_a	ΔsR_q	ΔsR_a	ΔsR_q	ΔsR_a	ΔsR_q
Uncoated reference	-	-	-	-	0.17	0.25
1 Layer	0.3	0.5	0.97	1.15	0.7	2
1 Layer (smoothed)	0.1	0.27	0.1	0.87	0.9	4.14
2 Layers	0.2	0.9	0.59	1.1	0.9	3.1
2 Layers (smoothed)	0.2	0.34	0.38	0.93	1.75	5.06

scenario. The smallest change in surface roughness on the coated surfaces was measured at $\Delta sR_a \approx 0.2 \mu\text{m}$ ($\Delta sR_q \approx 0.5 \mu\text{m}$) after the 10 MW m^{-2} and 200 cycles scenario. These values are higher than the roughness changes of the uncoated surfaces with $\Delta sR_a = 0.17 \mu\text{m}$ ($\Delta sR_q = 0.25 \mu\text{m}$) at 15 MW m^{-2} and 200 cycles. But overall, all measured roughness increases for the cyclic steady state loading tend to be relatively low.

6.4 Combined steady state and transient thermal load exposure

Experimental

In the case of ELMs (see section 2.3), heat fluxes in the GW m^{-2} range impinge on the surface of the PFCs. The influence of the combined steady state and transient thermal loads that act on the armor of PFCs during ELMs was also tested on LMD-w coated components as part of these HHF experiments. The four single layer tungsten coatings, framed in red in fig. 6.5, were subjected to the loads at the positions indicated by the red numbers in accordance with the plan specified in table 6.5. The electron beam moves in a grid pattern over the surface of the tiles to generate the steady state thermal loading and the resulting desired base temperatures T_{base} at the surface of the tungsten coating, as listed in table 6.5. Additionally, the transient loading is achieved locally with the circular beam loading method as discussed in chapter 4.2 [80, 256]. On the basis of equation 2.16, the heat flux factor F_{HF} , which is used as a measure of the rise in temperature in relation to the base temperature T_{base} , can be calculated using the pulse length Δt and intensity I . Experiments

Table 6.5: The positions visible with the red numbers in fig. 6.5 are subjected to combined steady state and transient heat loads according to the parameters specified in this table [171].

Position #	Pulses	Pulse length Δt [ms]	Heat flux factor F_{HF} [MW s ^{0.5} m ⁻²]	Intensity I [GW m ⁻²]	Base temperature T_{base} [°C]	Peak temperature T_{max} [°C]
1	10 ⁵	0.5	3	0.13	700	900
2	10 ⁵	0.5	6	0.27	700	1050
3	10 ³	0.5	12	0.55	700	1450
4	10 ⁴	0.5	12	0.55	700	1450
5	10 ⁵	0.5	12	0.55	700	1450
6	10 ⁵	0.5	3	0.13	200	400
7	10 ⁵	0.5	6	0.27	200	550
8	10 ⁵	0.5	12	0.55	200	950

were conducted with $F_{HF} = 3, 6$ and $12 \text{ MW s}^{0.5} \text{ m}^{-2}$ at base temperatures of 200 and 700 °C and different numbers of pulses. The temperature increase ΔT between the base temperature T_{base} and the peak temperature T_{max} can be calculated using equation 2.14.

Results

Fig. 6.11 displays the coated tungsten tiles, which are framed in red, after the combined steady state and transient thermal exposure tests. Partially circular surface modifications are visible. Table 6.6 lists the changes in roughness before and after the exposure of the coated surfaces, ΔsR_a and ΔsR_q . The sR_q values are somewhat higher than the sR_a values, as the R_q reacts more sensitively to individual peaks, as they occur while measuring cracked surfaces. The highest changes in surface roughness were measured at the exposure with 10^5 pulses at $F_{HF} = 12 \text{ MW s}^{0.5} \text{ m}^{-2}$ and 700 °C base temperature. Fig. 6.15 B) displays the surface of the coating subjected to this load. In direct comparison, fig. 6.15 A) shows an uncoated ITER grade Plansee (IGP) tungsten (W) tile from the experiments discussed in chapter 7, that was subjected to the same loading scenario. The cracks in the uncoated surface appear larger and deeper than on the coated tile, which is supported by the increased ΔsR_q value of 23 μm for the uncoated surface compared to 14 μm for the coated surface. At 16 μm , the ΔsR_a value for the uncoated tile is also higher than for the coated tile, where the ΔsR_a was measured at 10 μm . In direct comparison, the surface changes caused by combined steady state and transient loads on the coated surfaces are in a similar range, as those on uncoated surfaces. However, on coated surfaces the exposure to these loads resulted in fewer cracks of significant size.

Table 6.6: Listed are the measured roughness increases ΔsR_a and ΔsR_q of the coated surfaces after exposure to various combined steady state and transient thermal loads. The sR_q values are higher, as these are more sensitive to cracks compared to the sR_a values [171].

Pulses	Base tem- perature	$I = 0.13 \text{ GW m}^{-2}$		$I = 0.27 \text{ GW m}^{-2}$		$I = 0.55 \text{ GW m}^{-2}$	
#	[°C]	ΔsR_a [μm]	ΔsR_q [μm]	ΔsR_a [μm]	ΔsR_q [μm]	ΔsR_a [μm]	ΔsR_q [μm]
10^3	700	-	-	-	-	0	0
10^4	700	-	-	-	-	2.35	3.25
10^5	700	0.08	0.15	0.98	1.21	9.92	14.44
10^3	200	0	0.03	0.36	0.48	0.76	1.47

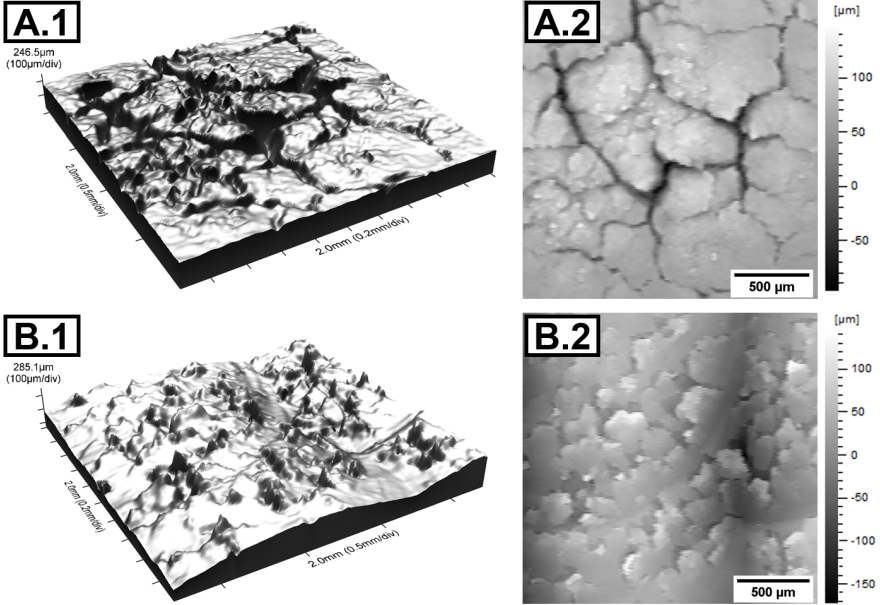


Figure 6.15: Profile measurements of surfaces after combined steady state and transient loading (0.55 GW m^{-2} ; $F_{HF} = 12 \text{ MW s}^{0.5} \text{ m}^{-2}$; at $700 \text{ }^{\circ}\text{C}$ base temperature) are visualized. A.1) shows the isometric view of a flat uncoated reference tungsten surface from the experiments in chapter 7 after exposure and A.2) in top view. B.1) shows the isometric view of a surface coated by LMD-w after exposure and B.2) in top view.

6.5 Discussion

Coating of the components by LMD-w

It was successfully demonstrated that tungsten surfaces measuring $28 \times 12 \text{ mm}^2$ can be coated with tungsten using LMD-w with good reproducibility. Even when subjected to various HHF scenarios, the coatings did not flake off. The bonding between the coating and the substrate was still good with a seamless transition even after the stress tests. This excellent bonding ensured efficient heat transfer from the coating to the substrate.

However, pores and cracks are visible in the cross-sectional micrographs of the coatings (see fig. 6.13) that were prepared after HHF exposure. Some of these cracks were probably already present in the coatings prior to the HHF exposure and may have been caused by high temperature gradients during the LMD-w process as discussed in chapter 5. During cooling after deposition, these temperature gradients cause high residual tensions in the material, which can release in the form of cracks. This type of crack can be avoided by flattening the temperature gradients between the melt pool and the substrate material.

In addition, pores are visible in the cross sections of the coatings (see fig. 6.13), which are distributed without any clear pattern. The origin of these pores is most likely due to process instabilities during LMD-w in the form of lateral oscillations of the wire as it is fed into the melt pool (see fig. 3.8). Suboptimal alignment of the inert gas nozzle could also lead to pore formation (fig. 3.7). By reduction of the distance between the wire nozzle and the melt pool, the amplitude of these oscillations can be reduced and thus the process stability improved. However, this is only possible to a certain extent, as the distance must remain large enough to prevent the wire nozzle from melting. Under the same caveat, the inert gas nozzle should also be positioned closer to the melt pool to ensure sufficient and uniform inert gas coverage of the melt pool. Studies of this process under vacuum should be conducted in the future to investigate whether certain defects are attributable to insufficient shielding from ambient gases or the inert gas flow.

In the case of stacked layers, unevenness in the underlying layers can also lead to process instability due to wire deflection. This effect results in a slightly increased defect density for stacked layers (see fig. 6.5 and fig. 6.13). However, the connection between the stacked layers is seamless, enabling thermal energy to dissipate to the substrate properly during HHF exposure.

Steady state thermal load exposure

The resilience of LMD-w coated tungsten double tiles under cyclic exposure to steady state thermal loads was tested and compared with the performance uncoated refer-

ences. Each cycle is intended to simulate the effect of one plasma discharge during fusion reactor operation, at least in terms of the resulting thermal stresses. First of all, it should be mentioned that the LMD-w tungsten layers retained their flawless connection to the underlying tungsten across all exposure scenarios (see fig. 6.11 and fig. 6.13). Due to the good connection to the substrate, the LMD-w coatings were able to fulfill their main task as PFM, which is to withstand the incoming stresses and transfer the thermal energy to the underlying cooling structures. Furthermore, no loss of material of the coatings due to spalling was observed. With regard to resilience to cyclically occurring thermal loads, no difference could be detected between coatings with one layer and two layers, or nonsmoothed and smoothed layers. It has already been demonstrated (see chapter 5) that layer thickness of 2 mm can be achieved by stacking four layers [169]. However, this does not represent the maximum achievable layer height, but merely one value that has been demonstrated so far. Compared to alternative coating processes such as plasma spraying or sputtering, the freedom offered by the LMD-w process in terms of coating height adjustment is an advantage. In addition to the advantages already mentioned in chapter 3 the LMD-w process is able to deposit material under atmospheric conditions, because it only needs local inert gas shielding. The deposition is even possible on cold substrates, although this is not ideal. And probably most importantly, LMD-w tungsten coatings are able to resist fusion-relevant HHFs due to their seamless connection to the substrate. The following section discusses how the various coatings performed under different thermal loading scenarios. The following discusses how the various LMD-coatings performed under the respective exposure scenarios.

Across all cycles of the 10 MW m^{-2} loading scenario, the measured average temperature at the surface of the coatings was $\sim 1000 \text{ }^{\circ}\text{C}$ (see fig. 6.7). This applies to both the coatings loaded with 200 cycles (green framed in fig. 6.11) and those loaded with 1000 cycles (purple framed in fig. 6.11). After a while, cracks running parallel to the cooling tube became visible on the IR camera images of the 1000 cycle ones (see fig. 6.7). These cracks grew with increasing cycle count, so after the exposure scenario, cracks were clearly visible in the purple framed coatings of fig. 6.11. Unfortunately, it is not possible to determine with certainty whether the cracks were caused by the HHF exposure or originated earlier during the deposition of the material. However, the position of the cracks on the double tiles (centrally) and their orientation (parallel to the cooling channel) indicate that these cracks arose due to the HHF exposure and not the LMD-w coating. A combination of factors contributed to this crack development. On the one hand, the cooling channel located centrally under the tungsten double tiles resulted in uneven temperature distribution within these (see fig. 6.2 B)). At the edges of the loaded surface of the tungsten tiles, which are located furthest from the cooling channel, the temperatures are higher than in the center. Furthermore, the alternating exposure and no exposure phases cause temperature shifts of $\sim 900 \text{ K}$. These temperature shifts cause resid-

ual tensions in the material due to temperature-induced expansion and shrinkage. This uneven temperature distribution in combination with the cyclic temperature shifts could therefore be the origin of these cracks and is in any case the reason for their growth. Further factors influencing crack development are the dimensions of the tungsten double tiles, as larger components are associated with larger absolute temperature-induced dimensional changes and thus greater stresses. The quality of the selected tungsten material is also a factor which can influence crack development.

In the cyclic steady state 15 MW m^{-2} HHF exposure experiments, after few cycles cracks were already visible on the IR camera images of the exposed tungsten double tiles. These cracks occurred both on the uncoated reference (fig. 6.8) and the coated double tiles (fig. 6.9). These cracks run parallel to the cooling channel and are similar to those observed during the 10 MW m^{-2} 1000 cycle exposure scenario. Since these cracks also appeared on the uncoated reference, the origin of this defects seems to be independent of the LMD-w coatings. The previously discussed factors, such as uneven temperature distribution in the test components, cyclical temperature shifts, component dimensions, and the tungsten material quality, are probably responsible for the development of these cracks. Due to the significantly higher temperature shifts of $\sim 1400 \text{ K}$ between the exposure and no exposure phases in the 15 MW m^{-2} experiments, larger residual tensions arise in the material. This has a facilitating effect on crack growth progressing with increasing cycle count. Cross-sectional micrographs were prepared on planes perpendicular to the cooling channel, showing these cracks in the uncoated reference (fig. 6.12 A)) and a basic single layer coated tile (fig. 6.12 B)). This type of crack, running perpendicular to the cooling channel and perpendicular to the loaded surface, has also been observed on monoblocks, as reported in the literature [288–293]. However, in comparable studies, this type of crack has only been observed on monoblocks under cyclic HHF loads of $\sim 20 \text{ MW m}^{-2}$. The fact that these cracks already occurred in the experiments discussed here at lower cyclic steady state heat fluxes could be due to the specific tungsten material used for the double tiles. It was initially expected that the coatings rather than the test components themselves would be the weak point. Accordingly, due to availability, double forged pure tungsten (DW) was used for the tungsten double tiles of the test components instead of the IGP tungsten that is typically used for monoblocks [81, 294]. Both the DW and IGP tungsten materials are 99.97 wt% pure and manufactured by the supplier Plansee SE, Austria [80, 294]. However, the composition of the impurities present in the materials may vary slightly between the two variants. Both materials exhibit an elongated grain structure transversal to the surface loaded in the HHF experiments. Since this is the preferred grain orientation for tungsten as PFM to inhibit the growth of longitudinal cracks and loss of armor material [111]. Due to different forging processes, the DW material features pancake shaped grains, while the IGP tungsten features needle like shaped grains. The factors of impurities and grain structure can

influence the behavior of these tungsten materials towards fusion relevant thermal loads and were most likely the reason for the development of these cracks at loads below 20 MW m^{-2} [81].

As visible on the IR camera image of fig. 6.8 B), the measured temperature on the uncoated double tile, during loading cycle 200 of the 15 MW m^{-2} exposure scenario, is significantly higher in the upper left area, than in the right. This is most likely caused by partial delamination at the W/Cu brazing joint, as depicted in fig. 6.14 C). Unfortunately at this point in time, it cannot be determined whether the delamination was caused by the cyclic HHF exposure or whether it already developed during the brazing process and grew as a result of the thermal loading scenario. Such partial delamination can lead to heat accumulation in certain areas due to reduced conduction of thermal energy to the cooling component. Since the tungsten double tiles did not show elevated temperatures in the areas where delamination was later detected at the W/Cu joint during the first load cycles on the IR camera images, it can be assumed that the delamination first developed during the HHF exposure scenarios. To detect such defects in a non-destructive manner prior to HHF experiments, the IRINA (InfraRed Inspection for Non-destructive Analysis) test facility could be used in future experiments of this type [295]. On the IR camera images of the coated double tiles at exposure cycle 200 with 15 MW m^{-2} loading depicted in fig. 6.9, an elevated temperature was measured on the upper half of the right double tile. The elevated temperatures in certain regions became even more amplified with rising cycle count (fig. 6.10). To quantify the respective cooling performances during the 1st and 200th loading cycles of various uncoated and coated surfaces, the cooling constants were determined. To determine these cooling constants, two-color pyrometry data, obtained during cooling after switching off the thermal load, was used. The duration, over which the cooling curves behave as exponential decay functions, is the respective cooling constant [s]. Thus, the cooling constants presented in fig. 6.16 indicate how quickly the tungsten tiles cooled off after the first and last loading cycle. For the 1-layer coating, the cooling constant did not change between the first and last loading cycle. But, for all other surfaces of the 15 MW m^{-2} exposure scenario, the cooling constants increased with rising cycle count, which means that the respective cooling performance deteriorated. Since this behavior was observed for both coated and uncoated tungsten double tiles, the coating is probably not the decisive factor for the decrease in cooling performance. Instead, this effect is most likely the result of the increasing local delamination at the W/Cu brazing joints of the test components.

The influence of the cyclically acting steady state thermal loads on the various tungsten surfaces was quantified by laser profilometry measurements in the condition before and after the exposure. The determined changes in the surface topology are expressed here as roughness changes using ΔsR_a and ΔsR_q (see 4.5) in table 6.4.

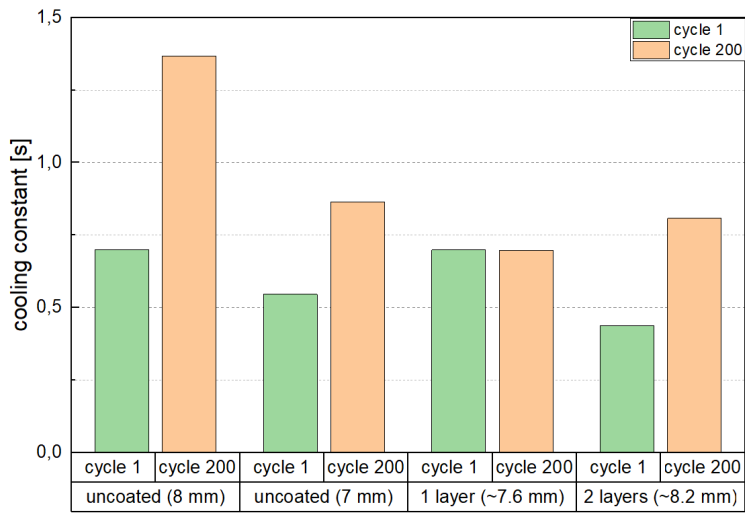


Figure 6.16: The measured cooling constants of various coated and uncoated tungsten surfaces are plotted for the 1st and 200th loading cycle. The cooling constant is a measure of the ability of the actively cooled test components to return to the coolant temperature after HHF exposure [171].

In the experiments with the highest intensity of 15 MW m^{-2} and 200 cycles , the roughness changes were the highest relative to the other thermal loading scenarios. This was followed by the 1000 cycles of 10 MW m^{-2} scenario, and the lowest increase in roughness was measured for the 10 MW m^{-2} 200 cycle scenario. Thus, the increase in roughness correlates positively with the applied intensity, as well as the cycle count and exposure time of the thermal loading scenario. The increases in roughness on coated surfaces was measured to be higher than for uncoated surfaces. However, when comparing the increases in roughness after steady state HHFs with those measured after combined steady state and transient thermal loads, it becomes clear that the here measured steady state roughness increases are all relatively low and rather insignificant. These low increases in roughness during steady state HHF exposure are another indication of the high resilience of LMD-w coatings against spalling and crumbling. Due to the seamless bond between the coatings and the substrate, the coatings were expected to perform well in these experiments. According to the literature, pure tungsten only experiences significant roughness increase during steady state thermal loading, if it gets exposed to high power densities over long time periods, which then leads to microscopic and macroscopic plastic deformations as a result of creep [296].

Combined steady state and transient thermal load exposure

The behavior of LMD-w coated PFCs when subjected to combined steady state and

transient thermal loads was tested on the four single layer coated surfaces framed in red in fig. 6.5. Thermal loads were applied at the eight positions marked with red numbers in accordance to the exposure plan of table 6.5. The condition of the test components after HHF exposure is depicted in fig. 6.11. These HHF studies were conducted to observe the influence of ELMs on the LMD-w tungsten coatings.

First of all, it should be noted that despite the exposure to high thermal stresses, no tungsten coating visibly delaminated, spalled, or crumbled. In addition, the exposed coatings were able to dissipate the externally applied thermal energy over the entire duration of the experiments, since they kept their good connection to the substrate. Also, for these thermal shock experiments no undesired surface temperature rise was detected with advancing exposure duration, as was the case in parts of the HHF studies with cyclically occurring steady state loads of 15 MW m^{-2} (see fig. 6.8 and fig. 6.10).

At some of the loaded positions, surface changes are visible in the form of circular gray spots (fig. 6.11). Thermal shocks can cause surface changes on tungsten due to the well-known damage mechanism, caused by thermally induced cyclic expansion and contraction of the material (as schematically illustrated in fig. 2.10). The damage usually ranges from minor surface modifications to small cracks or even crack networks [80,81,109,297]. In the case of thermal shocks, the criterion for crack formation can be described by equation 2.19. If the temperature shift ΔT [K] in the material caused by thermal shocks exceeds a certain value, the resulting residual tensions exceed the ultimate tensile strength of the material σ_{UTS} [Pa] and cracks develop [115]. Since temperature shifts ΔT increase with rising F_{FH} , the thermal shock scenarios with high F_{FH} (table 6.5 and fig. 6.5) feature the strongest surface modifications and crack formation (fig. 6.11).

