

In situ study of vacancy disordering in crystalline phase-change materials under electron beam irradiation

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Abstract:

Unconventionally high amount of atomic vacancies up to more than 10% are known to form in Ge-Sb-Te crystals upon rapid crystallization from the amorphous phase. Upon thermal annealing, an ordering process of these atomic vacancies is observed, triggering a structural transition from the recrystallized rocksalt structure to a stable layered trigonal structure and a transition from insulator to metal. In this work, we demonstrate an opposite vacancy *disordering* process upon extensive electron beam irradiation, which is accompanied by the reverse transition from the stable trigonal phase to the metastable cubic phase. The combined in situ transmission electron microscopy experiments and density functional theory nudged elastic band calculations reveal three transition stages, including (I) the vacancy diffusion in the trigonal phase, (II) the change in atomic stacking, and (III) the disappearance of vacancy-rich planes. The mechanism of vacancy disordering is attributed to kinetic knock-on collision effects of the high-energy electron beam, which prevail over the heating effects.

Keywords: phase-change materials, GeSbTe, vacancy disordering, electron beam irradiation, in situ TEM

Introduction

Fast and reversible phase transitions in chalcogenide phase-change materials (PCMs) play a key role in next-generation memory and computing chips [1-20]. Ge-Sb-Te compounds, such as $\text{Ge}_2\text{Sb}_2\text{Te}_5$ and GeSb_2Te_4 , are under active investigation [21-28], not only because they serve as the key element in commercial products [29], but also because they provide a rich platform for fundamental research in materials science, such as disorder-induced phenomena and metal-insulator transitions [30-40]. Upon rapid crystallization from the amorphous phase, Ge-Sb-Te compounds form a cubic rocksalt (cub-) phase [36, 41-44], consisting of two sublattices, one of which contains Te atoms, while the other is occupied by Ge atoms, Sb atoms and atomic vacancies in a random fashion. A trigonal (tri-) phase is formed upon further thermal annealing [45-50], which consists of stacking blocks. These blocks consist of alternately stacked Ge/Sb and Te layers and are terminated by Te layers [51-54].

The cub-phase of Ge-Sb-Te compounds contains a huge amount of atomic vacancies, e.g. 12.5 % for GeSb_2Te_4 : such vacancy concentrations are orders of magnitude higher than those found in conventional semiconductors [55, 56]. The formation of vacancy clusters results in Anderson localization of electrons [57], making cub-Ge-Sb-Te compounds disorder-induced Anderson insulators [30, 31]. Note that, during fast crystallization, a polycrystalline phase is typically formed. Nevertheless, the grain boundary scattering has been shown to affect the transport properties only marginally [58]. The ordering of vacancies into two-dimensional layers upon thermal annealing leads to both a structural transition into the tri-phase and an insulator-to-metal transition [30, 31]. Although the two transitions occur in a relatively narrow range of temperature, they are fundamentally different: in particular, the insulator-to-metal transition is an Anderson transition primarily driven by a reduction in intra-grain lattice disorder, not by a change in structure [37, 38].

The vacancy ordering process in Ge-Sb-Te crystals was firstly predicted by density functional theory (DFT) calculations [31], and was later confirmed by transmission electron microscopy (TEM) experiments upon thermal annealing [36, 37] and electron beam irradiation [44]. In this work, we report on in situ TEM observations of a vacancy disordering induced metastabilization from the tri-phase to the cub-phase in crystalline GeSb_2Te_4 (GST) thin films upon extensive electron beam irradiation. Structural details are recorded and analyzed along the transition path, and are further compared with DFT calculations, providing a comprehensive view of vacancy-diffusion induced structural metastabilization in crystalline GST thin films.

Experimental and simulation details

GeSb_2Te_4 films of ~80 nm thickness were deposited with magnetron sputtering on ultra-thin carbon film (~5 nm) TEM grids and were covered by an electron-transparent ZnS-SiO_2 layer. The as-deposited samples were annealed in argon atmosphere (flow rates of 1200 sccm) in a regular tube furnace at 150 °C for 1 hour and 300 °C for 1 hour to obtain cubic and trigonal phases, respectively. Both sets of thin films were found to be poly-crystalline with grains of different

crystallographic orientations along the view direction. The in situ electron beam irradiation experiments were performed based on the trigonal sample in a JEOL JEM-2100F TEM operated at 200 kV. The microstructures were characterized by bright-field and high resolution TEM (HRTEM) imaging, selected area electron diffraction (SAED) analysis and fast Fourier transform (FFT) of HRTEM images.