Formation of these small cracks can be detected at the surface in the form of roughening. The resulting changes in surface roughness before and after the HHF exposure at the loaded positions are quantified as ΔsR_a and ΔsR_q by profile measurements and specified in table 6.6. Since R_q is more sensitive to cracks than R_a , the measured ΔsR_q is higher than the ΔsR_a across all samples. Fig. 6.15 displays the surface topologies of A) IGP tungsten and B) single layer LMD-w tungsten coating detected by laser profilometry after combined steady state and transient thermal loading (10^5 pulses; $F_{FH} = 12 \text{ MW s}^{0.5} \text{ m}^{-2}$; 700°C base temperature). The gray scale of the unsmoothed LMD-w single layer in fig. 6.15 A.2) is larger than that of the IGP tungsten B.2) simply because of its higher surface waviness. The IGP tungsten seems to contain slightly larger cracks than the LMD-w coated surface. Nevertheless, the damage pattern in the two tungsten variants appears to be rather comparable, which is confirmed by the measured ΔsR_a and ΔsR_q values.

Several factors may have contributed to the slightly different appearance of the IGP tungsten and the LMD-w single layer coating after thermal shock exposure (fig. 6.15). The microstructure of tungsten can influence its mechanical properties and thus its behavior under thermal shock loads. Table 6.7 lists some relevant characteristics of the two tungsten variants for this context. The term grain structure

Table 6.7: Comparison of characteristics of IGP tungsten and LMD-w tungsten [171].

Characteristics	IGP W	LMD-w W
Manufacturing history	forging	solidification
Grain structure	transversal	transversal
Grain size [μm]	550	4000
Grain aspect ratio	3.2	2
Hardness Vickers [HV1]	~ 440	~ 370

refers to the axis along which the grains are elongated in relation to the thermally loaded surface. IGP tungsten receives its grain structure through customized forging, which contributes to strengthening of the material. Hence, forged tungsten offers a higher σ_{UTS} , which, according to equation 2.19, leads to a higher damage threshold when exposed to thermal shocks. In comparison, recrystallized and solidified tungsten both exhibit lower σ_{UTS} , and thus lower damage thresholds during thermal shock loading. Furthermore, it should be noted that impurity accumulations during recrystallization normally lead to weakening of the grain boundaries and overall weakened material cohesion. LMD-w tungsten generally exhibits properties similar to recrystallized tungsten and is not as strong as IGP tungsten, which implies that it could fracture at lower thermal loads [81, 115]. Yet the results of the thermal shock experiments show no weaker behavior for LMD-w tungsten. For LMD-w tungsten, the elongation of the grains transversal to the surface is the result of vertically upwards oriented grain growth during the solidification process [169]. In terms of strength under forces acting longitudinally to the surface, the longitudinal grain structure is more advantageous, but the transversal grain structure, as found in IGP tungsten and LMD-w tungsten, provides the advantage of limiting the propagation of cracks running longitudinally to the surface. To minimize the risk of vacuum vessel contamination due to tungsten breakout, the transversal grain structure is preferred for PFM. Since, impurities of high Z materials such as tungsten can significantly reduce the the plasma temperature and thus impair or prevent the desired fusion reactor operation [10, 118, 119]. However, it should be mentioned that due to the high temperatures during fusion relevant HHF exposure, the surface-near areas (up to depths of ~ 2 mm) of whatever initial grain structure partly become recrystallized [15, 290, 293, 298]. This significantly reduces the relevance of the initial grain structure of PFM. Future studies should investigate whether the grain struc-

ture of solidified LMD-w tungsten also changes over time under thermal loading.

Overall, for the specific experiments discussed here, it can be concluded that the LMD-w coatings withstood combined steady state and transient thermal loading similarly well as uncoated IGP tungsten. And the cracks caused by thermal shocks had no effect on the performance of the test components under HHFs.

Test component design

Compared to monoblock chains, the test component design (see fig. 6.1 B)) offers the benefit to test LMD-w coatings under fusion-relevant thermal conditions more easily, in less time, and at lower cost. Three test components were therefore manufactured for the HHF experiments discussed here. The test components enable testing of LMD-w tungsten coatings under the same thermal conditions as those encountered on divertor monoblock components during operation of high-power fusion reactors. The thermal conditions involve the surface temperature and the temperature gradients in the material (fig. 6.2). In addition, the dimensions of the coated surface are similar to those of monoblocks (fig. 6.1). Due to slightly lower cooling capacity, the same thermal conditions are achieved on the test components at 1 MW m^{-2} less intensity as on monoblock components that are loaded with 15 MW m^{-2} . The surface temperatures measured during loading also correspond well with the temperatures calculated in advance via FEM.

The thermal loading experiments with these test components have revealed three aspects of their design that can be optimized for future thermal loading experiments of LMD-w coatings. With the test component design used in these experiments, at most the influence of thermal loads of 15 MW m^{-2} on the LMD-w coatings can be investigated, since this is the upper limit of the possible heat dissipation for this component (see fig. 6.2 and fig. 6.3). Swirl tape elements inside the cooling channels could be used to increase turbulences in the cooling water and thus increase the Reynolds number Re (see equation 4.19) and the convective heat transfer coefficient h_{cv} (see equation 4.17). These swirl tape elements are also included in ITER PFCs and visible in fig. 2.14 [144, 299].

Furthermore, the cooling water pressure, velocity and temperature could be increased to enhance heat removal at the inner surface of the cooling channel. To increase these parameters and to ensure that the clamping connections through which the components are connected to the cooling system of the electron beam facility are leak-proof, the components should be designed with a greater distance between the loaded surface and the clamping connection. Thermal expansion and shrinking of the components under cyclic exposure with steady state loads has in some cases led to slight leaks at the clamp connection. These leaks were counteracted by tightening the clamp connection with greater force, but the mentioned optimization of the

component design is necessary for experiments with higher cooling water pressures. Another option to enable the test components for the discussed enhanced coolant parameters and ensure leak tightness would be the application of swagelok instead of clamping joints.

The last aspect of the test component design worthy of improvement is that during loading of the tungsten double tiles, the copper next to the double tiles (in top view perspective) is also partially exposed to the electron beam. This can cause stress at the joint between the tungsten and copper at the edges of the tile. This could have influenced the local delamination of the brazing joint visible in fig. 6.14 C). The aforementioned copper edge could be removed by milling prior to the HHF experiments to eliminate the issue. With these improvements in test component design, the LMD-w layers could be exposed to conditions equivalent to steady state 20 MW m^{-2} loading on monoblock chains. Although 10 MW m^{-2} is the expected steady state heat flux in the divertor of high-power fusion reactors such as ITER, the aforementioned 20 MW m^{-2} can act in the divertor during so-called slow transients (10 s with a limited cycle count of 300 cycles) [144, 300].

6.6 Conclusion and outlook

The resilience of LMD-w coated PFCs was tested under fusion-relevant thermal loads in the electron beam facility JUDITH 2. For this purpose, actively cooled test components were designed, manufactured and coated with different tungsten layers via LMD-w. This way, the repair and compensation of erosion losses on divertor PFCs by deposition of LMD-w coatings was simulated.

Based on the results of the LMD-w deposition of tungsten on the plasma-facing surfaces of W-armored PFCs, the following conclusions were drawn. It was successfully demonstrated that the deposition is sufficiently precise to maintain the 0.5 mm gap, which is characteristic for the distance between monoblocks. Further, the coatings exhibit seamless bonds both at the connection to the substrate and between the individual stacked layers (see fig. 6.13). But, the deposited material exhibits defects in the form of thermally induced cracks that occurred during deposition, as well as pores caused by process instabilities, caused by lateral oscillations of the wire while entering the melt pool (fig. 3.8) and insufficient power of the laser source. The origin of these defects and how they could be avoided is discussed in detail in chapter 5.

The tungsten coatings were subjected to various thermal loading scenarios to test their resilience against multiple fusion reactor plasma discharges. The exposure scenarios include steady state thermal loading with up to 1000 cycles of 10 MW m^{-2} and 200 cycles of 15 MW m^{-2} loading, as well as ELMs (see fig. 6.5). The influence of these HHF scenarios on the test components and their LMD-w tungsten coatings

was comprehensively observed, analyzed and discussed.

All of the various LMD-w tungsten coatings withstood the steady state HHF scenarios without spalling or delamination. No significant variations between individual, stacked and smoothed tungsten layers in terms of their resilience to cyclic steady state thermal loads was observed. Compared to alternative processes for tungsten deposition, such as sputtering or plasma spraying, stacking the LMD-w layers allows deposition of several millimeter thick coatings [169].

Residual tensions arise during the cyclic thermal loading, due to thermally induced expansion and shrinkage of the material. These tensions can result in the formation of large cracks running parallel to the cooling channel and transversely to the thermally loaded surface (see fig. 6.12 A) uncoated B) coated). According to the literature, this effect was also observed for monoblock components [288–293]. Since these cracks occurred both in the uncoated and the coated tungsten double tiles, there seems to be no direct relation between the LMD-w tungsten coatings and the occurrence of these large cracks at cyclic steady state thermal loading.

The measured relative temperatures on most of the tungsten surfaces subjected to 15 MW m^{-2} thermal loading rose with increasing cycle count, as shown in fig. 6.10. The cooling constants provided in fig. 6.16 also increased with rising cycle counts, indicating deterioration in cooling performance. Since this behavior can also be observed with uncoated tiles, the LMD-w tungsten coatings are evidently not the origin of this deterioration. Even after the cyclic steady state HHF loading scenarios, the bonds between the tungsten layers and the tungsten substrate are intact and seamless, which also suggests that the degradation of cooling performance and increase in surface temperature are not related to the coatings. Instead, this behavior can most likely be attributed to partial delamination at the W/Cu brazing joint of the test components, as depicted in fig. 6.14 C), which most likely progressed with rising cycle count. The discussed changes to the test component design would prevent or even completely avoid the occurrence of such defects in future HHF studies of this kind.

The surface changes resulting from plastic deformation, which were determined by laser profilometry before and after the thermal loading scenarios are listed as ΔsR_a and ΔsR_q in table 6.4. Even though the increase in roughness is higher for the coated surfaces than for the uncoated ones, all roughness increase values determined for the cyclic steady state exposure scenarios are in a low and acceptable range.

The single layer LMD-w tungsten coatings also withstood the combination of steady state and transient thermal loads without spalling or delamination (red framed tungsten surfaces in fig. 6.11). No increase in surface temperature was observed with

growing count of thermal shock pulses, indicating that the bond between the coatings and the substrate remained unaffected. When exposed to the same thermal loading scenarios, IGP tungsten and the LMD-w tungsten coatings show similar damage patterns (see fig. 6.15) [80, 81, 109, 297]. Thus, at thermal loading experienced by the PFM in the divertor region of large scale fusion reactors during giant ELMs, the LMD-w tungsten behaves comparable to IGP tungsten.

The HHF trials performed here are initial experiments to obtain an assessment concerning the resilience of LMD-w coated PFCs under fusion relevant thermal loading scenarios. Future HHF studies of this type should be performed with IGP tungsten to prevent cracking. Furthermore, adjustments to the test component design should be implemented, which should help assess the capability of the coating process and clarify the influence of the brazed joints. Considering that the PFCs in the divertor of future FPPs will mainly be exposed to steady state heat loads of $\sim 10 \text{ MW m}^{-2}$, the obtained results are already promising. The initial surface condition influences the formation of thermally induced cracks, their growth, and detectability. Consequently, for more conclusive results and easier comparability of coated and uncoated surfaces, these should be polished to a mirror finish before HHF loading. In addition, the improvements to the LMD-w tungsten process discussed in chapter 5 should be implemented to achieve coatings with fewer defects.

Overall, the LMD-w tungsten coatings behaved similarly to uncoated tungsten at fusion-relevant steady state as well as combined steady state and transient thermal loading. These results are promising for the in situ regeneration of tungsten armored PFCs using LMD-w.

7 Repair of heat load damaged PFM using LMD-w

7.1 Test component design and manufacturing

During the occurrence of the type-I ELMs in large scale fusion reactors, which are characterized in section 2.3, exceptionally high transient heat loads act on the armor of the PFCs. These loads can damage the tungsten used as PFM. Depending on the intensity and number of ELMs, surface changes occur in the form of roughening, crack formation and recrystallization [114, 301]. For the in situ repair of PFCs, it follows that the tungsten layers for the compensation of eroded tungsten must be applied onto this pre-damaged substrate. The question arises whether the melt pool formed during the LMD-w process extends deep enough into the substrate material to melt away pre-existing damage, such as cracks, and thus heal it. Alternatively, prior to material deposition, the surface of the damaged PFM could be melted using only the laser to heal the defects in this way. This chapter addresses the potential of healing pre-existing defects using LMD-w or laser remelting. For this purpose, tungsten tiles were pre-damaged with combined steady state and transient heat fluxes and then subjected to various LMD-w and laser remelting processing treatments. Subsequently, these samples were examined using several analytical methods and it was determined whether the pre-existing defects in the substrate had been removed.

IGP tungsten was used as the material for the tungsten tiles. To cause ELM-like damage to these tungsten tiles, the established circular beam loading method in the electron beam facility JUDITH 2, which is also described in section 6.4, was used. To expose these tungsten tiles to controlled high thermal loads in the electron beam facility, they must be actively cooled. Therefore, these six tungsten tiles, are brazed to a copper heat sink, which is specially designed for this type of HHF experiments, as schematically visualized in fig. 7.1 [80]. To achieve similar surface structures on all tiles to ensure an equal initial state, the exposed surfaces of the tungsten tiles were grinded and polished before brazing. AgCu28Ge2Co0.3 brazing alloy foil of 0.1 mm thickness, which is only used for the purpose of mock-up fabrication and not suitable for real divertor components, is used. The brazing process is conducted as specified in section 6.2. The copper heat sink features a channel for active cooling

with water. For HHF experiments this cooling channel is connected to the sample cooling system of the electron beam facility by means of clamping elements, which ensures the necessary heat dissipation. To be able to apply thermal loading to the tungsten tiles while keeping them reasonably thermally isolated from each other, 3 mm deep and 1 mm wide gaps are made at defined distances in the copper cooling structure.

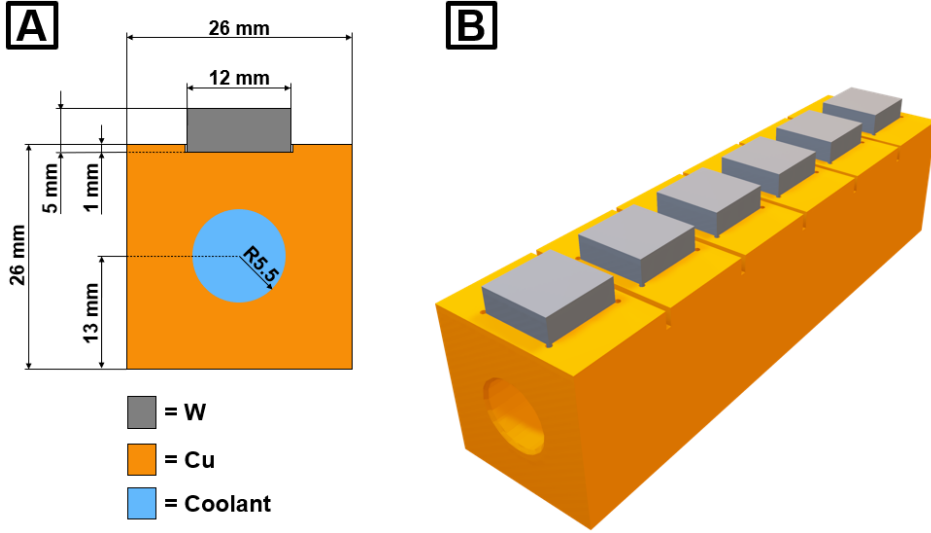


Figure 7.1: A) visualizes the cross-section of the test component on a plane perpendicular to the cooling channel axis, including its essential dimensions. Depicted in B) is the CAD rendering of the copper heat sink including six $12 \times 12 \times 5 \text{ mm}^3$ ITER grade tungsten tiles in their designated slots.

7.2 Damaging by transient thermal loads

Experimental

After successful manufacturing of the component, the six tungsten tiles are now subjected to the thermal loads that act on the PFM during ELMs. In the first step, the component is connected to the cooling system of the electron beam system JUDITH 2 using clamping elements as depicted in fig. 7.2. During the HHF experiments, 40 l min^{-1} of purified cooling water with an inlet temperature of 70°C and pressure of 20 bar is pumped through the cooling pipe of the component.

To simulate the influence of thermal loads experienced during ELMs on the PFM in the divertor region of large scale fusion reactors, the tungsten tiles of the test component are loaded with combined steady state and transient heat loads. The

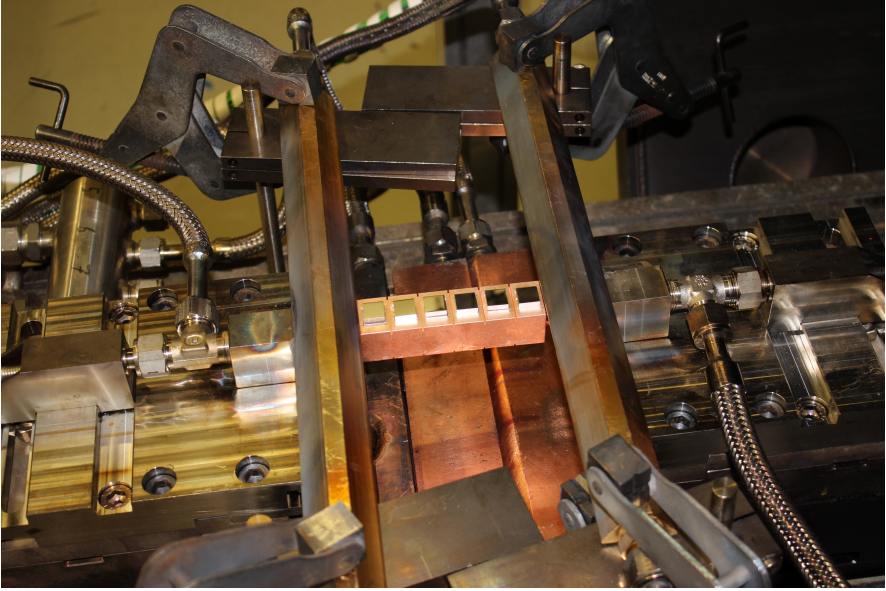


Figure 7.2: Shortly before the thermal loading, the test component is connected to the cooling system of the electron beam facility JUDITH 2. The actively cooled copper beam dumps are also visible. The continuously running electron beam is positioned on these beams dumps whenever it is not required on the tungsten samples [170].

six tungsten surfaces were subjected to 10^5 thermal shocks at a base temperature of $700\text{ }^\circ\text{C}$ with an absorbed intensity of $I = 0.55\text{ GW m}^{-2}$ ($F_{HF} = 12\text{ MW s}^{0.5}\text{ m}^{-2}$) using the circular beam loading method (as described in section 4.2). As specified in table 7.1, this results in a cyclic temperature change of 700 K to a maximum temperature of $1400\text{ }^\circ\text{C}$, calculated using equation 2.14. This value was validated during the HHF experiments using two-color pyrometer and high-speed single-color measurements. Here, the two-color pyrometer was used to calibrate the single-color pyrometer. The temperatures on the surface of the tungsten tile during HHF expo-

Table 7.1: The following loading parameters were used in the electron beam facility JUDITH 2 to simulate the damage caused by ELMs on the tungsten tiles [170].

Transient pulses	Pulse length Δt	Heat flux factor F_{HF}	Intensity I	Base temperature T_{base}	Peak temperature T_{max}
10^5	0.5 ms	$12\text{ MW s}^{0.5}\text{ m}^{-2}$	0.55 GW m^{-2}	$\sim 700\text{ }^\circ\text{C}$	$\sim 1400\text{ }^\circ\text{C}$

sure measured by the high-speed single-color pyrometer are plotted in Fig. 7.4. And fig. 7.3 depicts an IR camera image of a loaded tungsten tile during a transient pulse.

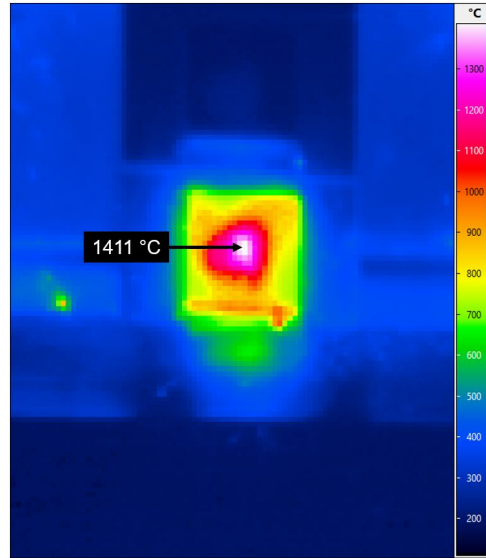


Figure 7.3: Depicted is an IR camera image of a tungsten tile subjected to combined steady state and transient heat loads as specified in table 7.1 [170].

Results

The damage to pure tungsten caused by combined steady state and transient thermal stresses has already been extensively researched [110,111,113,114,301,302]. As the thermal load increases, the damage patterns range from minor surface changes, crack formation, crack networks, crack networks with partial melting, up to severe melting of the surface. After thermal loading as described in table 7.1, the formation of a crack network with partial melting of the material is expected according to the literature [106,303,304]. This damage pattern is also evident on the tiles discussed here. Fig. 7.5 depicts SEM images of the damaged area on tile 6, which is also representative of tiles 1 to 5. In fig. 7.5 A) the damaged surface is displayed in a top view and B) displays a cross-sectional micrograph at the center of the damaged area. The network of cracks and the partial melting of the material in the form of small spheres are both visible in A). The depth of the cracks from the surface to the deepest point is visible in the cross-section shown in fig. 7.5 B) and reaches up to 450 μm . This maximum crack depth is consistent with values found in the literature. In the stress scenario described here, the literature assumes a mean crack depth of 250 μm [297,303]. Moreover, recrystallization is visible in the surface-near area down to the depth of 150 μm .

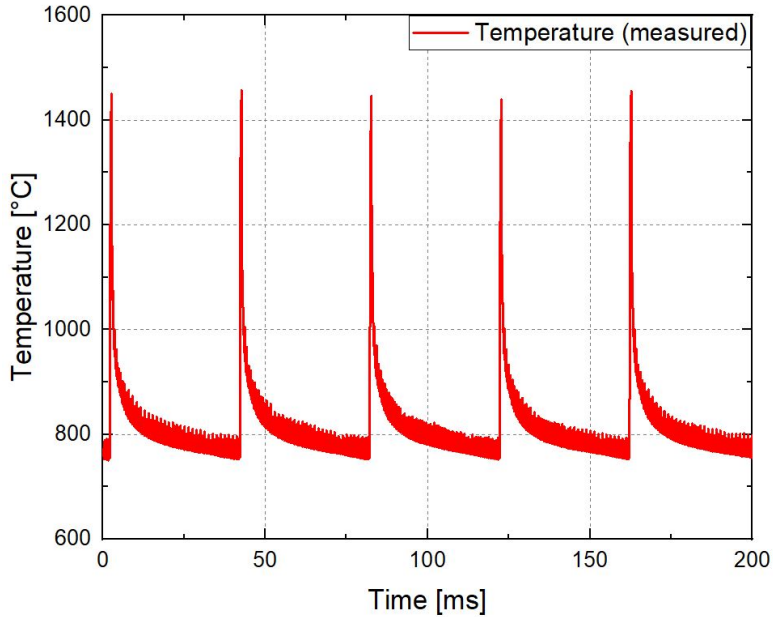


Figure 7.4: The high-speed single-color temperature measurement on the surface of a tungsten tile loaded as specified in table 7.1 is plotted here [170].

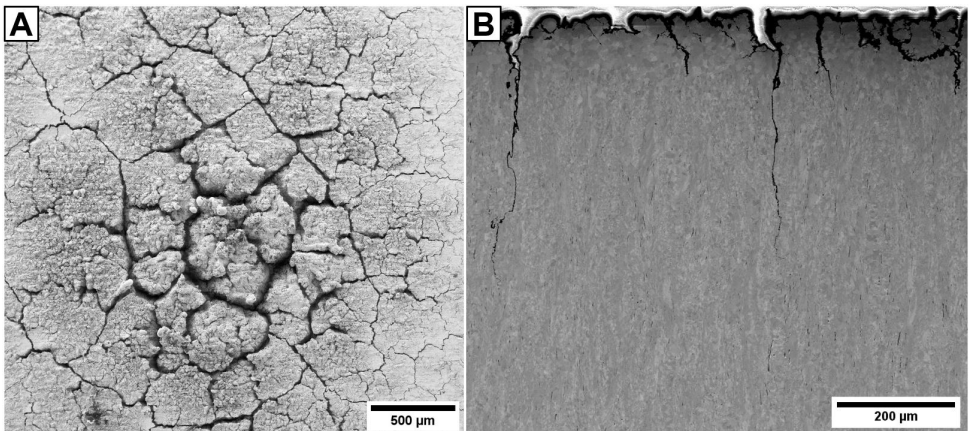


Figure 7.5: The area on tungsten tile number 6 that was exposed to combined steady state and transient thermal loads is depicted in the form of SEM images. A) displays the top view of the loaded surface, revealing a network of cracks and partial melting of the material in the form of small spheres. In B) the cross-section micrograph in the center of the loaded area reveals the depth of the cracks and the surface-near recrystallization [170].

7.3 Laser remelting FEM simulation

Experimental

As a result of the thermal shocks, the six tungsten tiles were significantly damaged, as visible in fig. 7.5. To determine the energy density required during the laser remelting process to melt deeply enough to heal these defects, thermal simulations of this process were developed. These thermal transient finite element method (FEM) simulations provide valuable insight regarding the temperature profile inside the tungsten during laser remelting. Of particular interest is the depth of the melt pool. The ANSYS 2020 R1 software was used for these FEM simulations. This involved the utilization of the Moving Heat Source V4.1 extension in the transient thermal module to appropriately simulate the process. As the name suggests, this extension allows movement of the heat source. Thereby, the distribution of thermal energy by conduction in three-dimensional space is calculated using equation 4.15. As substrate, a tungsten plate with the dimensions $60 \times 36 \times 6 \text{ mm}^3$ is used. The moving circular heat source positioned on the substrate represents the laser spot with a diameter of 2 mm. Exactly the same scenario was replicated experimentally with rolled L-oriented (grain structure elongated longitudinally to the surface plane) tungsten plate. Thus, in the evaluation, the calculated temperature inside the material can be related to the change in the grain structure by solidification of molten material and recrystallization in the HAZ. For the thermophysical properties of the substrate material, temperature-dependent data of pure tungsten is used in the simulation [275]. For calculation of the absorption and emission of tungsten, a coefficient of 0.4 is considered [106]. The emitted power by radiation on the upper side of the plate is considered and calculated according to equation 4.7. And for the bottom side of the plate, cooling by convection of 5000 W m^{-2} is assumed, to consider the energy dissipation on the actively cooled sample mount. With 5 kW of laser power and 400 mm min^{-1} scanning velocity, the same process parameters were employed for the comparison of the simulation and the experimental results. For reasons of complexity, the process is only thermally modeled without taking the fluid dynamics in the melt pool into account. Convergence analyses were performed to test the stability of the simulation and to verify the obtained results. For this purpose, the melt pool depth reached exactly at the midpoint of the scan distance was calculated and compared for different quantities of elements and time step durations. Calculations were performed with 73010, 264786 and 761402 elements with time step durations of 0.05 s, 0.01 s and 0.005 s. The values for melt pool depth and maximum surface temperature showed only small deviations of less than 0.5 %, which proves the network and time step duration convergence. To validate the temperature values calculated by the simulation, experimentally measured data is compared with the simulation data. This ensures that the thermal behavior of the real system is reflected with reasonable accuracy by the simulation. For this purpose, the temperatures resulting on the tungsten surface were measured during the scanning of one

melt track by means of an infrared (IR) camera (InfraTec VarioCAM® HDx). Since the surface temperatures reached during the process are above the melting point of tungsten ($T_{melt} = 3422\text{ °C}$) and thus also above the upper measurement limit of the IR camera, the temperatures in the cooling curves were compared. Fig. 7.6 plots the temperatures for the calculated cooling curve against the actual measured data for comparison. The sufficient agreement of the curves further validated the calculated temperature data and the assumed value for the convection cooling. In addition to the context of grain structure and process temperatures, the simulation was used to extrapolate beyond the process parameters achievable with the specified setup (see section 4.1). The resulting maximum surface temperatures and melt pool dimensions for laser powers above the currently available maximum of 5 kW were calculated.

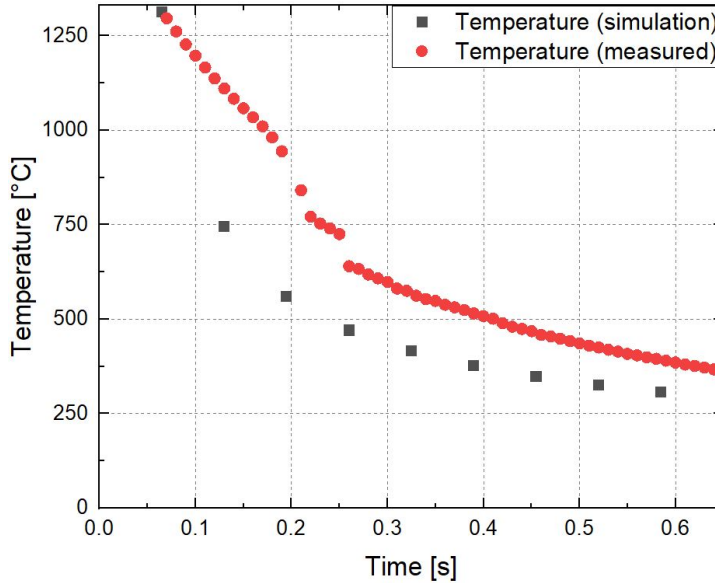


Figure 7.6: This plot compares the measured and simulated temperatures on a tungsten surface during the cooling phase directly after termination of the laser remelting process [170].

Results

For the tungsten plate on which the melt track was scanned, a cross-sectional micrograph was prepared. This micrograph was prepared on a plane perpendicular to the deposition direction. To visualize the grain structure, the BSE image of the micrograph was recorded. In fig. 7.7 A), this image is set alongside the temperature profile in the tungsten, which was generated using FEM simulation and is visualized in B). In A), an area with large grains, which is enclosed by a dashed line, is clearly

visible. In this area, the tungsten solidified from the molten phase. The resulting grains possess an aspect ratio of 1.9 and an average surface area of $4000 \mu\text{m}^2$. This zone extends, visible by the black dashed line, to a depth of $370 \mu\text{m}$. Below that lies the area of the heat-affected zone (HAZ), which extends up to the black dotted line and to the maximum depth of $1170 \mu\text{m}$. In this HAZ, the substrate material is recrystallized. As the distance from the melt pool decreases, the recrystallized grains exhibit increased sizes. According to simulation (see fig. 7.7 B), the temperatures in the HAZ were above 1800°C . Below the black dotted line, the material in A) still retains the original L-oriented grain structure of the rolled tungsten specimen. In fig. 7.7 B), the dashed line marks the area in which the temperature of 3422°C , i.e. the melting point of tungsten, was exceeded. The simulation predicts that the maximum melt pool depth is $383 \mu\text{m}$, which matches the experimentally determined value of $370 \mu\text{m}$. The calculated melt pool depths, widths and maximum surface

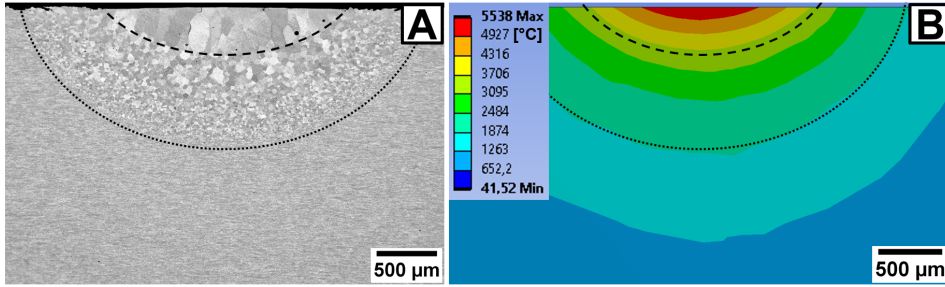


Figure 7.7: A) displays the BSE cross-section image of a pure tungsten sample subjected to the laser remelting process. One melt track perpendicular to the image plane with an energy density of 477 J mm^{-2} was used [170].

temperatures for various parameter sets of the laser remelting process are specified in table 7.2. Here, the laser power and scanning velocity were selected to yield the same energy density of 477 J mm^{-2} for all parameter sets. This energy density showed the best results in previously conducted studies [169]. The laser power was increased beyond the real limit of the specific system of 5 kW to 7.5 kW and 10 kW. It was found that with increasing laser power and scanning speed, the melt pool width and depth are increased, provided that the energy density remains constant. Thereby, the melt pool depth increases proportionally more than the width, leading to a decreasing aspect ratio of the melt pool width and depth.

Table 7.2: The maximum surface temperatures, melt pool widths and depths calculated through thermal transient SEM simulations of the laser remelting process are listed here. In all three scenarios, the laser power and scanning velocity were selected accordingly so that the energy density is 477 J mm^{-2} . Increased scanning velocity and laser power result in increased melt pool depth and width, with the depth increasing proportionally more than the width. The aspect ratio values serve to elucidate this effect [170].

Laser power [kW]	Scanning velocity [mm min ⁻¹]	Surface temperature [°C]	Melt pool depth [μm]	Melt pool width [μm]	Aspect ratio [width/depth]
5	400	5487	383	1866	4.9
7.5	600	7556	729	2664	3.7
10	800	9208	923	3066	3.3

7.4 LMD-w coating and laser remelting

Experimental

After exposure of the six tungsten tiles to combined steady state and transient thermal loads, as specified in table 7.1, five of the tiles were subjected to various LMD-w and laser remelting scanning strategies. Here, the intention was to investigate whether this process is capable of melting deeply enough to heal the pre-existing damage, and whether laser remelting prior to the LMD-w is necessary to do so. The scanning strategies were performed according to table 7.3. Thereby, the LMD-w setup as described in section 4.1 and 5.1 was utilized. For the depositions with 477 J mm^{-2} , the process parameters according to table 6.3 were used. For the laser remelting process, almost the same process parameters were adopted, with the exception of the wire feeding, which was deactivated for this case.

Table 7.3: The thermally pre-damaged tungsten tiles were subjected to different laser remelting and LMD-w scan strategies. The energy densities calculated according to equation 3.10 provide comparative values for the thermal energies applied during the respective processes [170].

Sample	Scan strategy
1	Layer (477 J mm^{-2})
2	Layer (637 J mm^{-2})
3	Remelting (477 J mm^{-2})
4	Remelting (477 J mm^{-2}) + Layer (477 J mm^{-2})
5	Remelting (477 J mm^{-2}) + Layer (637 J mm^{-2})
6	-

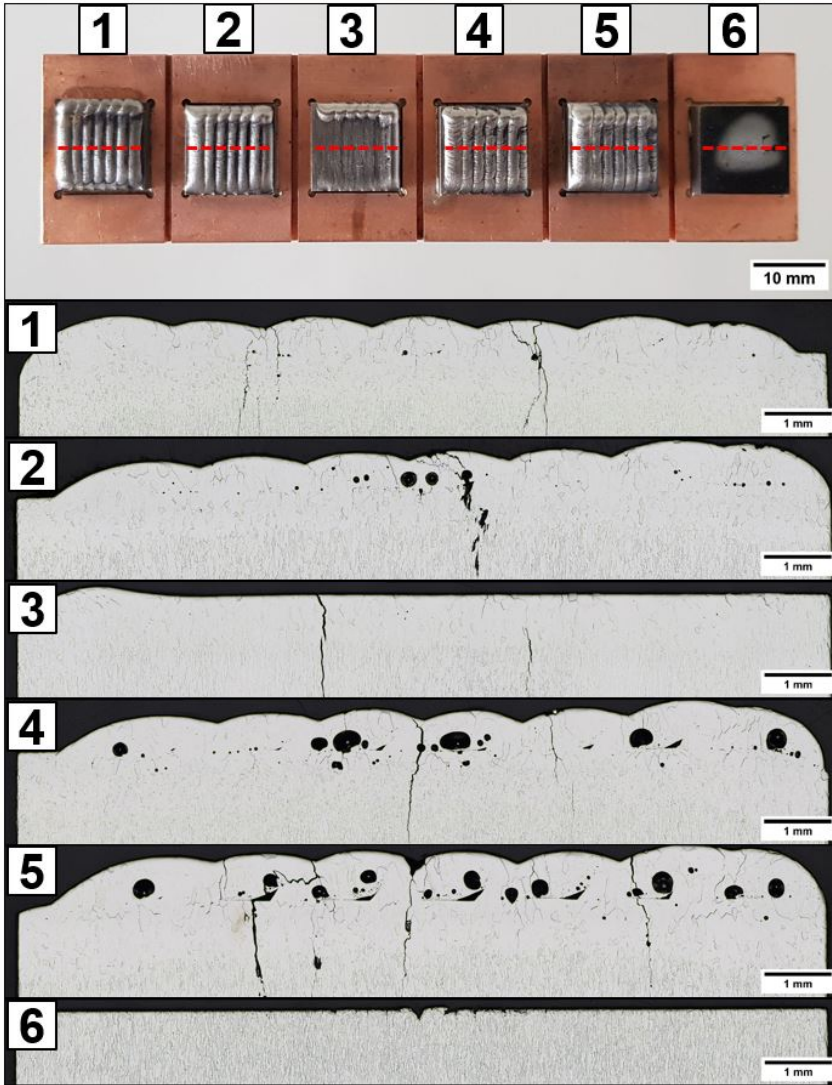


Figure 7.8: Depicted here are the 6 tungsten tiles that were exposed to the thermal loads specified in table 7.1 and afterwards treated with procedures noted in table 7.3, with tile 6 showing the state after the HHF exposure. Cross-section micrographs of these tiles taken by light microscopy are displayed [170].

A second applied energy density to achieve deeper melting was tried. For this, the scanning velocity was reduced from 400 mm min^{-1} to 300 mm min^{-1} , since 5 kW is the maximum laser power of the utilized laser source and optical system.

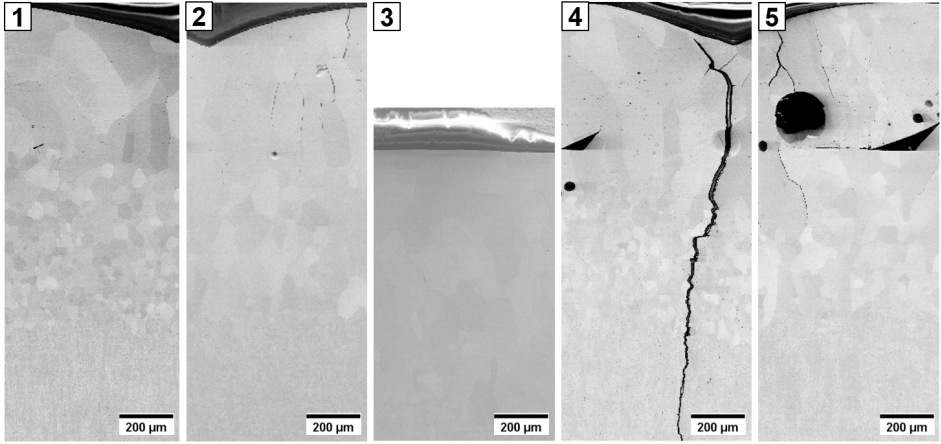


Figure 7.9: In addition to fig. 7.8, SEM images of the cross-section micrographs of samples 1 to 5 in the pre-damaged area are displayed. The damage caused by the HHF exposure depicted in fig. 7.5 has been eliminated, although, as discussed in chapter 5, macro cracks did develop [170].

Fig. 7.8 (top) depicts the state of the samples from the top view after coating and laser remelting. Sample 6 remained unaltered and serves as the reference sample, representing the state of the samples after HHF exposure.

Results

Cross-sectional micrographs were prepared of the tiles after treatment as described in table 7.3. These micrographs were recorded using light microscopy. Red lines on the overview image of the samples in fig. 7.8 indicate where specimens were cut. All samples were prepared with the cross-section at the location of the thermal loading. Tile 6 represents the state of all six samples directly after the HHF loading. Tiles 1 to 5, which were treated as described in table 7.3, exhibit macrocracks. These macrocracks are the result of steep temperature gradients in the material during the process, which lead to high residual stresses. Options for avoidance of these cracks are discussed in section 5.3. However, the damage caused by the HHF exposure, as described in section 7.2, is no longer evident in the coated and remelted samples on fig. 7.8 and fig. 7.9. The layers deposited using LMD-w measure up to $\sim 650 \mu\text{m}$ in height.

In particular, tile 1 (477 J mm^{-2}) displays excellent bonding with the substrate, such that the transition between substrate and deposited material is virtually unrecognizable. Furthermore, tile 1 displays hardly any other defects. The image analysis program ImageJ was used to quantify the porosity of the deposited material in the form of area density. With an area density of 99 % and thus porosity of

1 %, tile 1 shows the best result.

The LMD-w layer on tile 2 was coated with an increased energy density of 637 J mm^{-2} . The area density in the deposited material on tile 2 is lower at 97 %. This is mainly due to the spherical pores with diameters of up to $230 \text{ }\mu\text{m}$.

No material was applied to sample number 3. Instead, the surface was treated by laser with an energy density of 477 J mm^{-2} . Similar to the samples coated with LMD-w W, this laser remelting process repaired the damage caused by the HHF exposure.

On tile 4, the surface was first treated with the laser remelting process and afterwards coated with one layer using LMD-w (477 J mm^{-2}). In the cross-section micrograph of tile 4, large round pores are visible with maximum diameters of $400 \text{ }\mu\text{m}$. Therefore, the measured area density of 89 % is also lower than for tiles 1 and 2.

Tile 5 was also surface treated by laser remelting and then coated using LMD-w. However, the energy density used here was the highest at 637 J mm^{-2} . With an area density of 88 %, the deposited material here exhibits the lowest area density, which can be attributed to the high defect density. These defects include gas inclusions in between the welding beads and the large number of round pores with maximum diameters of $400 \text{ }\mu\text{m}$.