DFT calculations were performed using the Perdew-Burke-Ernzerhof functionals [59] and the projected augmented plane-wave (PAW) pseudopotentials [60] as implemented in the Vienna *ab initio* simulation package (VASP) [61]. The cut-off energy for the plane-wave basis expansion was set to be 400 eV. Self-consistent calculations were carried out with a Gaussian smearing width of 0.05 eV and a convergence tolerance of 1×10^{-6} eV. K-point grids for the Brillouin zone integration were chosen as $3 \times 3 \times 1$. The initial trigonal model contains 21 atomic layers with 12 Ge, 24 Sb and 48 Te atoms. The nudged elastic band (NEB) calculations were performed using the climbing-image NEB (CI-NEB) implementation [62, 63] in VASP. The relaxation convergence tolerance for CI-NEB was set to be 0.03 eV/Å. In the NEB study, the lattice parameters of the supercell were fixed as $L_x = L_y = 2 \times 4.27$ Å and $L_z = 41.87$ Å, corresponding to the experimental values of the cub-phase, which are very close to that of the trigonal phase with $a_{tri} = 4.27$ Å and $c_{tri} = 41.69$ Å [42]. A denser k point mesh of $6 \times 6 \times 2$ was tested for a typical diffusion calculation, yielding almost identical results (Figure S1). The van der Waals correction was also tested [64] and shown to slightly enlarge the relevant energy barriers by $\sim 0.05 - 0.1$ eV (Figure S1).

Results and Discussion

The TEM images of thermally-annealed samples and the atomic models of the tri- and cub-phase are presented in Figure 1. For the tri-phase (Figure 1a), most grains show a needle-like morphology (red box). The HRTEM images and corresponding FFT patterns of selected areas show the structural details of the typical $[01\bar{1}0]_t$ - and $[0001]_t$ -oriented grains. The yellow arrows indicate structural gaps, which are usually referred to as van der Waals (vdW) gaps in literature, although it has been shown recently that weak covalent interactions are present in addition to van der Waals forces [65-67]. For the cub-phase (Figure 1b), the grains look equiaxial (red box). Two typical crystallographic orientations $[100]_c$ and $[111]_c$ are highlighted. The measured plane spacings are in line with previous measurements on the two crystalline phases of GST [41, 42]. It is important to note that the atomic arrangement and lattice parameters in the $(0001)_t$ plane of the tri-phase and the $(111)_c$ plane of the cub-phase are identical [31], as evidenced by the HRTEM images and the corresponding diffraction patterns (yellow box).

Figure 1c shows the side view of a $2 \times 2 \times 1$ tri-phase model, containing three septuple-layer (SL) blocks and three structural gaps. The compositional Ge/Sb disorder is considered for the cation-like layers with 50% Ge and 50% Sb for the center layer, 25% Ge and 75% Sb for the two outer layers [42]. Each atomic layer occupies one of the three special positions, i.e. “a” = (0, 0), “b” = (2/3, 1/3) and “c” = (1/3, 2/3), of the unit cell (Figure 1d). Figure 1e shows a standard cubic supercell, which

can also be built along its [111] direction (Figure 1f), giving very similar structural features as the tri-phase [68]. The major difference comes from the statistical distribution of atomic vacancies on the cation-like layers in the cub-phase. If all the atomic vacancies are arranged into three particular layers, an ordered phase with cubic stacking (o-cub) can be obtained (Figure 1g). The stacking sequence of the tri-phase is “-g-abcabca-g-bcabcab-g-cabcab-” (where “g” represents the structural gap), and can be compactly denoted as “-g-A-g-B-g-C-” (Figure 1c), whereas the o-cub-phase consists of “-g-A-g-C-g-B-” blocks (Figure 1g).

If the electron beam was focused to a small area (~ 200 nm in radius) with a high beam intensity of $\sim 1.1 \times 10^{24} \text{ e m}^{-2} \text{ s}^{-1}$ (current density $\sim 1.76 \times 10^{13} \text{ pA cm}^{-2}$), a gradual structural transition from the tri-phase to the cub-phase was observed as a function of time (Figure 2). The tri-phase grain in the center of the irradiation area is indexed to be [0001]-oriented from the SAED pattern, which appeared to be darker as compared to the surrounding grains (Figure 2a). After 45 min irradiation, a clear change of the image contrast of the center tri-phase grain was found, and extra spots appeared in the corresponding SAED pattern, such as $[\bar{1}2\bar{1}0]_t$ and $[\bar{1}016]_t$ (marked by the yellow arrows in Figure 2b), indicating that the lattice re-oriented in parts of the central grain. Small lattice distortions were also observed, which could be attributed to the effects of strain during the re-orientation process. Further exposure to the electron beam triggered a structural transition to the cub-phase in parts of the irradiated area, as evidenced by the cub-phase spots (marked by white circles) in the SAED pattern in Figure 2c. After 115 min irradiation, the irradiated area was comprised of small cub-phase grains with multiple lattice orientations (Figure 2d). The structural features of this irradiation-induced cubic sample closely resemble the thermally-annealed one.