Electron backscatter diffraction (EBSD) measurements were performed on the cross-sections of the tiles to gain insights into the crystal orientation and grain structure. Fig. 7.10 shows the inverse pole figure (IPF) maps visualized in color code in z direction, which is the direction perpendicular to the surface plane. The results indicate that none of the samples have a preferred grain orientation. It was found that neither the samples coated with LMD-w nor the laser remelted samples or reference samples exhibit predominant grain orientations. All EBSD maps in fig. 7.10 are aligned with the substrate surface plane on the horizontal plane. The upper $\sim 600 \text{ }\mu\text{m}$ of material on tile 1 is tungsten deposited by LMD-w. This material was applied as melt. During solidification, the melt is primarily cooled by the substrate below, with the grains at the solid part of the substrate acting as nucleation sites. Further up, the temperature of the melt is still higher, which is why nucleation is not initiated there. The grains grow epitaxially from the substrate in the direction of the temperature gradients and thus vertically upwards, perpendicular to the surface. The grains in this area of solidified material are vertically elongated with an aspect ratio of 2 and exhibit an average measured cross-sectional area of $4000 \text{ }\mu\text{m}^2$. Below this area is the heat-affected zone (HAZ) of the substrate. In this zone, the material was recrystallized. The HAZ in tile 1 reaches down to the depth of $700 \text{ }\mu\text{m}$. Here the

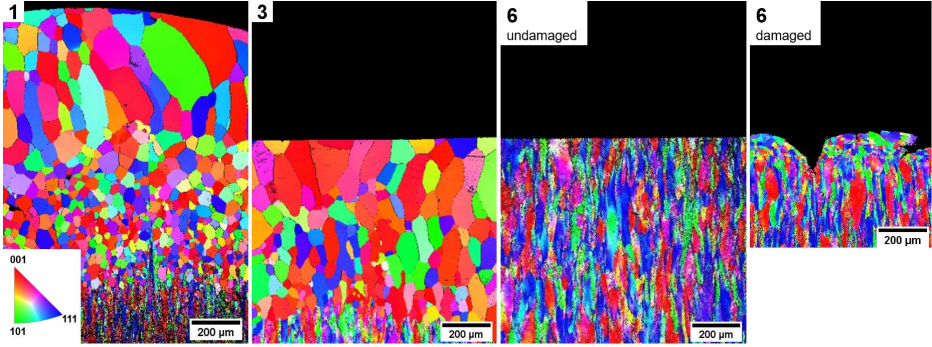


Figure 7.10: Visualized here are the inverse-pole-figure (IPF) maps of samples 1, 3 and 6, which were arranged with the surface plane of the substrate material aligned on the same level. Tile 1 was coated with the standard (477 J mm^{-2}) LMD-w layer and tile 3 was treated with the standard (477 J mm^{-2}) laser remelting process. ‘6 undamaged’ represents the state of the substrate material before exposure and ‘6 damaged’ represents the state of this material after the thermal exposure as specified in table 7.1 [170].

grains no longer extend vertically and instead feature a lower aspect ratio of 1. The size of the grains decreases vertically until the smaller grain size of the ITER Grade Plansee (IGP) tungsten material is reached. IGP tungsten is multi-forged and characterized by vertically elongated grains with an aspect ratio of 3.2 and measured grain area of $550 \text{ } \mu\text{m}^2$. Tile ‘6 undamaged’ displays the original grain structure of the IGP tungsten material. ‘6 damaged’, on the other hand, shows the condition of the IGP tungsten material after the exposure with ELM like combined steady state and transient heat loads. Here, strong surface modifications and surface-near recrystallization down to the depth of $150 \text{ } \mu\text{m}$ are visible. No material was deposited on tile 3, but only the surface was treated with the laser remelting process. It is clearly visible that the surface modifications after HHF exposure (visible on tile ‘6 damaged’) have been repaired and that the surface is smooth again. For specimen 3, the HAZ extends down to the depth of $800 \text{ } \mu\text{m}$. The upper area of this zone was melted and the lower area recrystallized. Here, most of the grains feature vertical elongation with an average aspect ratio of 1.9 and mean surface area of $4000 \text{ } \mu\text{m}^2$. Accordingly, the grain structure generated by the laser remelting process is comparable to that in material deposited by LMD-w.

Hardness measurements were performed on the cross-sections of samples 1 to 6 to gain insight into the mechanical properties. Fig. 7.11 provides an overview of the measurement results, including their positions on the vertical axis of the cross-sections. On the x-axis of the diagram in fig. 7.11, the measured hardness Vickers (HV1) is plotted. Here, 0 on the y-axis represents the surface plane of the

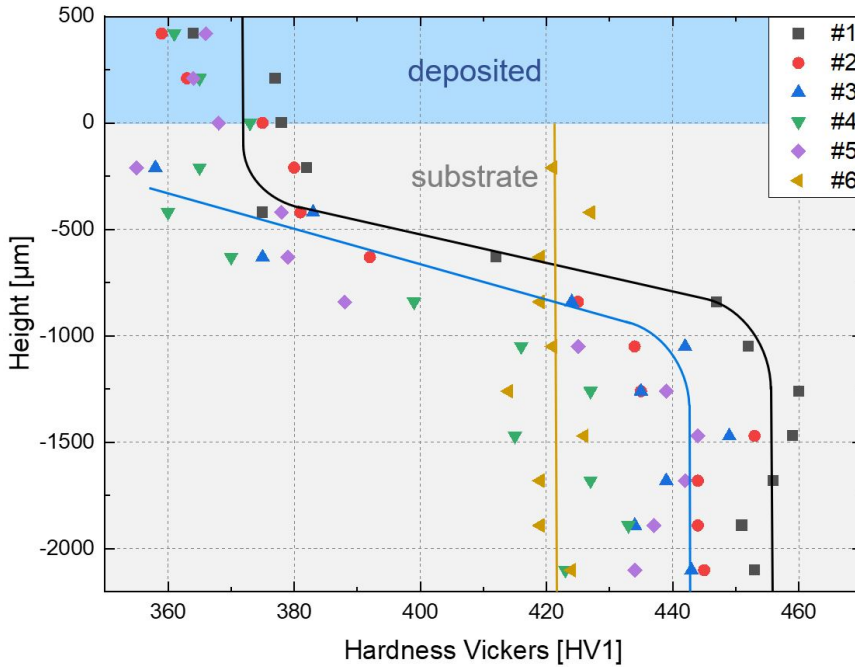


Figure 7.11: The results of the HV1 Vickers hardness measurements performed on the cross sections of samples 1 to 6 in the pre-damaged area are plotted. Positive height values indicate the tungsten deposited by LMD-w, while negative height values refer to the harder IGP tungsten substrate material. For more clarity, guide the eye lines have been added for the samples 1, 3, and 6 [170].

tungsten tiles. Given height values above 0 μm are located in the deposited material and below 0 μm are located in the substrate material. Each value displayed in fig. 7.11 is the arithmetic mean of five measured values. To avoid any mutual influence of the individual measured values, these are located at least 210 μm apart from one another, which corresponds to three times the diagonal length of the hardness indents. As visible in fig. 7.11, the yielded hardness values in tile 6 are close together across all depths. The arithmetic mean value of 420 HV1 indicates that the initial hardness of IGP tungsten of ~ 440 HV1 remained almost unchanged. The deposited material (height > 0 μm) exhibits lower average hardness values compared to IGP tungsten. The mean hardness values in this area for the samples 1, 2, 4 and 5 are ~ 370 HV1. The hardness increases with depth in the HAZs, which, depending on the sample, range from 0 μm to -700 μm or -900 μm , until the hardness of the IGP tungsten is reached at ~ 440 HV1. This dependence of hardness on depth is analogous to the change in the microstructure, which is particularly visible in

fig. 7.10.

7.5 Discussion

This chapter primarily addresses the question of whether the PFM is melted deeply enough during the deposition of tungsten by LMD-w and during laser remelting to heal damage caused by fusion-relevant HHF loads. And it was investigated whether it is necessary to perform a laser remelting step prior to the deposition to achieve this. To meet the objective of damage removal, the depth of the melt pool must be sufficient to reach surface modifications and cracks that occur in the PFM as result of the loading during the operation of the fusion reactor. For the purpose of healing thermally caused cracks, it has already been demonstrated that tungsten can be melted to depths of 300 μm by exposing it to laser irradiation at appropriate intensities [305]. Here, the tungsten tiles subjected to the combined steady state and transient thermal loads as specified in table 7.1 exhibited even deeper cracks of up to 450 μm depth (see fig. 7.5).

Melt pool depth calculation

To gain insight into the dimensions of the melt pool and, in particular, its depth, transient thermal FEM simulations of tungsten melting induced by laser irradiation were performed. Adequate agreement between the experimentally measured surface temperatures during the cooling phase of the laser remelting process and the calculated temperatures validated the temperature data obtained from the FEM simulations. Accordingly, the discussed FEM simulation can be utilized as a suitable thermal approximation for the calculation of temperatures inside the tungsten substrate material when subjected to laser irradiation. In fig. 7.7, A) the grain structure observed in the BSE micrograph of laser-remelted tungsten (using an energy density of 477 J mm^{-2}) was compared with B) the calculated temperature profile during the melting process for this case. According to the temperature profile of fig. 7.7 B), the melting temperature of tungsten is exceeded in an area down to a depth of 383 μm . In this area (up to a depth of 370 μm), large grains with an average cross-sectional surface area of 4000 μm^2 , which is typical of material solidified from the melt, can be observed on the micrograph in fig. 7.7 B). These results suggest that using energy densities of 477 J mm^{-2} should cause the majority of cracks to be melted and disappear [297]. Furthermore, the simulation was used to extrapolate beyond the laser powers that are experimentally feasible with the specified setup. The results in table 7.2 indicate that, at constant energy density, the melt pool width and depth increase with increasing laser power and scanning velocity. The effect of a steeper increase in melt depth compared to melt width has also been observed in the literature [306,307]. Thereby, the melt pool depth increases disproportionately more than the width. This is also reflected in the decreasing aspect ratio in table 7.2. At increased laser powers and scanning velocities, more energy

is deposited in less time while maintaining constant energy density. Since less time is available for the thermal energy to be distributed in the tungsten the melting temperature of tungsten is exceeded in a larger area.

Repair of damaged PFM by LMD-w coating and laser remelting

Damage caused by ELM-like thermal loads was effectively removed from the tungsten substrate of tile 1 to 5 by melting the damaged area using the LMD-w and laser remelting process (fig. 7.8 and fig. 7.9). Unfortunately, both laser processes caused the formation of new cracks, which are particularly visible in the cross-sectional micrographs. These are the macrocracks covered in chapter 5. During the laser process, steep temperature gradients occur between the heated material and the comparatively cooler substrate, which lead to residual stresses during the solidification of the melt. These residual stresses can be released in the material in the form of these macrocracks. The origin of these macrocracks and how they could be avoided is discussed in detail in chapter 5. The observed surface waviness perpendicular to the deposition direction of the LMD-w layers can be prevented by subsequent laser remelting (with the scanning direction parallel to the prior deposition direction), as demonstrated in chapter 5 (see fig. 5.4). In tiles 2, 4 and 5, large spherical pores occurred that have the appearance of keyhole defects. Such keyhole defects are signs of excessive energy input. In tiles 4 and 5, gas inclusions between the melt tracks indicate poor wetting of the substrate. Before the LMD-w coating, the surfaces of tiles 4 and 5 were treated by the laser remelting process. Apparently, this procedure had no positive effect on the quality of the coating. Apparently, more energy was absorbed by the deposited material in tile 4 compared to sample 1, which could explain the formation of these keyhole defects. This could also explain the increased defect density when comparing tile 2 and 5. This suggests that lower energy densities may be advantageous when applying LMD-w coatings after treating surfaces by laser remelting. Further studies must be conducted to answer this question conclusively. The change in hardness (fig. 7.11) from the deposited material (~ 370 HV1), across the HAZ and down to the original substrate material (~ 440 HV1) progresses analogously to the change in the grain structure (see fig. 7.10) and can also be attributed to the latter.

7.6 Conclusion and outlook

After certain periods of fusion reactor operation, specific damage is expected on the plasma-facing armor of the inner wall. Studies were conducted to determine the capability of the LMD-w and laser remelting process, to heal thermally initiated damage in PFM resulting from ELMs. The question is, whether during these laser processes the substrate material is melted deeply enough to remove the damage. In addition, the aim is to test whether the LMD-w process alone is sufficient to remove the defects or whether a laser remelting step must be performed first. It was also

investigated whether the deposited tungsten bonds with the previously damaged substrate.

In this context, transient thermal FEM simulations of the laser remelting process were developed, verified and validated. These simulations were utilized to calculate temperature profiles and melt pool dimensions in relation to specific process parameters. Fig. 7.7 A) displays the tungsten grain structure in form of a BSE cross-section micrograph of a melt track cut perpendicular to the scanning direction. Analogue to this, the temperature profile during the process, calculated using the FEM simulation, is visualized in color code in fig. 7.7 B). This investigation suggested, that with application of energy densities of 477 J mm^{-2} , the melt pool depth of $\sim 370 \text{ }\mu\text{m}$ should reach deep enough to melt away the majority of the damage [297]. Furthermore, the FEM simulation was used to obtain information on the thermal behavior of tungsten at laser powers above the maximum power of 5 kW experimentally available with the utilized laser source. The results revealed that the melt pool depth and width increase with increasing laser power and scanning velocity for a constant energy density. The melt pool depth increases disproportionately more compared to the melt pool width. This effect is attributed to heat accumulation caused by the concentrated input of energy within a shorter period of time.

The question of whether the characteristic damage to PFCs caused by exposure in high-power fusion reactors can really be repaired by LMD-w and laser remelting was addressed experimentally. With this in mind, tungsten tiles were exposed to fusion-relevant combined steady state and transient thermal loads (as specified in table 7.1) in the electron beam facility JUDITH 2 to generate ELM-like damage. The results of this exposure include surface changes, partial melting of material, surface-near recrystallization to depths of $150 \text{ }\mu\text{m}$ and cracks with maximum depths of $450 \text{ }\mu\text{m}$ (see fig. 7.5). Afterwards, the tiles 1 to 5 were subjected to various LMD-w and laser remelting scanning strategies according to the plan specified in table 7.3. Both the LMD-w and the laser remelting process proved to be capable of healing the characteristic initial damage. This means that no laser remelting step is necessary prior to tungsten deposition to heal the damage. However, new defects were caused by these laser processes (see fig. 7.8 and fig. 7.9). The origin of the macrocracks, which occur with the current LMD-w setup during tungsten deposition, and how these defects could be avoided, is discussed in detail in the discussion section of chapter 5. The coating deposited with an energy density of 477 J mm^{-2} on tile 1 exhibits the best quality with an area density in the deposited material of 99 % measured on its cross-section. This coating displays flawless bonding to the substrate.

The HAZ and thus the change in grain structure in form of recrystallization extends down to depths of 700 to 900 μm . Analogous to the change in grain structure

(see fig. 7.10), material softening was also measured (fig. 7.11) from the hard original IGP tungsten substrate material (~ 440 HV1), through the HAZ to the new deposited tungsten (~ 370 HV1).

Overall, the studies covered in this chapter have demonstrated that healing of damaged PFM (tungsten) is possible using LMD-w and laser remelting. However, further process optimizations, as outlined in chapter 5, are necessary to avoid crack formation during the laser treatment.

8 Overall conclusion and outlook

8.1 Conclusion

The limited durability of plasma-facing components (PFCs) will represent one of the greatest challenges for the realization of economically viable fusion power plants (FPPs) [308]. Despite utilization of the most promising candidate for the plasma-facing material (PFM) in the form of tungsten, the service life of PFCs remains limited [9, 120, 309]. Since the replacement of these components for large-scale fusion reactors can occupy the time frame of years (if ever possible), this solution is insufficient [43]. Frequency and duration of in-vessel maintenance operations are the key factors influencing the availability and thus the profitability of FPPs [310]. One alternative approach to solve this issue would be the in situ repair of damaged and eroded PFM. By direct energy deposition (DED) of new armor material, simultaneous healing of surface damage and compensation for erosion losses could be achieved. This in-vessel repair operation would be performed by a deposition unit, acting as end-effector, mounted on a multipurpose deployer (MPD) (see fig. 2.19 for the current MPD design iteration of the European DEMO FPP). The wire-based laser metal deposition (LMD-w) process provides all the necessary characteristics for in-vessel repair of PFC armor. Hence, the focus of this work is research on the surface regeneration of the PFM (tungsten armor of the PFCs) by LMD-w. This subject was explored in detail from various perspectives to answer several key questions. These include whether deposition of tungsten on a tungsten substrate using LMD-w is even feasible. How the deposited tungsten coatings behave under fusion-relevant conditions and whether healing of pre-damaged PFM by LMD-w is achievable. In the course of this research, further related questions arose that were also addressed in this work.

With regard to the feasibility of the deposition of tungsten on tungsten substrate using LMD-w, comprehensive studies were conducted and presented in chapter 5 [169]. It was demonstrated that by arranging welding beads next to each other, layers of 99 % density and even stacked layers can be applied. By adaptation of the LMD-w process specifically for tungsten, the majority of the defect types that occurred were successively addressed and largely eliminated. The origin of macrocracks and ways to avoid them have been detailed in the discussion section of chapter 5 and will

be elaborated in the following outlook section. In principle, excessively steep temperature gradients during the process between the melt and the comparatively cold substrate must be flattened to reduce the formation of high residual stresses, which arise during the solidification process. In the cross-section micrographs, the deposited tungsten, which solidified from the melt, is characterized by comparatively large grains ($\sim 4000 \mu\text{m}^2$). These vertically elongated tungsten grains (average aspect ratio of 1.9) grow epitaxially upwards from the substrate during solidification of the melt. The wavy topology of the layers perpendicular to the deposition direction, which is the result of the adjacent arrangement of welding beads, was examined in the plasma-wall interaction context using the ERO2.0 code. It was found that the total heat flux, erosion, and deposition are not influenced by surface waviness or plasma incidence direction ϕ . But heat flux and erosion at wavy topologies (for $\phi > 0^\circ$) were concentrated in certain areas (as visible in fig. 5.8 and fig. 5.10), which led to higher local peak heat flux and erosion values. Especially in the case of wavy surface topologies, with the plasma incidence direction perpendicular to the LMD-w deposition direction ($\phi = 90^\circ$), the heat flux and erosion are concentrated. However, if the plasma incidence direction is chosen parallel to the deposition direction ($\phi = 0^\circ$), heat flux and erosion are more evenly distributed, and comparable to those of flat surfaces. For vertical targets of a tokamak divertor, it would therefore be advantageous to apply the coating with the deposition direction parallel to the long side of the monoblocks (toroidal axis of the reactor). But it was also demonstrated that the surface waviness can be significantly reduced by remelting the tungsten layers via laser. This particularly applies to the scanning direction of the laser remelting parallel to the deposition direction of the coating.

The resilience of LMD-w coated tungsten armored test components when exposed to fusion-relevant thermal loads was tested in the electron beam facility JUDITH 2 (chapter 6). Tungsten tiles coated using various scan strategies were cyclically exposed to conditions equivalent to the steady state thermal loads of 10 and 15 MW m^{-2} that are expected in the divertor area of high-power fusion reactors. Each cycle, which consists of a loading and cooling phase, simulates one plasma discharge in the fusion reactor. Prior to the experiments, thermal FEM simulations were carried out to determine the correct loading parameters and the upper threshold of heat dissipation for the test component. At 10 MW m^{-2} , the LMD-w coatings exhibited only the macrocracks already present from the LMD-w deposition. With increasing number of cycles, these cracks grew particularly centrally on the tiles parallel to the cooling channel. At 15 MW m^{-2} , the tiles coated with one and two layers behaved exactly like the uncoated references and developed large cracks centrally on the tile, running parallel to the cooling tube of the component. Additionally, LMD-w coated tungsten tiles were subjected to combined steady state and transient thermal loads (up to 10^5 thermal shocks at 700 °C base temperature; $I = 0.55 \text{ GW m}^{-2}$; $F_{HF} = 12 \text{ MW s}^{0.5} \text{ m}^{-2}$) to simulate the effect of edge localized modes (ELMs) on

these layers. Depending on the loading scenario, crack networks formed, along with partial melting of the material, which was recognizable afterwards as small spheres on the surface. Overall, both in the steady state as well as in the combined steady state and transient loading scenarios, the LMD-w coated tiles behaved comparable to uncoated references.

The studies detailed in chapter 7 were conducted to determine whether damage to the PFM that develops over a certain period of operation of the fusion reactor can be healed by LMD-w and laser remelting [170]. To achieve this healing, the melt pool during the laser treatment must extend deep enough to reach the defects present in the PFM. Thermal transient FEM simulations were performed to determine the energy density required to melt deep enough to heal the characteristic damage. At the energy density of 477 J mm^{-2} , the melt pool with depths of $\sim 380 \text{ }\mu\text{m}$ should extend deep enough to repair the majority of defects. Subsequently, this was experimentally validated. Therefore, tungsten tiles were brazed onto test components specially designed for this purpose and subjected to combined steady state and transient thermal loads in the electron beam facility JUDITH 2 to simulate ELMs. On these tiles, tungsten was then deposited by LMD-w, or the surface was treated by the laser remelting. Despite the formation of new cracks, whose avoidance is discussed in the outlook section, the process successfully demonstrated that healing of surface defects, previously inflicted to W tiles, is indeed possible.

8.2 Outlook

Based on these research findings, the deposition of resilient tungsten layers on tungsten armored PFCs using LMD-w can be evaluated as a potentially viable approach. However, for this approach to be considered promising, certain characteristics of the tungsten deposition still need to be improved and certain aspects of the deposited tungsten, need to be investigated in more detail in the context of nuclear fusion.

Cracks that develop during the deposition of tungsten and extend all the way into the underlying substrate are absolutely unacceptable for the PFCs of fusion reactors. These cracks are caused by steep temperature gradients during deposition and the subsequent residual stresses in the material. By flattening these temperature gradients, the formation of these cracks can be avoided. For powder bed fusion of tungsten, the probability of crack formation was significantly reduced by substrate heating to $600 \text{ }^{\circ}\text{C}$ (which raises the tungsten above its ductile to brittle transition temperature) and entirely avoided for $1000 \text{ }^{\circ}\text{C}$ [193, 201, 205, 282]. These cracks can be avoided for LMD-w by temperature homogenization in the substrate material in the form of wider laser spots with flatter flanks of the energy density profile or by utilization of a second laser spot for pre- and post-heating. It is conceivable to utilize coaxial wire feeding, whereby the laser spot that melts the material is surrounded

by a second laser ring projected for better temperature control. This concept would thus enable pre- and post-heating and promises enhanced quality for the welding beads. For coaxial systems of this kind, the welding quality could potentially be even further improved by preheating the wire with the laser. Furthermore, LMD-w of tungsten should be performed and analyzed in vacuum environment, since in most cases vacuum leads to significantly improved welding quality.

Results were published, that demonstrate the crack-free deposition of tungsten by electron beam additive manufacturing (EBAM) [283]. In these experiments, a comparatively wide effective beam diameter was achieved using circular beam figures. By utilization of a sufficiently compact and lightweight electron beam gun, which would allow in-vessel operation on the MPD even in the narrow divertor area, the repair of PFMs using EBAM would even be conceivable. But this would necessarily require maintaining vacuum during the maintenance operation. For remote maintenance (RM) operations the electronics of the electron beam gun would need to be shielded from radiation exposure, which would further increase the total weight of the end-effector.

In comparison to the replacement of damaged or eroded PFCs, which may take years, the LMD-w process enables repairs to be performed within days or weeks, which would significantly increase the availability and thus the profitability of future FPPs. If, for example, ~ 1 mm of material is eroded along the strike line of the highly loaded outer vertical target of the DEMO FPP divertor across the width of three monoblocks, compensation for these erosion losses by LMD-w coating would require ~ 95 hours of pure deposition time (deposition of 2 stacked layers on a total of 7128 monoblocks). Including the time required to position and maneuver the deposition unit, such repairs could be completed within the time frame of one or two weeks. To increase the deposition rate, several welding beads could be deposited simultaneously by parallel arrangement of multiple deposition units.

The interaction of the hydrogen isotopes used as fuel in fusion reactors with LMD-w tungsten remains to be investigated, although the expected results will likely be comparable to those for recrystallized tungsten. The effect of tungsten deposition on the surface of PFCs coated with boron, which is used in tungsten armored fusion reactors to reduce plasma impurities, also remains to be investigated [71, 311, 312].

Equipping the in-vessel deposition unit with sensors would also be advantageous (or even necessary) for the envisaged RM application inside the fusion reactor. In situ laser profilometry measurements of the coating surfaces could be utilized to monitor the process and detect the occurrence of irregularities that could be corrected by laser remelting. Also, the identification of potential defects (before and after LMD-w deposition) could be achieved by thermal monitoring. Through heat-

ing up large areas of PFCs using laser irradiation while simultaneously monitoring this area with an IR camera, irregularities in the dissipation of thermal energy to the heat sink could be identified.

To advance this surface regeneration concept to the next level, the development and construction of a customized LMD-w processing unit specifically designed for tungsten deposition inside a nuclear fusion reactor vacuum vessel would be desirable. To utilize LMD-w tungsten coatings as PFM in FPPs, further comprehensive material qualification studies are necessary.

Bibliography

- [1] EUROfusion; European research roadmap to the realisation of fusion energy (2018) (cited on page: 1)
- [2] M. Wiatros-Motyka, D. Jones et al.; Global Electricity Review 2023; *Ember* (2023) (cited on page: 1)
- [3] J. Ongena, G. Van Oost; ENERGY FOR FUTURE CENTURIES Prospects for fusion power as a future energy source; *Fusion Science and Technology* 61 (2t) 3–16 (2012); ISSN 1536-1055; doi: <https://doi.org/10.13182/Fst12-A13488> (cited on page: 1)
- [4] J. Mailloux, N. Abid et al.; Overview of JET results for optimising ITER operation; *Nuclear Fusion* 62 (4) (2022); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/ac47b4> (cited on pages: 1, 14, and 18)
- [5] R. A. Pitts, S. Carpentier et al.; A full tungsten divertor for ITER: Physics issues and design status; *Journal of Nuclear Materials* 438 S48–S56 (2013); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2013.01.008> (cited on pages: 1, 96, and 97)
- [6] A. J. H. Donné; The European roadmap towards fusion electricity; *Philosophical Transactions of the Royal Society a-Mathematical Physical and Engineering Sciences* 377 (2141) (2019); ISSN 1364-503x; doi: <https://doi.org/10.1098/rsta.2017.0432> (cited on page: 1)
- [7] G. Federici, C. H. Skinner et al.; Plasma-material interactions in current tokamaks and their implications for next step fusion reactors; *Nuclear Fusion* 41 (12r) 1967–2137 (2001); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/41/12/218> (cited on pages: 2 and 48)
- [8] R. A. Causey, J. N. Brooks et al.; Tritium inventory and recovery in next-step fusion devices; *Fusion Engineering and Design* 61-62 525–536 (2002); ISSN 0920-3796; doi: [https://doi.org/10.1016/S0920-3796\(02\)00248-X](https://doi.org/10.1016/S0920-3796(02)00248-X) (cited on pages: 2 and 69)
- [9] J. Linke, J. Du et al.; Challenges for plasma-facing components in nuclear fusion; *Matter and Radiation at Extremes* 4 (5) (2019); ISSN 2468-2047;

- doi: <https://doi.org/10.1063/1.5090100> (cited on pages: 2, 18, 21, 22, 23, 48, 69, and 143)
- [10] V. Philipps; Tungsten as material for plasma-facing components in fusion devices; *Journal of Nuclear Materials* 415 (1) S2–S9 (2011); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2011.01.110> (cited on pages: 2, 24, 49, and 120)
- [11] R. A. Pitts, X. Bonnin et al.; Physics basis for the first ITER tungsten divertor; *Nuclear Materials and Energy* 20 (2019); doi: <https://doi.org/10.1016/j.nme.2019.100696> (cited on pages: 2 and 32)
- [12] Y. Ueda, J. W. Coenen et al.; Research status and issues of tungsten plasma facing materials for ITER and beyond; *Fusion Engineering and Design* 89 (7-8) 901–906 (2014); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2014.02.078> (cited on page: 2)
- [13] M. Kaufmann, R. Neu; Tungsten as first wall material in fusion devices; *Fusion Engineering and Design* 82 (5-14) 521–527 (2007); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2007.03.045> (cited on page: 2)
- [14] G. Federici, L. Boccaccini et al.; An overview of the EU breeding blanket design strategy as an integral part of the DEMO design effort; *Fusion Engineering and Design* 141 30–42 (2019); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2019.01.141> (cited on pages: 2, 12, 15, and 30)
- [15] J. H. You, G. Mazzone et al.; Divertor of the European DEMO: Engineering and technologies for power exhaust; *Fusion Engineering and Design* 175 (2022); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2022.113010> (cited on pages: 2 and 120)
- [16] S. Brezinsek, J. W. Coenen et al.; Plasma-wall interaction studies within the EUROfusion consortium: progress on plasma-facing components development and qualification; *Nuclear Fusion* 57 (11) (2017); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/aa796e> (cited on page: 2)
- [17] T. R. Barrett, G. Ellwood et al.; Progress in the engineering design and assessment of the European DEMO first wall and divertor plasma facing components; *Fusion Engineering and Design* 109-111 917–924 (2016); ISSN 09203796; doi: <https://doi.org/10.1016/j.fusengdes.2016.01.052> (cited on pages: 2, 22, 29, and 69)
- [18] J. Palmer, M. Irving et al.; The design and development of divertor remote handling equipment for ITER; *Fusion Engineer-*

- ing and Design* 82 (15-24) 1977–1982 (2007); ISSN 09203796; doi: <https://doi.org/10.1016/j.fusengdes.2007.05.073> (cited on page: 2)
- [19] Y. Noguchi, M. Saito et al.; Design progress of ITER blanket remote handling system towards manufacturing; *Fusion Engineering and Design* 136 722–728 (2018); ISSN 09203796; doi: <https://doi.org/10.1016/j.fusengdes.2018.03.068> (cited on page: 2)
- [20] D. Naujoks, K. Asmussen et al.; Tungsten as target material in fusion devices; *Nuclear Fusion* 36 (6) 671–687 (1996); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/36/6/I01> (cited on page: 2)
- [21] S. Brezinsek, A. Kirschner et al.; Erosion, screening, and migration of tungsten in the JET divertor; *Nuclear Fusion* 59 (9) (2019); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/ab2aef> (cited on page: 2)
- [22] T. DebRoy, H. L. Wei et al.; Additive manufacturing of metallic components – Process, structure and properties; *Progress in Materials Science* 92 112–224 (2018); ISSN 00796425; doi: <https://doi.org/10.1016/j.pmatsci.2017.10.001> (cited on pages: 2 and 88)
- [23] M. Gipperich, C. Peters et al.; Reflectometry-based investigation of temperature fields during dual-beam Laser Metal Deposition; In *Laser 3D Manufacturing VIII*; doi: <https://doi.org/10.1117/12.2576112> (cited on pages: 2 and 50)
- [24] M. P. Fewell; The Atomic Nuclide with the Highest Mean Binding-Energy; *American Journal of Physics* 63 (7) 653–658 (1995); ISSN 0002-9505; doi: <https://doi.org/10.1119/1.17828> (cited on page: 7)
- [25] H. Yamada; Nuclear Fusion pp. 1901–1945; Springer International Publishing (2022); ISBN 978-3-030-72579-2; doi: https://doi.org/10.1007/978-3-030-72579-2_31 (cited on page: 8)
- [26] F. Ajzenbergselove; Energy-Levels of Light-Nuclei $a = 5-10$; *Nuclear Physics A* 490 (1) 1–225 (1988); ISSN 0375-9474; doi: [https://doi.org/10.1016/0375-9474\(88\)90124-8](https://doi.org/10.1016/0375-9474(88)90124-8) (cited on pages: 8 and 9)
- [27] A. Einstein; Is the inertia of a body dependent on its energy content?; *Annalen Der Physik* 18 (13) 639–641 (1905); ISSN 0003-3804 (cited on page: 9)
- [28] J. D. Lawson; Some Criteria for a Power Producing Thermonuclear Reactor; *Proceedings of the Physical Society of London Section B* 70 (1) 6–10 (1957); doi: <https://doi.org/10.1088/0370-1301/70/1/303> (cited on page: 9)
- [29] K. Pinkau, U. Schuhmacher; Kernfusion mit magnetisch eingeschlossenen Plasmen; *Physik in unserer Zeit* 13 138–154 (1982) (cited on pages: 9 and 10)

- [30] J. Wesson, D. J. Campbell; Tokamaks; Oxford University Press (2011); ISBN 9780199592234 (cited on pages: 9, 10, and 21)
- [31] E. Böhm-Vitense; Introduction to stellar astrophysics Volume 3; Cambridge University Press (1992); ISBN 9780521348713 (cited on page: 10)
- [32] R. Betti, O. A. Hurricane; Inertial-confinement fusion with lasers; *Nature Physics* 12 (5) 435–448 (2016); ISSN 1745-2473; doi: <https://doi.org/10.1038/Nphys3736> (cited on page: 10)
- [33] J. Ongena, R. Koch et al.; Magnetic-confinement fusion; *Nature Physics* 12 (5) 398–410 (2016); ISSN 1745-2473; doi: <https://doi.org/10.1038/Nphys3745> (cited on pages: 10 and 15)
- [34] P. R. Thomas, P. Andrew et al.; Observation of alpha heating in JET DT plasmas; *Physical Review Letters* 80 (25) 5548–5551 (1998); ISSN 0031-9007; doi: <https://doi.org/10.1103/PhysRevLett.80.5548> (cited on page: 10)
- [35] J. Paméla, E. R. Solano; From JET to ITER. Preparing the next step in fusion research; EURATOM/UKAEA Fusion Association (2001); <https://scipub.euro-fusion.org/archives/jet-archive/from-jet-to-iter-preparing-the-next-step-in-fusion-research> (cited on page: 10)
- [36] F. G. Rimini, P. Andrew et al.; Combined heating experiments in ELM-free H modes in JET; *Nuclear Fusion* 39 (11) 1591 (1999); ISSN 0029-5515; doi: <https://dx.doi.org/10.1088/0029-5515/39/11/310> (cited on page: 10)
- [37] J. P. Friedberg; Plasma Physics and Fusion Energy; Cambridge University Press (2010); ISBN 9780511755705; doi: <https://doi.org/10.1017/CBO9780511755705> (cited on page: 10)
- [38] S. Ciattaglia, G. Federici et al.; The European DEMO Fusion Reactor: Design Status and Challenges from Balance of Plant Point of View (2017); doi: <https://doi.org/10.1109/EEEIC.2017.7977853> (cited on page: 11)
- [39] I. Ignatov, O. Mosin; Modeling of possible processes for origin of life and living matter in hot mineral and seawater with deuterium; *Journal of Environment and Earth Science* 3 (14) 103–118 (2013); ISSN 2225-0948 (cited on page: 11)
- [40] L. L. Lucas, M. P. Unterweger; Comprehensive review and critical evaluation of the half-life of tritium; *Journal of Research of the National Institute of Standards and Technology* 105 (4) 541–549 (2000); ISSN 1044-677x; doi: <https://doi.org/10.6028/jres.105.043> (cited on page: 11)
- [41] N. K.; Nuclear Fusion; Cambridge university press (1989) (cited on page: 12)

-
- [42] R. A. Gross; Fusion Energy; John Wiley Sons (1984); ISBN 978-0471884705 (cited on page: 12)
 - [43] G. Van Oost; Fundamentals of magnetic fusion technology; International Atomic Energy Agency, Vienna (2023); ISBN 9789201104212 (cited on pages: 12, 15, 21, 34, 36, 48, and 143)
 - [44] L. A. El-Guebaly, S. Malang; Toward the ultimate goal of tritium self-sufficiency: Technical issues and requirements imposed on ARIES advanced power plants; *Fusion Engineering and Design* 84 (12) 2072–2083 (2009); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2008.12.098> (cited on pages: 12 and 16)
 - [45] M. E. Sawan, M. A. Abdou; Physics and technology conditions for attaining tritium self-sufficiency for the DT fuel cycle; *Fusion Engineering and Design* 81 (8-14) 1131–1144 (2006); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2005.07.035> (cited on pages: 12 and 16)
 - [46] S. L. Zheng, T. N. Todd; Study of impacts on tritium breeding ratio of a fusion DEMO reactor; *Fusion Engineering and Design* 98-99 1915–1918 (2015); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2015.06.171> (cited on page: 12)
 - [47] L. A. ElGuebaly; Breeding potential of candidate breeders for the US demo power plant; *Sofo '95 - 16th Ieee/Npss Symposium on Fusion Engineering, Vols 1 and 2* pp. 1198–1201 (1995); doi: <https://doi.org/10.1109/FUSION.1995.534441> (cited on page: 12)
 - [48] A. Pironti, M. Walker; Control of Tokamak plasmas - Introduction to the special section; *Ieee Control Systems Magazine* 25 (5) 24–29 (2005); ISSN 1066-033x; doi: <https://doi.org/10.1109/MCS.2005.1512793> (cited on page: 12)
 - [49] V. D. Shafranov; The initial period in the history of nuclear fusion research at the Kurchatov Institute; *Physics-Uspekhi* 44 (8) 835–843 (2001); ISSN 1063-7869; doi: <https://doi.org/10.1070/PU2001v044n08ABEH001068> (cited on page: 12)
 - [50] S. Li, H. Jiang et al.; Optimal Tracking for a Divergent-Type Parabolic PDE System in Current Profile Control; *Abstract and Applied Analysis* (2014); ISSN 1085-3375; doi: <https://doi.org/10.1155/2014/940965> (cited on page: 13)
 - [51] G. McCracken, P. Stott; Fusion : The Energy of the Universe; Elsevier Science Technology, Amsterdam (2005); ISBN 9780080492711 (cited on pages: 12 and 15)
-

- [52] A. H. Boozer; Plasma Confinement pp. 373–390; Academic Press, New York (2003); ISBN 978-0-12-227410-7; doi: <https://doi.org/10.1016/B0-12-227410-5/00582-2> (cited on page: 12)
- [53] B. Unterberg; Magnetically confined fusion plasmas with a radiating boundary and improved energy confinement; Berichte des Forschungszentrums Jülich, Jülich (2004); ISBN 0944-2952 (cited on pages: 13 and 14)
- [54] D. G. Whyte, A. E. Hubbard et al.; I-mode: an H-mode energy confinement regime with L-mode particle transport in Alcator C-Mod; *Nuclear Fusion* 50 (10) (2010); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/50/10/105005> (cited on page: 14)
- [55] S. C. Jardin, C. E. Kessel et al.; Physics basis for the advanced tokamak fusion power plant, ARIES-AT; *Fusion Engineering and Design* 80 (1-4) 25–62 (2006); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2005.06.352> (cited on page: 14)
- [56] M. Shimada, D. J. Campbell et al.; Progress in the ITER Physics Basis - Chapter 1: Overview and summary; *Nuclear Fusion* 47 (6) S1–S17 (2007); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/47/6/S01> (cited on page: 14)
- [57] R. Dux, A. Loarte et al.; Transport of tungsten in the H-mode edge transport barrier of ITER; *Plasma Physics and Controlled Fusion* 56 (12) (2014); ISSN 0741-3335; doi: <https://doi.org/10.1088/0741-3335/56/12/124003> (cited on page: 14)
- [58] K. Toi, F. Watanabe et al.; L-H Transition and Edge Transport Barrier Formation on Lhd; *Fusion Science and Technology* 58 (1) 61–69 (2010); ISSN 1536-1055; doi: <https://doi.org/10.13182/Fst10-A10794> (cited on page: 14)
- [59] H. Zohm; Edge localized modes (ELMs); *Plasma Physics and Controlled Fusion* 38 (2) 105–128 (1996); ISSN 0741-3335; doi: <https://dx.doi.org/10.1088/0741-3335/38/2/001> (cited on pages: 14 and 18)
- [60] M. Greenwald, J. L. Terry et al.; A new look at density limits in tokamaks; *Nuclear Fusion* 28 (12) 2199 (1988); ISSN 0029-5515; doi: <https://dx.doi.org/10.1088/0029-5515/28/12/009> (cited on page: 14)
- [61] S. Ishida, P. Barabaschi et al.; Overview of the JT-60SA project; *Nuclear Fusion* 51 (9) (2011); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/51/9/094018> (cited on page: 14)

-
- [62] I. Barlow, M. Bigi et al.; The error field correction coils on the JET machine; *Fusion Engineering and Design* 58-59 189–193 (2001); ISSN 0920-3796; doi: [https://doi.org/10.1016/S0920-3796\(01\)00430-6](https://doi.org/10.1016/S0920-3796(01)00430-6) (cited on page: 14)
- [63] G. Matsunaga, M. Takechi et al.; In-vessel coils for magnetic error field correction in JT-60SA; *Fusion Engineering and Design* 98-99 1113–1117 (2015); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2015.06.024> (cited on page: 14)
- [64] B. Bigot; Progress toward ITER’s First Plasma; *Nuclear Fusion* 59 (11) (2019); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/ab0f84> (cited on page: 14)
- [65] R. Aymar, P. Barabaschi et al.; The ITER design; *Plasma Physics and Controlled Fusion* 44 (5) 519–565 (2002); ISSN 0741-3335; doi: <https://doi.org/10.1088/0741-3335/44/5/304> (cited on pages: 15, 17, 18, and 34)
- [66] Y. Shimomura; ITER and plasma surface interaction issues in a fusion reactor; *Journal of Nuclear Materials* 363 467–475 (2007); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2007.01.215> (cited on page: 15)
- [67] V. Barabash; Role and contribution of ITER in research of materials and reactor components; *Journal of Nuclear Materials* 329 156–160 (2004); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2004.04.015> (cited on page: 15)
- [68] ITER Organization (2024); <https://www.iter.org> (cited on pages: 15 and 16)
- [69] F. Wagner; Fusion energy; *MRS Energy Sustainability* 5 (2018); doi: <https://doi.org/10.1557/mre.2018.8> (cited on page: 15)
- [70] L. M. Kovrizhnykh; The current status of the stellarator program; *Physics-Uspekhi* 52 (7) 730–735 (2009); ISSN 1063-7869; doi: <https://doi.org/10.3367/UFNe.0179.200907g.0772> (cited on page: 15)
- [71] T. Klinger, T. Andreeva et al.; Overview of first Wendelstein 7-X high-performance operation; *Nuclear Fusion* 59 (11) (2019); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/ab03a7> (cited on pages: 16 and 146)
- [72] A. Dinklage, C. D. Beidler et al.; Magnetic configuration effects on the Wendelstein 7-X stellarator; *Nature Physics* 14 (8) 855–+ (2018); ISSN 1745-2473; doi: <https://doi.org/10.1038/s41567-018-0141-9> (cited on page: 16)
- [73] Max-Planck-Gesellschaft (2024); <https://www.ipp.mpg.de/9296/einfuehrung> (cited on page: 16)
-

- [74] EUROfusion Consortium Research Institutions (2024); <https://euro-fusion.org/programme/demo/> (cited on page: 16)
- [75] A. J. Creely, M. J. Greenwald et al.; Overview of the SPARC tokamak; *Journal of Plasma Physics* 86 (5) (2020); ISSN 0022-3778; doi: <https://doi.org/10.1017/S0022377820001257> (cited on page: 16)
- [76] M. Windridge; Tokamak Energy; IOP Publishing (2020); ISBN 978-0-7503-2719-0; doi: <https://dx.doi.org/10.1088/978-0-7503-2719-0ch5> (cited on page: 16)
- [77] G. Federici, C. Bachmann et al.; DEMO design activity in Europe: Progress and updates; *Fusion Engineering and Design* 136 729–741 (2018); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2018.04.001> (cited on pages: 17 and 18)
- [78] P. Magaud, G. Marbach et al.; Nuclear Fusion Reactors pp. 365–381; Elsevier, New York (2004); ISBN 978-0-12-176480-7; doi: <https://doi.org/10.1016/B0-12-176480-X/00305-3> (cited on pages: 16 and 17)
- [79] A. W. Kleyn, W. Koppers et al.; Plasma-surface interaction in ITER; *Vacuum* 80 (10) 1098–1106 (2006); ISSN 0042-207x; doi: <https://doi.org/10.1016/j.vacuum.2006.02.019> (cited on page: 18)
- [80] T. Loewenhoff; Combined Steady State and High Cycle Transient Heat Load Simulation with the Electron Beam Facility JUDITH 2 volume 173; Forschungszentrum Jülich GmbH, Zentralbibliothek Verlag, Jülich (2012); ISBN 978-3-89336-869-3 (cited on pages: 19, 24, 59, 60, 61, 62, 111, 116, 119, 124, and 125)
- [81] M. Wirtz; Thermal shock behaviour of different tungsten grades under varying conditions; Forschungszentrum Jülich GmbH, Zentralbibliothek Verlag, Jülich (2012); ISBN 9783893368426 (cited on pages: 19, 23, 59, 69, 116, 117, 119, 120, and 124)
- [82] A. Loarte, G. Saibene et al.; Characteristics of type I ELM energy and particle losses in existing devices and their extrapolation to ITER; *Plasma Physics and Controlled Fusion* 45 (9) 1549–1569 (2003); ISSN 0741-3335; doi: <https://doi.org/10.1088/0741-3335/45/9/302> (cited on page: 19)
- [83] J. Linke; High heat flux performance of plasma facing materials and components under service conditions in future fusion reactors; *Fusion Science and Technology* 49 (2t) 455–464 (2006); ISSN 1536-1055; doi: <https://doi.org/10.13182/Fst06-A1144> (cited on pages: 19 and 21)

-
- [84] D. N. Hill; A review of ELMs in divertor tokamaks; *Journal of Nuclear Materials* 241 182–198 (1997); ISSN 0022-3115; doi: [https://dx.doi.org/10.1016/S0022-3115\(97\)80039-6](https://dx.doi.org/10.1016/S0022-3115(97)80039-6) (cited on page: 18)
- [85] J. Linke, P. Lorenzetto et al.; EU development of high heat flux components; *Fusion Science and Technology* 47 (3) 678–685 (2005); ISSN 1536-1055; doi: <https://doi.org/10.13182/Fst05-A764> (cited on pages: 18, 19, and 20)
- [86] M. Becoulet, G. Huysmans et al.; Edge localized modes control:: experiment and theory; *Journal of Nuclear Materials* 337 (1-3) 677–683 (2005); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2004.09.074> (cited on pages: 18 and 19)
- [87] E. J. Doyle, R. J. Groebner et al.; Modifications in Turbulence and Edge Electric-Fields at the L-H Transition in the DIII-D Tokamak; *Physics of Fluids B-Plasma Physics* 3 (8) 2300–2307 (1991); ISSN 0899-8221; doi: <https://doi.org/10.1063/1.859597> (cited on page: 18)
- [88] T. E. Evans; ELM mitigation techniques; *Journal of Nuclear Materials* 438 S11–S18 (2013); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2013.01.283> (cited on page: 19)
- [89] P. T. Lang, J. Neuhauser et al.; ELM frequency control by continuous small pellet injection in ASDEX Upgrade; *Nuclear Fusion* 43 (10) 1110–1120 (2003); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/43/10/012> (cited on page: 19)
- [90] P. T. Lang, A. Alonso et al.; ELM pacing investigations at JET with the new pellet launcher; *Nuclear Fusion* 51 (3) (2011); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/51/3/033010> (cited on page: 19)
- [91] G. Kocsis, S. Kalvin et al.; Spatio-temporal investigations on the triggering of pellet induced ELMs; *Nuclear Fusion* 47 (9) 1166–1175 (2007); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/47/9/013> (cited on page: 19)
- [92] Y. Liang, H. R. Koslowski et al.; Overview of ELM Control by Low n Magnetic Perturbations on JET; *Plasma and Fusion Research* 5 S2018–S2018 (2010); ISSN 1880-6821; doi: <https://doi.org/10.1585/pfr.5.S2018> (cited on page: 19)
- [93] N. Oyama; Progress and issues in understanding the physics of ELM dynamics, ELM mitigation, and ELM control; *11th Iaea Technical Meeting on H-Mode Physics and Transport Barriers* 123 (2008); ISSN 1742-6588; doi: <https://doi.org/10.1088/1742-6596/123/1/012002> (cited on page: 19)
- [94] A. W. Leonard, T. H. Osborne et al.; Tolerable ELMs at high density in DIII-D; *Journal of Nuclear Materials* 290 1097–1101 (2001); ISSN 0022-3115; doi: [https://doi.org/10.1016/S0022-3115\(00\)00503-1](https://doi.org/10.1016/S0022-3115(00)00503-1) (cited on page: 20)
-

- [95] O. Gruber, K. Lackner et al.; Vertical Displacement Events and Halo Currents; *Plasma Physics and Controlled Fusion* 35 B191–B204 (1993); ISSN 0741-3335; doi: <https://doi.org/10.1088/0741-3335/35/Sb/015> (cited on page: 20)
- [96] A. Hassanein, T. Sizyuk et al.; Vertical displacement events: A serious concern in future ITER operation; *Fusion Engineering and Design* 83 (7-9) 1020–1024 (2008); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2008.05.032> (cited on page: 20)
- [97] M. Merola, G. Vieider et al.; European achievements for ITER high heat flux components; *Fusion Engineering and Design* 56-57 173–178 (2001); ISSN 0920-3796; doi: [https://doi.org/10.1016/S0920-3796\(01\)00253-8](https://doi.org/10.1016/S0920-3796(01)00253-8) (cited on page: 20)
- [98] J. Roth, E. Tsitrone et al.; Recent analysis of key plasma wall interactions issues for ITER; *Journal of Nuclear Materials* 390-91 1–9 (2009); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2009.01.037> (cited on page: 21)
- [99] A. S. Kukushkin, H. D. Pacher et al.; Effect of neutral transport on ITER divertor performance; *Nuclear Fusion* 45 (7) 608–616 (2005); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/45/7/008> (cited on page: 21)
- [100] G. Federici, J. N. Brooks et al.; Assessment of erosion and tritium codeposition in ITER-FEAT; *Journal of Nuclear Materials* 290 260–265 (2001); ISSN 0022-3115; doi: [https://doi.org/10.1016/S0022-3115\(00\)00627-9](https://doi.org/10.1016/S0022-3115(00)00627-9) (cited on page: 21)
- [101] H. Bolt, V. Barabash et al.; Plasma facing and high heat flux materials - needs for ITER and beyond; *Journal of Nuclear Materials* 307 43–52 (2002); ISSN 0022-3115; doi: [https://doi.org/10.1016/S0022-3115\(02\)01175-3](https://doi.org/10.1016/S0022-3115(02)01175-3) (cited on page: 21)
- [102] J. W. Coenen, S. Antusch et al.; Materials for DEMO and reactor applications- boundary conditions and new concepts; *Physica Scripta* T167 (2016); ISSN 0031-8949; doi: <https://doi.org/10.1088/0031-8949/2016/T167/014002> (cited on pages: 21, 25, and 26)
- [103] B. T. Kelly; *Irradiation Damage to Solids*; Pergamon Press, Oxford (1966); ISBN 978-0080119106 (cited on page: 21)
- [104] C. P. C. Wong; Neutron wall loading of Tokamak reactors; *Journal of Nuclear Materials* 283 588–592 (2000); ISSN 0022-3115; doi: [https://doi.org/10.1016/S0022-3115\(00\)00145-8](https://doi.org/10.1016/S0022-3115(00)00145-8) (cited on page: 21)
- [105] C. H. Wu, C. Alessandrini et al.; Progress of the European RD on plasma-wall interactions, neutron effects and tritium removal in ITER plasma facing materials; *Fusion Engineering and Design* 56-57 179–187 (2001); ISSN 0920-3796; doi: [https://doi.org/10.1016/S0920-3796\(01\)00255-1](https://doi.org/10.1016/S0920-3796(01)00255-1) (cited on page: 21)

- [106] E. Lassner, W. Schubert; Tungsten: Properties, Chemistry, Technology of the Element, Alloys, and Chemical Compound; 1st ed. (1999); doi: <https://doi.org/10.1007/978-1-4615-4907-9> (cited on pages: 22, 24, 31, 69, 70, 71, 73, 89, 128, and 130)
- [107] M. Rieth, R. Doerner et al.; Behavior of tungsten under irradiation and plasma interaction; *Journal of Nuclear Materials* 519 334–368 (2019); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2019.03.035> (cited on page: 22)
- [108] T. Hirai, G. Pintsuk; Thermo-mechanical calculations on operation temperature limits of tungsten as plasma facing material; *Fusion Engineering and Design* 82 (4) 389–393 (2007); ISSN 09203796; doi: <https://doi.org/10.1016/j.fusengdes.2007.03.032> (cited on page: 23)
- [109] Y. Ueda, K. Schmid et al.; Baseline high heat flux and plasma facing materials for fusion; *Nuclear Fusion* 57 (9) (2017); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/aa6b60> (cited on pages: 23, 24, 119, and 124)
- [110] M. Wirtz, I. Uytdenhouwen et al.; Material properties and their influence on the behaviour of tungsten as plasma facing material; *Nuclear Fusion* 57 (6) (2017); ISSN 0029-5515 1741-4326; doi: <https://doi.org/10.1088/1741-4326/aa6938> (cited on pages: 23 and 128)
- [111] M. Wirtz, J. Linke et al.; Transient heat load challenges for plasma-facing materials during long-term operation; *Nuclear Materials and Energy* 12 148–155 (2017); ISSN 23521791; doi: <https://doi.org/10.1016/j.nme.2016.12.024> (cited on pages: 23, 100, 116, and 128)
- [112] M. Wirtz, J. Linke et al.; Investigation of materials under high repetition and intense fusion pulses pp. p. 38–42; International Atomic Energy Agency (IAEA) (2017); ISBN ISBN 978-92-0-108217-6 (cited on page: 23)
- [113] M. Wirtz, J. Linke et al.; Thermal shock tests to qualify different tungsten grades as plasma facing material; *Physica Scripta* T167 (2016); ISSN 0031-8949 1402-4896; doi: <https://doi.org/10.1088/0031-8949/t167/1/014015> (cited on pages: 23, 88, and 128)
- [114] J. Linke, T. Loewenhoff et al.; Performance of different tungsten grades under transient thermal loads; *Nuclear Fusion* 51 (7) (2011); ISSN 0029-5515 1741-4326; doi: <https://doi.org/10.1088/0029-5515/51/7/073017> (cited on pages: 23, 125, and 128)
- [115] A. S. Arakcheev, A. Huber et al.; Theoretical investigation of crack formation in tungsten after heat loads; *Journal of Nuclear Materials* 463 246–249 (2015);

- ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2014.10.090> (cited on pages: 23, 119, and 120)
- [116] D. Wang, K. Li et al.; Cracking Behavior in Additively Manufactured Pure Tungsten; *Acta Metallurgica Sinica (English Letters)* 32 (1) 127–135 (2019); ISSN 2194-1289; doi: <https://doi.org/10.1007/s40195-018-0752-2> (cited on page: 23)
- [117] S. Nogami, A. Hasegawa et al.; Mechanical properties of tungsten: Recent research on modified tungsten materials in Japan; *Journal of Nuclear Materials* 543 152506 (2021); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2020.152506> (cited on page: 24)
- [118] T. Pütterich, R. Neu et al.; Calculation and experimental test of the cooling factor of tungsten; *Nuclear Fusion* 50 (2) (2010); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/50/2/025012> (cited on pages: 24 and 120)
- [119] L. Singheiser, T. Hirai et al.; Plasma-facing materials for thermo-nuclear fusion devices; *Transactions of the Indian Institute of Metals* 62 (2) 123–128 (2009); ISSN 0972-2815; doi: <https://doi.org/10.1007/s12666-009-0016-y> (cited on pages: 24 and 120)
- [120] R. Behrisch, G. Federici et al.; Material erosion at the vessel walls of future fusion devices; *Journal of Nuclear Materials* 313 388–392 (2003); ISSN 0022-3115; doi: [https://doi.org/10.1016/S0022-3115\(02\)01580-5](https://doi.org/10.1016/S0022-3115(02)01580-5) (cited on pages: 25, 48, and 143)
- [121] I. Tanihata, H. Toki et al.; Handbook of Nuclear Physics; Springer, Singapore 1st edition (2023); ISBN 978-981-19-6345-2; doi: <https://doi.org/10.1007/978-981-19-6345-2> (cited on page: 25)
- [122] R. A. Forrest; Handbook of activation data calculated using Easy-2007 (2009) (cited on pages: 25 and 26)
- [123] D. Maisonnier, I. Cook et al.; A conceptual study of commercial fusion power plants. Final Report of the European Fusion Power Plant Conceptual Study (PPCS); *EFDA report number EFDA (05)-27/4.10* 1 (2005) (cited on page: 25)
- [124] G. Federici, G. Giruzzi et al.; EU demo design and rd studies; In 2013 IEEE 25th Symposium on Fusion Engineering (SOFE) pp. 1–8. IEEE; ISBN 1479901717 (cited on page: 25)
- [125] J. Knaster, A. Moeslang et al.; Materials research for fusion; *Nature Physics* 12 (5) 424–434 (2016); ISSN 1745-2473; doi: <https://doi.org/10.1038/Nphys3735> (cited on page: 25)

- [126] A. Ibarra, F. Arbeiter et al.; The IFMIF-DONES project: preliminary engineering design; *Nuclear Fusion* 58 (10) (2018); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/aad91f> (cited on page: 25)
- [127] J. P. Bonal, C. H. Wu; Neutron irradiation effects on the thermal conductivity and the dimensional stability of carbon fiber composites; *Physica Scripta* T64 26–31 (1996); ISSN 0031-8949; doi: <https://doi.org/10.1088/0031-8949/1996/T64/003>; <GotoISI>://WOS:A1996VD17600004; Vd176 Times Cited:9 Cited References Count:9 (cited on page: 25)
- [128] J. Pamela, G. F. Matthews et al.; An ITER-like wall for JET; *Journal of Nuclear Materials* 363 1–11 (2007); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2006.12.056> (cited on pages: 27 and 30)
- [129] F. Romanelli, M. Laxåback et al.; Overview of JET results; *Nuclear Fusion* 51 (9) (2011); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/51/9/094008> (cited on pages: 27 and 30)
- [130] B. Spilker; The impact of transient thermal loads on beryllium as plasma facing material; Forschungszentrum Jülich GmbH, Zentralbibliothek Verlag, Jülich (2017); ISBN 978-3-95806-227-6; doi: <https://doi.org/10.18154/RWTH-2017-01947> (cited on page: 28)
- [131] A. Paredes, H. Bufferand et al.; A penalization technique to model plasma facing components in a tokamak with temperature variations; *Journal of Computational Physics* 274 283–298 (2014); ISSN 0021-9991; doi: <https://doi.org/10.1016/j.jcp.2014.05.025> (cited on page: 28)
- [132] Y. Feng, F. Sardei et al.; Physics of island divertors as highlighted by the example of W7-AS; *Nuclear Fusion* 46 (8) 807–819 (2006); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/46/8/006> (cited on page: 28)
- [133] M. Kotschenreuther, P. Valanju et al.; The super X divertor (SXD) and a compact fusion neutron source (CFNS); *Nuclear Fusion* 50 (3) (2010); ISSN 0029-5515; doi: <https://doi.org/10.1088/0029-5515/50/3/035003> (cited on page: 28)
- [134] R. A. Pitts, A. Kukushkin et al.; Status and physics basis of the ITER divertor; *Physica Scripta* 2009 (T138) 014001 (2009); ISSN 1402-4896 0031-8949; doi: <https://doi.org/10.1088/0031-8949/2009/T138/014001> (cited on pages: 29 and 33)
- [135] R. Mitteau, R. Eaton et al.; Allowable heat load on the edge of the ITER first wall panel beryllium flat tiles; *Nuclear Materials and Energy* 12 1067–

- 1070 (2017); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2017.02.001> (cited on page: 29)
- [136] S. Banetta, F. Zaccchia et al.; Manufacturing of small-scale mock-ups and of a semi-prototype of the ITER Normal Heat Flux First Wall; *Fusion Engineering and Design* 89 (7) 970–974 (2014); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2014.04.020> (cited on page: 29)
- [137] S. Heuer, J. W. Coenen et al.; Overview of challenges and developments in joining tungsten and steel for future fusion reactors; *Physica Scripta* T171 (1) (2020); ISSN 0031-8949; doi: <https://doi.org/10.1088/1402-4896/ab47a4> (cited on pages: 29 and 30)
- [138] F. Maviglia, C. Bachmann et al.; Integrated design strategy for EU-DEMO first wall protection from plasma transients; *Fusion Engineering and Design* 177 (2022); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2022.113067> (cited on page: 29)
- [139] V. Ganesh; Joining of tungsten and steel for the first wall of a future fusion reactor; Forschungszentrum Jülich GmbH, Zentralbibliothek, Verlag, Jülich (2023); ISBN 978-3-95806-715-8; doi: <https://doi.org/10.34734/FZJ-2023-03599> (cited on page: 29)
- [140] S. Heuer; Charakterisierung gradiert Eisen/Wolfram-Schichten für die erste Wand von Fusionsreaktoren; Forschungszentrum Jülich GmbH, Zentralbibliothek, Verlag, Jülich (2017); ISBN 978-3-95806-252-8 (cited on page: 29)
- [141] H. Tanigawa, K. Shiba et al.; Status and key issues of reduced activation ferritic/martensitic steels as the structural material for a DEMO blanket; *Journal of Nuclear Materials* 417 (1) 9–15 (2011); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2011.05.023> (cited on page: 29)
- [142] M. Rieth, M. Schirra et al.; EUROFER 97. Tensile, charpy, creep and structural tests; Karlsruhe Institut für Technologie, Karlsruhe (2003); doi: <https://doi.org/10.5445/IR/270055720> (cited on page: 30)
- [143] EUROfusion; DEMO Material Property Handbook (MPH): EUROFER 97 (cited on page: 30)
- [144] R. A. Pitts, S. Bardin et al.; Physics conclusions in support of ITER W divertor monoblock shaping; *Nuclear Materials and Energy* 12 60–74 (2017); doi: <https://doi.org/10.1016/j.nme.2017.03.005> (cited on pages: 31, 34, 121, and 122)
- [145] H. Sicius; Handbook of the Chemical Elements; Springer-Verlag GmbH, Berlin (2024); ISBN 9783662689219; doi: <https://doi.org/10.1007/978-3-662-68921-9> (cited on page: 31)

- [146] K. Zhang, E. Gaganidze et al.; Development of the material property handbook and database of CuCrZr; *Fusion Engineering and Design* 144 148–153 (2019); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2019.04.094> (cited on page: 31)
- [147] J. H. You, G. Mazzone et al.; Conceptual design studies for the European DEMO divertor: Rationale and first results; *Fusion Engineering and Design* 109-111 1598–1603 (2016); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2015.11.012> (cited on page: 31)
- [148] J. H. You; Copper matrix composites as heat sink materials for water-cooled divertor target; *Nuclear Materials and Energy* 5 7–18 (2015); doi: <https://doi.org/10.1016/j.nme.2015.10.001> (cited on page: 31)
- [149] V. Barabash, A. Peacock et al.; Materials challenges for ITER – Current status and future activities; *Journal of Nuclear Materials* 367-370 21–32 (2007); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2007.03.017> (cited on page: 31)
- [150] E. Visca, S. Libera et al.; Hot radial pressing: An alternative technique for the manufacturing of plasma-facing components; *Fusion Engineering and Design* 75-79 485–489 (2005); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2005.06.123> (cited on page: 31)
- [151] K. Poleshchuk, D. Terentyev et al.; Strength of a circumferential W/Cu/CuCrZr joint of ITER monoblock; *Nuclear Materials and Energy* 43 101947 (2025); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2025.101947> (cited on page: 31)
- [152] S. Kakudate, N. Takeda et al.; RD Activities for ITER Blanket Remote Handling Equipment; In Proc. 22nd IAEA Fusion Energy Conf. pp. 1–8. International Atomic Energy Agency; https://www-pub.iaea.org/mtcd/meetings/fec2008/it_p7-18.pdf (cited on page: 32)
- [153] D. Leichtle, P. Pereslavytsev et al.; Global shutdown dose rate maps for a DEMO conceptual design; *Fusion Engineering and Design* 98-99 1524–1527 (2015); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2015.03.006> (cited on page: 32)
- [154] J. Palmer, M. Irving et al.; The design and development of divertor remote handling equipment for ITER; *Fusion Engineering and Design* 82 (15) 1977–1982 (2007); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2007.05.073> (cited on page: 32)
- [155] Y. Noguchi, M. Saito et al.; Design progress of ITER blanket remote handling system towards manufacturing; *Fusion Engineering and Design* 136 722–728

- (2018); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2018.03.068> (cited on pages: 33 and 34)
- [156] T. Maruyama, A. Aburadani et al.; Robot vision system RD for ITER blanket remote-handling system; *Fusion Engineering and Design* 89 (9) 2404–2408 (2014); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2014.01.004> (cited on page: 34)
- [157] C.-H. Choi, A. Tesini et al.; Multi-purpose deployer for ITER in-vessel maintenance; *Fusion Engineering and Design* 98-99 1448–1452 (2015); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2015.06.156> (cited on pages: 34 and 35)
- [158] D. McGarrigle, E. Flynn et al.; Analysis of Existing and Proposed Maintenance Deployment Systems Toward DEMO MPD Development; *IEEE Transactions on Plasma Science* 50 (11) 4485–4490 (2022); ISSN 1939-9375; doi: <https://doi.org/10.1109/TPS.2022.3163466> (cited on pages: 34, 35, and 36)
- [159] DIN Deutsches Institut für Normung e.V.; Manufacturing processes - Terms and definitions, division (DIN 8580:2022-12) (2022); doi: <https://dx.doi.org/10.31030/3217872> (cited on page: 37)
- [160] DIN Deutsches Institut für Normung e.V.; Additive manufacturing - General principles - Fundamentals and vocabulary (ISO/ASTM 52900:2022-03) (2022); doi: <https://dx.doi.org/10.31030/3290011> (cited on pages: 37, 38, and 47)
- [161] H. Zeyn; Industrialisierung der Additiven Fertigung. Digitalisierte Prozesskette - von der Entwicklung bis zum einsetzbaren Artikel; DIN Media GmbH, Berlin, Wien, Zürich (2017); ISBN 978-3-410-26919-9 (cited on pages: 37 and 38)
- [162] G. Witt, A. Wegner et al.; Neue Entwicklungen in der Additiven Fertigung; Springer Nature, Berlin, Heidelberg (2015); ISBN 978-3662484722; doi: <https://doi.org/10.1007/978-3-662-48473-9> (cited on page: 37)
- [163] S. Antusch, A. Klein et al.; Additive manufacturing of novel complex tungsten components via electron beam melting: Basic properties and evaluation of the high heat flux behavior; *Nuclear Materials and Energy* 39 101683 (2024); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2024.101683> (cited on pages: 37, 45, and 46)
- [164] T. Childerhouse, M. Jackson; Near Net Shape Manufacture of Titanium Alloy Components from Powder and Wire: A Review of State-of-the-Art Process Routes; *Metals* 9 (6) (2019); doi: <https://doi.org/10.3390/met9060689> (cited on page: 37)

- [165] T. Heeling, K. Wegener; The effect of multi-beam strategies on selective laser melting of stainless steel 316L; *Additive Manufacturing* 22 334–342 (2018); ISSN 2214-8604; doi: <https://doi.org/10.1016/j.addma.2018.05.026> (cited on page: 37)
- [166] A. Gebhardt; Additive Fertigungsverfahren. Additive Manufacturing und 3D-Drucken für Prototyping - Tooling- Produktion; Carl Hanser Verlag, München (2017); ISBN 978-3-446-44401-0 (cited on page: 38)
- [167] I. Tabernero, A. Paskual et al.; Study on Arc Welding Processes for High Deposition Rate Additive Manufacturing; *Procedia CIRP* 68 358–362 (2018); ISSN 2212-8271; doi: <https://doi.org/10.1016/j.procir.2017.12.095> (cited on pages: 38 and 49)
- [168] H. A. Richard, B. Schramm et al.; Additive Fertigung von Bauteilen und Strukturen; Springer Vieweg, Wiesbaden (2017); ISBN 9783658177799; doi: <https://doi.org/10.1007/978-3-658-27412-2> (cited on page: 38)
- [169] J. Tweer, R. Day et al.; Initial experiments to regenerate the surface of plasma-facing components by wire-based laser metal deposition; *Nuclear Materials and Energy* 38 101577 (2024); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2023.101577> (cited on pages: 38, 70, 72, 74, 75, 76, 77, 78, 79, 115, 120, 123, 132, and 143)
- [170] J. Tweer, R. Day et al.; Repair of heat load damaged plasma-facing material using the wire-based laser metal deposition process; *Nuclear Materials and Energy* 41 101787 (2024); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2024.101787> (cited on pages: 38, 127, 128, 129, 131, 132, 133, 134, 135, 137, 138, and 145)
- [171] J. Tweer, T. Derra et al.; High heat flux testing of wire-based laser metal deposition coated plasma-facing components; *Nuclear Materials and Energy* 45 101983 (2025); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2025.101983> (cited on pages: 38, 96, 97, 98, 99, 100, 101, 102, 103, 104, 105, 106, 107, 108, 109, 110, 111, 112, 113, 118, and 120)
- [172] E. O. Olakanmi, R. F. Cochrane et al.; A review on selective laser sintering/melting (SLS/SLM) of aluminium alloy powders: Processing, microstructure, and properties; *Progress in Materials Science* 74 401–477 (2015); ISSN 0079-6425; doi: <https://doi.org/10.1016/j.pmatsci.2015.03.002> (cited on page: 38)
- [173] C. Körner; Additive manufacturing of metallic components by selective electron beam melting — a review; *International Materials Reviews* 61 (5) 361–377

- (2016); ISSN 0950-6608; doi: <https://doi.org/10.1080/09506608.2016.1176289> (cited on page: 38)
- [174] J. Parthasarathy, B. Starly et al.; Mechanical evaluation of porous titanium (Ti6Al4V) structures with electron beam melting (EBM); *Journal of the Mechanical Behavior of Biomedical Materials* 3 (3) 249–259 (2010); ISSN 1751-6161; doi: <https://doi.org/10.1016/j.jmbbm.2009.10.006> (cited on page: 38)
- [175] W. Meiners; Direktes Selektives Laser Sintern einkomponentiger metallischer Werkstoffe; Shaker Verlag (1999); ISBN 978-3826565717 (cited on page: 40)
- [176] U. Scipioni Bertoli, A. J. Wolfer et al.; On the limitations of Volumetric Energy Density as a design parameter for Selective Laser Melting; *Materials Design* 113 331–340 (2017); ISSN 0264-1275; doi: <https://doi.org/10.1016/j.matdes.2016.10.037> (cited on page: 40)
- [177] F. Caiazzo, V. Alfieri et al.; On the Relevance of Volumetric Energy Density in the Investigation of Inconel 718 Laser Powder Bed Fusion; *Materials* 13 (3) (2020); doi: <https://doi.org/10.3390/ma13030538> (cited on page: 40)
- [178] G. O. Barrionuevo, J. Ramos-Grez et al.; Numerical analysis of the effect of processing parameters on the microstructure of stainless steel 316L manufactured by laser-based powder bed fusion; *Materials Today: Proceedings* 59 93–100 (2022); ISSN 2214-7853; doi: <https://doi.org/10.1016/j.matpr.2021.10.209> (cited on page: 40)
- [179] G. V. de Leon Nope, L. I. Perez-Andrade et al.; Study of volumetric energy density limitations on the IN718 mesostructure and microstructure in laser powder bed fusion process; *Journal of Manufacturing Processes* 64 1261–1272 (2021); ISSN 1526-6125; doi: <https://doi.org/10.1016/j.jmapro.2021.02.043> (cited on page: 40)
- [180] W. Xing, D. Ouyang et al.; Effect of energy density on defect evolution in 3D printed Zr-based metallic glasses by selective laser melting; *Science China-Physics Mechanics Astronomy* 63 (2) (2020); ISSN 1674-7348; doi: <https://doi.org/10.1007/s11433-019-1485-8> (cited on page: 40)
- [181] S. K. Nayak, S. K. Mishra et al.; Effect of energy density on laser powder bed fusion built single tracks and thin wall structures with 100 μm preplaced powder layer thickness; *Optics Laser Technology* 125 106016 (2020); ISSN 0030-3992; doi: <https://doi.org/10.1016/j.optlastec.2019.106016> (cited on page: 40)
- [182] M. Thomas, G. J. Baxter et al.; Normalised model-based processing diagrams for additive layer manufacture of engineering alloys; *Acta Materialia* 108 26–35

- (2016); ISSN 1359-6454; doi: <https://doi.org/10.1016/j.actamat.2016.02.025> (cited on page: 40)
- [183] J. Wegner; Functionalisation of the process-inherent cooling rates of the laser powder bed fusion process for the fabrication of bulk metallic glasses; Universität Duisburg-Essen, Duisburg (2023); doi: <https://doi.org/10.17185/dupublico/78650> (cited on pages: 41, 42, 43, and 60)
- [184] K. C. Mills, B. J. Keene et al.; Marangoni effects in welding; *Philosophical Transactions of the Royal Society a-Mathematical Physical and Engineering Sciences* 356 (1739) 911–925 (1998); ISSN 1364-503x; doi: <https://doi.org/10.1098/rsta.1998.0196> (cited on pages: 42 and 43)
- [185] R. S. Vidyarthi, D. K. Dwivedi; Activating flux tungsten inert gas welding for enhanced weld penetration; *Journal of Manufacturing Processes* 22 211–228 (2016); ISSN 1526-6125; doi: <https://doi.org/10.1016/j.jmapro.2016.03.012> (cited on page: 42)
- [186] S. H. Pan, G. C. Yao et al.; Additive manufacturing of tungsten, tungsten-based alloys, and tungsten matrix composites; *Tungsten* 5 (1) 1–31 (2023); ISSN 2661-8036; doi: <https://doi.org/10.1007/s42864-022-00153-6> (cited on pages: 42, 45, and 74)
- [187] J. Xie, H. Lu et al.; Additive manufacturing of tungsten using directed energy deposition for potential nuclear fusion application; *Surface and Coatings Technology* 409 126884 (2021); ISSN 0257-8972; doi: <https://doi.org/10.1016/j.surfcoat.2021.126884> (cited on pages: 42 and 74)
- [188] D. D. Gu, W. Meiners et al.; Laser additive manufacturing of metallic components: materials, processes and mechanisms; *International Materials Reviews* 57 (3) 133–164 (2013); ISSN 0950-6608 1743-2804; doi: <https://doi.org/10.1179/1743280411y.0000000014> (cited on pages: 42 and 74)
- [189] C. Tan, K. Zhou et al.; Selective laser melting of high-performance pure tungsten: parameter design, densification behavior and mechanical properties; *Sci Technol Adv Mater* 19 (1) 370–380 (2018); ISSN 1878-5514; doi: <https://doi.org/10.1080/14686996.2018.1455154> (cited on pages: 42 and 74)
- [190] I. Takamichi, I. Roderick et al.; The Physical Properties of Liquid Metals; Clarendon Press, Oxford (1993); ISBN 978-0198563945 (cited on page: 42)

- [191] E. Ahmadi, A. R. Ebrahimi; Welding of 316L Austenitic Stainless Steel with Activated Tungsten Inert Gas Process; *Journal of Materials Engineering and Performance* 24 (2) 1065–1071 (2015); ISSN 1544-1024; doi: <https://doi.org/10.1007/s11665-014-1336-6> (cited on page: 42)
- [192] M. Rombouts, J. P. Kruth et al.; Fundamentals of Selective Laser Melting of alloyed steel powders; *CIRP Annals* 55 (1) 187–192 (2006); ISSN 0007-8506; doi: [https://doi.org/10.1016/S0007-8506\(07\)60395-3](https://doi.org/10.1016/S0007-8506(07)60395-3) (cited on pages: 42 and 43)
- [193] A. v. Müller, G. Schlick et al.; Additive manufacturing of pure tungsten by means of selective laser beam melting with substrate preheating temperatures up to 1000 C; *Nuclear Materials and Energy* 19 184–188 (2019); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2019.02.034> (cited on pages: 44, 45, 46, 89, 91, and 145)
- [194] P. Bajaj, A. Hariharan et al.; Steels in additive manufacturing: A review of their microstructure and properties; *Materials Science and Engineering: A* 772 138633 (2020); ISSN 0921-5093; doi: <https://doi.org/10.1016/j.msea.2019.138633> (cited on page: 44)
- [195] X. Tang, X. Chen et al.; The current state of CuCrZr and CuCrNb alloys manufactured by additive manufacturing: A review; *Materials Design* 224 111419 (2022); ISSN 0264-1275; doi: <https://doi.org/10.1016/j.matdes.2022.111419> (cited on page: 44)
- [196] P. Morcos, A. Elwany et al.; Review: additive manufacturing of pure tungsten and tungsten-based alloys; *Journal of Materials Science* 57 (21) 9769–9806 (2022); ISSN 1573-4803; doi: <https://doi.org/10.1007/s10853-022-07183-y> (cited on page: 45)
- [197] O. Gokcekaya, T. Ishimoto et al.; Influence of powder characteristics on densification via crystallographic texture formation: Pure tungsten prepared by laser powder bed fusion; *Additive Manufacturing Letters* 1 100016 (2021); ISSN 2772-3690; doi: <https://doi.org/10.1016/j.addlet.2021.100016> (cited on page: 45)
- [198] B. Vrancken, R. K. Ganeriwala et al.; Analysis of laser-induced microcracking in tungsten under additive manufacturing conditions: Experiment and simulation; *Acta Materialia* 194 464–472 (2020); ISSN 1359-6454; doi: <https://doi.org/10.1016/j.actamat.2020.04.060> (cited on page: 45)
- [199] D.-Z. Wang, K.-L. Li et al.; Cracking Behavior in Additively Manufactured Pure Tungsten; *Acta Metallurgica Sinica (English Letters)* 32 (1) 127–135 (2019); ISSN 2194-1289; doi: <https://doi.org/10.1007/s40195-018-0752-2> (cited on page: 45)

- [200] R. Neu, J. W. Coenen et al.; Material and component developments for the DEMO divertor using fibre reinforcement and additive manufacturing; *Materials Research Express* 10 (11) 116516 (2023); ISSN 2053-1591; doi: <https://dx.doi.org/10.1088/2053-1591/ad0c7c> (cited on pages: 45 and 46)
- [201] A. v. Müller, D. Dorow-Gerspach et al.; Progress in additive manufacturing of pure tungsten for plasma-facing component applications; *Journal of Nuclear Materials* 566 (2022); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2022.153760> (cited on pages: 45, 89, 91, and 145)
- [202] B. Curzadd, A. von Müller et al.; Topology optimization of tungsten/copper structures for plasma-facing component applications; *Nuclear Fusion* 59 (8) (2019); ISSN 0029-5515; doi: <https://doi.org/10.1088/1741-4326/ab1ff5> (cited on page: 45)
- [203] A. Talignani, R. Seede et al.; A review on additive manufacturing of refractory tungsten and tungsten alloys; *Additive Manufacturing* 58 103009 (2022); ISSN 2214-8604; doi: <https://doi.org/10.1016/j.addma.2022.103009> (cited on page: 45)
- [204] L. Howard, G. D. Parker et al.; Progress and Challenges of Additive Manufacturing of Tungsten and Alloys as Plasma-Facing Materials; *Materials* 17 (9) (2024); doi: <https://doi.org/10.3390/ma17092104> (cited on pages: 45 and 47)
- [205] D. Dorow-Gerspach, A. Kirchner et al.; Additive manufacturing of high density pure tungsten by electron beam melting; *Nuclear Materials and Energy* 28 (2021); ISSN 23521791; doi: <https://doi.org/10.1016/j.nme.2021.101046> (cited on pages: 45, 46, 47, 88, 89, 91, and 145)
- [206] Y. Tan, J. Wang et al.; Additive manufacturing of W/RAFM hypervapotron plasma-facing components and the steady state thermal fatigue behavior; *Journal of Nuclear Materials* 601 155333 (2024); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2024.155333> (cited on page: 46)
- [207] F. Escourbiac, J. Schlosser et al.; Experimental optimisation of a hypervapotron® concept for ITER plasma facing components; *Fusion Engineering and Design* 66-68 301–304 (2003); ISSN 0920-3796; doi: [https://doi.org/10.1016/S0920-3796\(03\)00172-8](https://doi.org/10.1016/S0920-3796(03)00172-8) (cited on page: 46)
- [208] D. G. Ahn; Directed Energy Deposition (DED) Process: State of the Art; *International Journal of Precision Engineering and Manufacturing-Green Technology* 8 (2) 703–742 (2021); ISSN 2198-0810; doi: <https://doi.org/10.1007/s40684-020-00302-7> (cited on pages: 47, 48, and 49)

- [209] A. Saboori, A. Aversa et al.; Application of Directed Energy Deposition-Based Additive Manufacturing in Repair (2019); doi: <https://doi.org/10.3390/app9163316> (cited on page: 48)
- [210] H. Koehler, K. Partes et al.; Laser reconditioning of crankshafts: From lab to application; *Physics Procedia* 5 387–397 (2010); ISSN 1875-3892; doi: <https://doi.org/10.1016/j.phpro.2010.08.160> (cited on page: 48)
- [211] J. M. Wilson, C. Piya et al.; Remanufacturing of turbine blades by laser direct deposition with its energy and environmental impact analysis; *Journal of Cleaner Production* 80 170–178 (2014); ISSN 0959-6526; doi: <https://doi.org/10.1016/j.jclepro.2014.05.084> (cited on page: 48)
- [212] O. Yilmaz, N. Gindy et al.; A repair and overhaul methodology for aeroengine components; *Robotics and Computer-Integrated Manufacturing* 26 (2) 190–201 (2010); ISSN 0736-5845; doi: <https://doi.org/10.1016/j.rcim.2009.07.001> (cited on page: 48)
- [213] A. Dass, A. Moridi; State of the Art in Directed Energy Deposition: From Additive Manufacturing to Materials Design (2019); doi: <https://doi.org/10.3390/coatings9070418> (cited on pages: 48 and 49)
- [214] M. Z. Tokar; An assessment for the erosion rate of DEMO first wall; *Nuclear Fusion* 58 (1) (2018); ISSN 0029-5515 1741-4326; doi: <https://doi.org/10.1088/1741-4326/aa92dd> (cited on page: 48)
- [215] M. Gipperich; Investigation of a Dual-Beam Process as an Approach to Increase Process Stability in Wire-Based Laser Metal Deposition; Apprimus Verlag, Aachen (2023); ISBN 978-3-98555-151-4; doi: 10.18154/RWTH-2023-02618 (cited on pages: 48 and 57)
- [216] J. Foster, C. Cullen et al.; Remanufacture of hot forging tools and dies using laser metal deposition with powder and a hard-facing alloy Stellite 21®; *Journal of Remanufacturing* 9 (3) 189–203 (2019); ISSN 2210-4690; doi: <https://doi.org/10.1007/s13243-018-0063-9> (cited on page: 48)
- [217] A. Azarniya, X. G. Colera et al.; Additive manufacturing of Ti-6Al-4V parts through laser metal deposition (LMD): Process, microstructure, and mechanical properties; *Journal of Alloys and Compounds* 804 163–191 (2019); ISSN 0925-8388; doi: <https://doi.org/10.1016/j.jallcom.2019.04.255> (cited on page: 48)
- [218] R. M. Mahamood, E. T. Akinlabi; Processing Parameters Optimization for Material Deposition Efficiency in Laser Metal Deposited Titanium Alloy; *Lasers in Manufacturing and Materials Processing* 3 (1) 9–21 (2016);

- ISSN 2196-7237; doi: <https://doi.org/10.1007/s40516-015-0020-5> (cited on page: 49)
- [219] M. O. Shaikh, C.-C. Chen et al.; Additive manufacturing using fine wire-based laser metal deposition; *Rapid Prototyping Journal* 26 (3) 473–483 (2019); ISSN 1355-2546; doi: <https://doi.org/10.1108/RPJ-04-2019-0110> (cited on page: 49)
- [220] A. G. Demir; Micro laser metal wire deposition for additive manufacturing of thin-walled structures; *Optics and Lasers in Engineering* 100 9–17 (2018); ISSN 0143-8166; doi: <https://doi.org/10.1016/j.optlaseng.2017.07.003> (cited on page: 49)
- [221] D. Ding, Z. Pan et al.; Wire-feed additive manufacturing of metal components: technologies, developments and future interests; *The International Journal of Advanced Manufacturing Technology* 81 (1) 465–481 (2015); ISSN 1433-3015; doi: <https://doi.org/10.1007/s00170-015-7077-3> (cited on page: 49)
- [222] N. Ghanadi, S. Pasebani; A Review on Wire-Laser Directed Energy Deposition: Parameter Control, Process Stability, and Future Research Paths (2024); doi: <https://doi.org/10.3390/jmmp8020084> (cited on pages: 49, 51, and 75)
- [223] M. Froend, S. Riekehr et al.; Process development for wire-based laser metal deposition of 5087 aluminium alloy by using fibre laser; *Journal of Manufacturing Processes* 34 721–732 (2018); ISSN 1526-6125; doi: <https://doi.org/10.1016/j.jmapro.2018.06.033> (cited on pages: 49 and 75)
- [224] M. Wang, N. Kashaev; Investigation of process window for AA7075 considering effects of different wire feed directions in lateral Laser Metal Deposition; *Procedia CIRP* 111 218–223 (2022); ISSN 2212-8271; doi: <https://doi.org/10.1016/j.procir.2022.08.053> (cited on pages: 49 and 89)
- [225] W. Williams, Martina et al.; Wire + Arc Additive Manufacturing; *Materials Science and Technology* 32 (7) 641–647 (2016); ISSN 0267-0836; doi: <https://doi.org/10.1179/1743284715Y.00000000073> (cited on page: 49)
- [226] A. Zapata, C. Bernauer et al.; Investigation on the Cause-Effect Relationships between the Process Parameters and the Resulting Geometric Properties for Wire-Based Coaxial Laser Metal Deposition (2022); doi: <https://doi.org/10.3390/met12030455> (cited on pages: 49 and 75)
- [227] S. Zhu, Y. Nakahara et al.; Optimization of Process Conditions for Additive Manufacturing Technology Combining High-Power Diode Laser and Hot Wire (2021); doi: <https://doi.org/10.3390/met11101583> (cited on pages: 49 and 75)

- [228] F. Caiazzo, V. Alfieri; Directed Energy Deposition of stainless steel wire with laser beam: evaluation of geometry and affection depth; *Procedia CIRP* 99 348–351 (2021); ISSN 2212-8271; doi: <https://doi.org/10.1016/j.procir.2021.03.121> (cited on pages: 49 and 75)
- [229] M. Wang, V. Ventzke et al.; Wire-based laser directed energy deposition of AA7075: effect of process parameters on microstructure and mechanical properties; *Journal of Materials Research and Technology* 21 388–403 (2022); ISSN 2238-7854; doi: <https://doi.org/10.1016/j.jmrt.2022.09.051> (cited on pages: 49 and 75)
- [230] X. Lu, M. V. Li et al.; Comparison of wire-arc and powder-laser additive manufacturing for IN718 superalloy: unified consideration for selecting process parameters based on volumetric energy density; *The International Journal of Advanced Manufacturing Technology* 114 (5) 1517–1531 (2021); ISSN 1433-3015; doi: <https://doi.org/10.1007/s00170-021-06990-y> (cited on page: 49)
- [231] N. Pépe, S. Egerland et al.; Measuring the process efficiency of controlled gas metal arc welding processes; *Science and Technology of Welding and Joining* 16 (5) 412–417 (2011); ISSN 1362-1718; doi: <https://doi.org/10.1179/1362171810Y.0000000029> (cited on page: 49)
- [232] B. Wu, Z. Pan et al.; A review of the wire arc additive manufacturing of metals: properties, defects and quality improvement; *Journal of Manufacturing Processes* 35 127–139 (2018); ISSN 1526-6125; doi: <https://doi.org/10.1016/j.jmapro.2018.08.001> (cited on page: 49)
- [233] G. Marinelli, F. Martina et al.; Development of Wire plus Arc additive manufacture for the production of large-scale unalloyed tungsten components; *International Journal of Refractory Metals Hard Materials* 82 329–335 (2019); ISSN 0263-4368; doi: <https://doi.org/10.1016/j.ijrmhm.2019.05.009> (cited on page: 49)
- [234] G. Marinelli, F. Martina et al.; Effect of shielding gas composition and welding speed on autogenous welds of unalloyed tungsten plates; *International Journal of Refractory Metals and Hard Materials* 85 105043 (2019); ISSN 0263-4368; doi: <https://doi.org/10.1016/j.ijrmhm.2019.105043> (cited on page: 49)
- [235] G. Marinelli, F. Martina et al.; Microstructure and thermal properties of unalloyed tungsten deposited by Wire plus Arc Additive Manufacture; *Journal of Nuclear Materials* 522 45–53 (2019); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2019.04.049> (cited on page: 49)
- [236] J. Fuchs, C. Schneider et al.; Wire-based additive manufacturing using an electron beam as heat source; *Welding in the World* 62 (2) 267–275 (2018);

- ISSN 1878-6669; doi: <https://doi.org/10.1007/s40194-017-0537-7> (cited on page: 50)
- [237] K. Osipovich, K. Kalashnikov et al.; Wire-Feed Electron Beam Additive Manufacturing: A Review (2023); doi: <https://doi.org/10.3390/met13020279> (cited on page: 50)
- [238] S. Negi, A. A. Nambolan et al.; Review on electron beam based additive manufacturing; *Rapid Prototyping Journal* 26 (3) 485–498 (2020); ISSN 1355-2546; doi: <https://doi.org/10.1108/RPJ-07-2019-0182> (cited on page: 50)
- [239] F. Pixner, R. Buzolin et al.; Wire-based electron beam additive manufacturing of tungsten; *International Journal of Refractory Metals and Hard Materials* 108 105917 (2022); ISSN 0263-4368; doi: <https://doi.org/10.1016/j.ijrmhm.2022.105917> (cited on page: 50)
- [240] M. Bambach, I. Sizova et al.; Comparison of laser metal deposition of Inconel 718 from powder, hot and cold wire; *Procedia CIRP* 74 206–209 (2018); ISSN 22128271; doi: <https://doi.org/10.1016/j.procir.2018.08.095> (cited on page: 50)
- [241] Y. Fu, A. G. Demir et al.; Additive manufacturing of Ti-6Al-4V alloy by micro-laser metal wire deposition with pulsed wave emission: processability and microstructure formation; *The International Journal of Advanced Manufacturing Technology* 126 (5) 2693–2711 (2023); ISSN 1433-3015; doi: <https://doi.org/10.1007/s00170-023-11238-y>; <https://doi.org/10.1007/s00170-023-11238-y> (cited on page: 50)
- [242] F. Brueckner, M. Riede et al.; Process characteristics in high-precision laser metal deposition using wire and powder; *Journal of Laser Applications* 29 (2) 022301 (2017); ISSN 1042-346X; doi: <https://doi.org/10.2351/1.4983237> (cited on page: 51)
- [243] N. Klingbeil; Untersuchung des Drahtsystems zur Prozessstabilisierung für das drahtbasierte Laserauftragschweißen in der additiven Fertigung; *Ap-primus Wissenschaftsverlag, Aachen* (2019); ISBN 978-3863597061 (cited on pages: 51, 53, and 54)
- [244] W. U. H. Syed, L. Li; Effects of wire feeding direction and location in multiple layer diode laser direct metal deposition; *Applied Surface Science* 248 (1) 518–524 (2005); ISSN 0169-4332; doi: <https://doi.org/10.1016/j.apsusc.2005.03.039> (cited on pages: 51 and 73)
- [245] C. Freyer; Schichtweises drahtbasiertes Laserauftragschweißen und Fräsen zum Aufbau metallischer Bauteile; *Shaker Verlag, Aachen* (2007); ISBN 978-3832259136 (cited on pages: 53 and 54)

- [246] W. Jia, S. Chen et al.; Influence of energy area densities on microstructure evolution and properties of laser direct energy deposited Ni-Co-Mn-Al-Y magnetic shape memory alloy; *Additive Manufacturing* 60 103232 (2022); ISSN 2214-8604; doi: <https://doi.org/10.1016/j.addma.2022.103232> (cited on page: 51)
- [247] M. Gipperich, J. Riepe et al.; Express Wire Coil Cladding (EW2C) As an Advanced Technology To Accelerate Additive Manufacturing and Coating; In ASME Turbo Expo 2021: Turbomachinery Technical Conference and Exposition volume Volume 7: Industrial and Cogeneration; Manufacturing Materials and Metallurgy. V007T17A001; doi: <https://doi.org/10.1115/gt2021-58709> (cited on pages: 51 and 55)
- [248] K. H. Leong, H. K. Geyer et al.; Threshold laser beam irradiances for melting and welding; *Journal of Laser Applications* 9 (5) 227–231 (1997); ISSN 1042-346X 1938-1387; doi: <https://doi.org/10.2351/1.4745464> (cited on pages: 54, 55, and 71)
- [249] M. Froend, V. Ventzke et al.; Microstructure by design: An approach of grain refinement and isotropy improvement in multi-layer wire-based laser metal deposition; *Materials Science and Engineering: A* 772 (2020); ISSN 09215093; doi: <https://doi.org/10.1016/j.msea.2019.138635> (cited on pages: 54, 55, and 88)
- [250] J. Rapp, C. Glumann et al.; Influence of Processing Parameters and Alloy Composition on the Weld Seam in Laser Welding of Aluminium Alloys; In ECLAT '92 p. pp. 99–104. DGM-Informationsgesellschaft (cited on page: 55)
- [251] P. Majerus, R. Duwe et al.; The new electron beam test facility JUDITH II for high heat flux experiments on plasma facing components; *Fusion Engineering and Design* 75-79 365–369 (2005); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2005.06.058> (cited on pages: 58, 59, and 60)
- [252] S. Schiller, U. Heisig et al.; Elektronenstrahltechnologie; Wissenschaftliche Verlagsgesellschaft Stuttgart, Lizenz des Verlag Technik, VEB Berlin (1977); ISBN 9783804705333 (cited on page: 59)
- [253] S. Flügge; Handbuch der Physik, Band XXI; Springer Verlag (1956) (cited on page: 59)
- [254] M. v. Ardenne; Tabellen zur angewandten Physik volume 1; Deutscher Verlag der Wissenschaften 2nd edition (1973) (cited on page: 59)
- [255] S. Keusemann; Feasibility study for the simulation of transient thermal events on plasma facing materials with electron beams; Thesis (2008) (cited on page: 59)

- [256] T. Loewenhoff, T. Hirai et al.; Experimental simulation of Edge Localised Modes using focused electron beams – features of a circular load pattern; *Journal of Nuclear Materials* 415 (1, Supplement) S51–S54 (2011); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2010.08.065> (cited on pages: 60 and 111)
- [257] A. Schmidt, T. Hirai et al.; First demonstration of non-destructive tests on tungsten-coated JET divertor CFC tiles in the electron beam facility JUDITH-2; *Physica Scripta* 2009 (T138) 014034 (2009); ISSN 1402-4896 0031-8949; doi: <https://dx.doi.org/10.1088/0031-8949/2009/T138/014034> (cited on page: 60)
- [258] F. Lieneweg; Handbuch der technischen Temperaturmessung; Vieweg, Braunschweig (1976); ISBN 9783528083588 (cited on pages: 60 and 61)
- [259] F. Bernhard; Handbuch der Technischen Temperaturmessung; Springer Berlin Heidelberg, Berlin, Heidelberg (2014) (cited on page: 60)
- [260] Z. M. M. Zhang, B. K. Tsai et al.; Experimental Methods in the Physical Sciences – Radiometric Temperature Measurements: I. Fundamentals volume 42; Academic Press, Amsterdam, Boston (2009); ISBN 9780123740212 (cited on page: 61)
- [261] W. Zhou, L. Zhong; Scanning Microscopy for Nanotechnology; Springer New York, New York (2007); ISBN 978-0-387-39620-0; doi: <https://doi.org/10.1007/978-0-387-39620-0> (cited on page: 63)
- [262] H. Seiler; Secondary electron emission in the scanning electron microscope; *Journal of Applied Physics* 54 (11) R1–R18 (1983); ISSN 0021-8979; doi: <https://doi.org/10.1063/1.332840> (cited on page: 63)
- [263] J. I. Goldstein, D. E. Newbury et al.; Backscattered Electrons pp. 15–28; Springer New York, New York, NY (2018); ISBN 978-1-4939-6676-9; doi: https://doi.org/10.1007/978-1-4939-6676-9_2 (cited on pages: 63 and 64)
- [264] M. Scimeca, S. Bischetti et al.; Energy Dispersive X-ray (EDX) microanalysis: A powerful tool in biomedical research and diagnosis; *Eur J Histochem* 62 (1) 2841 (2018); doi: <https://doi.org/10.4081/ejh.2018.2841> (cited on page: 64)
- [265] D. Shindo, T. Oikawa; Energy Dispersive X-ray Spectroscopy pp. 81–102; Springer Japan, Tokyo (2002); ISBN 978-4-431-66988-3; doi: https://doi.org/10.1007/978-4-431-66988-3_4 (cited on page: 64)
- [266] E. S. Gadelmawla, M. M. Koura et al.; Roughness parameters; *Journal of Materials Processing Technology* 123 (1) 133–145 (2002); ISSN 0924-0136; doi: [https://doi.org/10.1016/S0924-0136\(02\)00060-2](https://doi.org/10.1016/S0924-0136(02)00060-2) (cited on page: 64)

- [267] DIN Deutsches Institut für Normung e.V.; Metallic materials - Vickers hardness test - Part 1: Test method (ISO 6507-1:2018) (2018); doi: <https://dx.doi.org/10.31030/2778746> (cited on page: 65)
- [268] D. Nguyen; Finite Element Methods; Springer Cham 2nd edition (2024); ISBN 978-3-031-48788-0; doi: <https://doi.org/10.1007/978-3-031-48788-0> (cited on page: 65)
- [269] D. Hahn, M. Özişik; Heat Conduction Fundamentals pp. 1–39; John Wiley Sons, Inc. (2012); ISBN 9781118411285; doi: <https://doi.org/10.1002/9781118411285.ch1> (cited on page: 65)
- [270] S. Hall; 12 - Closed Loop Heat Transfer Systems pp. 203–219; Butterworth-Heinemann, Oxford (2012); ISBN 978-0-12-387785-7; doi: <https://doi.org/10.1016/B978-0-12-387785-7.00012-8> (cited on pages: 66 and 67)
- [271] E. N. Sieder, G. E. Tate; Heat transfer and pressure drop of liquids in tubes; *Industrial Engineering Chemistry* 28 (12) 1429–1435 (1936); ISSN 0019-7866 (cited on page: 66)
- [272] J. Romazanov; 3D-simulation of impurity transport in a fusion edge plasma by means of a massively parallelized Monte-Carlo code; Thesis (2018); doi: <https://doi.org/10.13154/294-6330> (cited on page: 67)
- [273] W. W. Webb, J. T. Norton et al.; Oxidation of Tungsten; *Journal of The Electrochemical Society* 103 (2) 107 (1956); ISSN 1945-7111 0013-4651; doi: <https://dx.doi.org/10.1149/1.2430238> (cited on page: 70)
- [274] W. Stopyra, K. Gruber et al.; Laser powder bed fusion of AA7075 alloy: Influence of process parameters on porosity and hot cracking; *Additive Manufacturing* 35 101270 (2020); ISSN 2214-8604; doi: <https://doi.org/10.1016/j.addma.2020.101270> (cited on page: 73)
- [275] P. Tolias; Analytical expressions for thermophysical properties of solid and liquid tungsten relevant for fusion applications; *Nuclear Materials and Energy* 13 42–57 (2017); ISSN 23521791; doi: <https://doi.org/10.1016/j.nme.2017.08.002> (cited on pages: 74 and 130)
- [276] P. F. Paradis, T. Ishikawa et al.; Physical properties of liquid and undercooled tungsten by levitation techniques; *Applied Physics Letters* 86 (4) (2005); ISSN 0003-6951; doi: <https://doi.org/10.1063/1.1853513> (cited on page: 74)
- [277] J. V. Gordon, S. P. Narra et al.; Defect structure process maps for laser powder bed fusion additive manufacturing; *Additive Manufacturing* 36 (2020); ISSN 22148604; doi: <https://doi.org/10.1016/j.addma.2020.101552> (cited on page: 74)

- [278] T. Bergs, S. Kammann et al.; Experimental investigations on the influence of temperature for Laser Metal Deposition with lateral Inconel 718 wire feeding; *Procedia CIRP* 94 29–34 (2020); ISSN 22128271; doi: <https://doi.org/10.1016/j.procir.2020.09.007> (cited on page: 75)
- [279] M. Froend, V. Ventzke et al.; Thermal analysis of wire-based direct energy deposition of Al-Mg using different laser irradiances; *Additive Manufacturing* 29 100800 (2019); ISSN 2214-8604; doi: <https://doi.org/10.1016/j.addma.2019.100800> (cited on page: 75)
- [280] Y. Chen, F. Lu et al.; Investigation of dendritic growth and liquation cracking in laser melting deposited Inconel 718 at different laser input angles; *Materials Design* 105 133–141 (2016); ISSN 02641275; doi: <https://doi.org/10.1016/j.matdes.2016.05.034> (cited on page: 88)
- [281] M. S. F. de Lima, S. Sankaré; Microstructure and mechanical behavior of laser additive manufactured AISI 316 stainless steel stringers; *Materials Design* 55 526–532 (2014); ISSN 02613069; doi: <https://doi.org/10.1016/j.matdes.2013.10.016> (cited on page: 88)
- [282] S.-H. Pan, G.-C. Yao et al.; Additive manufacturing of tungsten, tungsten-based alloys, and tungsten matrix composites; *Tungsten* 5 (1) 1–31 (2022); ISSN 2661-8028 2661-8036; doi: <https://doi.org/10.1007/s42864-022-00153-6> (cited on pages: 89, 91, and 145)
- [283] F. Pixner, R. Buzolin et al.; Wire-based electron beam additive manufacturing of tungsten; *International Journal of Refractory Metals Hard Materials* 108 (2022); ISSN 0263-4368; doi: <https://doi.org/10.1016/j.ijrmhm.2022.105917> (cited on pages: 89, 92, and 146)
- [284] T. Hirai, F. Escourbiac et al.; ITER tungsten divertor design development and qualification program; *Fusion Engineering and Design* 88 (9) 1798–1801 (2013); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2013.05.010> (cited on pages: 95 and 96)
- [285] M. Li, J.-H. You; Interpretation of the deep cracking phenomenon of tungsten monoblock targets observed in high-heat-flux fatigue tests at 20 MW/m²; *Fusion Engineering and Design* 101 1–8 (2015); ISSN 09203796; doi: <https://doi.org/10.1016/j.fusengdes.2015.09.008> (cited on pages: 96 and 97)
- [286] J. H. You, E. Visca et al.; European divertor target concepts for DEMO: Design rationales and high heat flux performance; *Nuclear Materials and Energy* 16 1–11 (2018); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2018.05.012> (cited on pages: 96 and 97)

- [287] S. Panayotis, T. Hirai et al.; Self-castellation of tungsten monoblock under high heat flux loading and impact of material properties; *Nuclear Materials and Energy* 12 200–204 (2017); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2016.10.025> (cited on pages: 96 and 97)
- [288] Z. Sun, Q. Li et al.; Post examination of tungsten monoblocks subjected to high heat flux tests of ITER full-tungsten divertor qualification program; *Fusion Engineering and Design* 121 60–69 (2017); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2017.06.009> (cited on pages: 116 and 123)
- [289] S. Panayotis, T. Hirai et al.; Fracture modes of ITER tungsten divertor monoblock under stationary thermal loads; *Fusion Engineering and Design* 125 256–262 (2017); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2017.08.009> (cited on pages: 116 and 123)
- [290] G. Pintsuk, M. Bednarek et al.; Characterization of ITER tungsten qualification mock-ups exposed to high cyclic thermal loads; *Fusion Engineering and Design* 98-99 1384–1388 (2015); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2015.01.037> (cited on pages: 116, 120, and 123)
- [291] T. Hirai, F. Escourbiac et al.; Status of technology RD for the ITER tungsten divertor monoblock; *Journal of Nuclear Materials* 463 1248–1251 (2015); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2014.12.027> (cited on pages: 116 and 123)
- [292] P. Gavila, B. Riccardi et al.; High heat flux testing of EU tungsten monoblock mock-ups for the ITER divertor; *Fusion Engineering and Design* 98-99 1305–1309 (2015); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2014.12.006> (cited on pages: 116 and 123)
- [293] G. Pintsuk, I. Bobin-Vastra et al.; Qualification and post-mortem characterization of tungsten mock-ups exposed to cyclic high heat flux loading; *Fusion Engineering and Design* 88 (9) 1858–1861 (2013); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2013.05.091> (cited on pages: 116, 120, and 123)
- [294] M. Wirtz, J. Linke et al.; Thermal shock tests to qualify different tungsten grades as plasma facing material; *Physica Scripta* T167 (2016); ISSN 0031-8949 1402-4896; doi: <https://doi.org/10.1088/0031-8949/t167/1/014015> (cited on page: 116)

- [295] A. Kapoustina, T. Hirai et al.; Detection of Heat Transfer Properties of CFC-Plasma Facing Components by IR-Observation and High Heat Flux Testing; *Physica Scripta* 2004 (T111) 181 (2004); ISSN 1402-4896 0031-8949; doi: <https://dx.doi.org/10.1238/Physica.Topical.111a00181> (cited on page: 117)
- [296] G. Pintsuk, E. Cacciotti et al.; Long-pulse high heat flux testing of tungsten monoblock target mock-ups for investigation of creep fatigue interaction; *Nuclear Materials and Energy* 39 101687 (2024); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2024.101687> (cited on page: 118)
- [297] T. Loewenhoff, S. Antusch et al.; High pulse number thermal shock testing of tungsten alloys produced by powder injection molding; *Nuclear Materials and Energy* 20 100680 (2019); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2019.100680> (cited on pages: 119, 124, 128, 139, and 141)
- [298] J. H. You, H. Greuner et al.; High-heat-flux performance limit of tungsten monoblock targets: Impact on the armor materials and implications for power exhaust capacity; *Nuclear Materials and Energy* 33 101307 (2022); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2022.101307> (cited on page: 120)
- [299] Y. Bournonville, M. Grandotto et al.; Numerical simulation of swirl-tube cooling concept, application to the ITER project; *Fusion Engineering and Design* 84 (2) 501–504 (2009); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2008.11.028> (cited on page: 121)
- [300] K. Krieger, S. Brezinsek et al.; Scrape-off layer and divertor physics: Chapter 5 of the special issue: on the path to tokamak burning plasma operation; *Nuclear Fusion* 65 (4) 043001 (2025); ISSN 0029-5515; doi: <https://dx.doi.org/10.1088/1741-4326/adaf42> (cited on page: 122)
- [301] T. Loewenhoff, S. Bardin et al.; Impact of combined transient plasma/heat loads on tungsten performance below and above recrystallization temperature; *Nuclear Fusion* 55 (12) (2015); ISSN 0029-5515 1741-4326; doi: <https://doi.org/10.1088/0029-5515/55/12/123004> (cited on pages: 125 and 128)
- [302] G. Pintsuk, A. Prokhodtseva et al.; Thermal shock characterization of tungsten deformed in two orthogonal directions; *Journal of Nuclear Materials* 417 (1-3) 481–486 (2011); ISSN 00223115; doi: <https://doi.org/10.1016/j.jnucmat.2010.12.109> (cited on page: 128)
- [303] T. Loewenhoff, J. Linke et al.; Tungsten and CFC degradation under combined high cycle transient and steady state heat loads; *Fu-*

- sion Engineering and Design* 87 (7-8) 1201–1205 (2012); ISSN 09203796; doi: <https://doi.org/10.1016/j.fusengdes.2012.02.106> (cited on page: 128)
- [304] D. Dorow-Gerspach, T. Derra et al.; Progress in the Realization of μ -Brush W for Plasma-Facing Components; *Journal of Nuclear Engineering* 3 (4) 333–341 (2022); ISSN 2673-4362; doi: <https://doi.org/10.3390/jne3040020> (cited on page: 128)
- [305] T. Loewenhoff, J. Linke et al.; Laser re-melting of tungsten damaged by transient heat loads; *Nuclear Materials and Energy* 9 165–170 (2016); ISSN 2352-1791; doi: <https://doi.org/10.1016/j.nme.2016.04.004> (cited on page: 139)
- [306] A. Fathi, E. Toyserkani et al.; Prediction of melt pool depth and dilution in laser powder deposition; *Journal of Physics D-Applied Physics* 39 (12) 2613–2623 (2006); ISSN 0022-3727; doi: <https://doi.org/10.1088/0022-3727/39/12/022> (cited on page: 139)
- [307] G. Q. Yu, D. D. Gu et al.; Influence of processing parameters on laser penetration depth and melting/re-melting densification during selective laser melting of aluminum alloy; *Applied Physics a-Materials Science Processing* 122 (10) (2016); ISSN 0947-8396; doi: <https://doi.org/10.1007/s00339-016-0428-6> (cited on page: 139)
- [308] D. J. Ward, I. Cook et al.; The economic viability of fusion power; *Fusion Engineering and Design* 75–79 1221–1227 (2005); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2005.06.160> (cited on page: 143)
- [309] H. Bolt, V. Barabash et al.; Materials for the plasma-facing components of fusion reactors; *Journal of Nuclear Materials* 329–333 66–73 (2004); ISSN 0022-3115; doi: <https://doi.org/10.1016/j.jnucmat.2004.04.005> (cited on page: 143)
- [310] D. Maisonnier, I. Cook et al.; DEMO and fusion power plant conceptual studies in Europe; *Fusion Engineering and Design* 81 (8) 1123–1130 (2006); ISSN 0920-3796; doi: <https://doi.org/10.1016/j.fusengdes.2005.08.055> (cited on page: 143)
- [311] J. Winter, H. G. Esser et al.; Boronization in textor; *Journal of Nuclear Materials* 162–164 713–723 (1989); ISSN 0022-3115; doi: [https://doi.org/10.1016/0022-3115\(89\)90352-8](https://doi.org/10.1016/0022-3115(89)90352-8) (cited on page: 146)
- [312] S. Sereda, S. Brezinsek et al.; Impact of boronizations on impurity sources and performance in Wendelstein 7-X; *Nuclear Fusion* 60 (8) 086007 (2020); ISSN 0029-5515; doi: <https://dx.doi.org/10.1088/1741-4326/ab937b> (cited on page: 146)

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