This direct tri-to-cub structural transition path under electron beam irradiation is in contrast to the conventional thermal-induced tri-to-cub path, which consists of a melt-quenched amorphization and then a rapid crystallization process into the cub-phase. The in situ TEM experiments revealed a progressive and continuous structural metastabilization process from the tri-phase to the cub-phase. To assess the structural changes in more detail, we carried out in situ HRTEM characterizations (Figure 3) of a target tri-phase grain showing the $[010\bar{1}03]_t$ -orientation, as evidenced by its FFT pattern (Figure 3a). Under exposure to electron beam irradiation with the same beam intensity of $\sim 1.1 \times 10^{24} \text{ e m}^{-2} \text{ s}^{-1}$ over 45 min, the target trigonal grain splits into two parts with different lattice orientations (Figure 3b). After 85 min irradiation, a phase transition occurred and two cubic phase grains with different orientations appeared (Figure 3c). At this stage, the structural gaps in the tri-grain with blue boundary were still visible, while they were gradually filled and the tri-grain transformed to a cub-phase grain with multiple intersecting vacancy-rich $(\bar{1}\bar{1}1)_c$ and $(1\bar{1}1)_c$ layers after 100 min irradiation (Figure 3d). Such intersecting vacancy-rich layers were also observed in cub-phase GST thin films under thermal annealing [36].

To make further analyses on intra-grain structural details, it is advantageous to record the HRTEM images from the side view of the tri- and cub-phase during the whole transition process. We

managed to find a suitable grain for this experiment, despite that the constant change of grain orientation under electron beam irradiation. This set of in situ HRTEM recordings is presented in Figure 4. After 20 min irradiation with the same beam intensity of $\sim 1.1 \times 10^{24} \text{ e m}^{-2} \text{ s}^{-1}$, a mixture of tri-phase and cub-phase structures was observed (Figure 4a). From these images the stacking sequence of the tri-phase is well distinguished from the cub-phase, marked by yellow and red dots, respectively (Figure 4a). After 40 min irradiation, the sliding of SL blocks took place (Figure 4b). Further irradiation could drive the filling of the vacancy-rich $(\bar{1}\bar{1}1)_c$ layers and/or the equivalent $(1\bar{1}1)_c$ layers in the cub-phase structures (Figure 4c). These vacancy-rich layers were less visible after 100 min irradiation (Figure 4d), and vanished completely after 130 min irradiation (Figure 4e). Clearly, this tri-to-cub transition corresponds to a vacancy disordering process, in contrast to the vacancy ordering process upon thermal annealing [31, 36].

Next, we performed DFT NEB calculations to study this vacancy disordering induced structural metastabilization process in detail. Within this method, the minimum energy path between known initial and final states is determined by relaxing simultaneously a chain of configurations – denoted as images – connected by springs, which interpolate between the two states. Furthermore, the highest-energy image is made to climb uphill to the saddle point. Starting from a $2 \times 2 \times 1$ tri-phase model, one Sb atom in an outer “cation” layer was moved into the gap region (Figure 5a). The Sb atom firstly broke six Sb-Te bonds, diffused across a Te layer and then formed six new Sb-Te bonds when it arrived in the gap region, raising the total energy of the system by ~ 1.0 eV. Three such transitions were calculated in a sequential way, and the barrier for each transition was calculated to be ~ 1.5 eV (Figure 5b), higher than the Sb diffusion barrier (~ 1.0 eV) in cub-phase of GST [36, 52]. These high energy barriers for diffusion are important to retain the respective crystalline phases, otherwise, structural transition may already take place spontaneously at low temperatures. The out-of-plane diffusion process created one atomic vacancy in the SL block and turned the structural gap into a vacancy-rich layer. Following our previous definition of vacancy concentration on three target layers I_{vac} , this process can be termed as a tri-100% $\rightarrow$ tri-75% transition, representing the Stage I of the whole vacancy disordering process. For the sake of convenience, we abbreviated the three target vacancy-rich layers as VLs.

According to our previous total energy calculations for large-scale GST models (consisting of 1008 atoms in a $\sim 3.0 \times 2.6 \times 4.2 \text{ nm}^3$ supercell) [31], for the system with $I_{vac}=75\%$ the tri-phase stacking is still energetically more favorable than the cubic one, but if I_{vac} lowers to 50%, the cub-phase stacking has a smaller total energy, suggesting a block shifting transition associated with vacancy diffusion [31]. Our small-scale models show the same trends in total energy and enable a direct transition calculation to understand the atomic details involved in this process. To reduce the I_{vac} and obtain the cub-phase stacking, atomic diffusion into vacancy gaps should continue and the shifting of the SL blocks should occur. This tri-75% $\rightarrow$ cub-50% transition process represents the Stage II of the whole vacancy disordering process.

A typical transition path of Stage II is presented in Figure 6a: the path can be characterized by three sub-processes, namely, (1) out-of-plane “cation” diffusion, (2) in-plane “cation” diffusion and (3) SL block shifting. The first two peaks in the NEB energy profile (Figure 6b) correspond to three out-of-plane diffusions and the other peaks correspond to six in-plane diffusions in the three VLs. The barrier for in-plane diffusion is on average ~ 0.3 eV, which is much smaller than the out-of-plane diffusion barrier of 1.2–1.6 eV. The top two SL blocks shifted continuously during the transition process. The shifting pathway was chosen to be a “snake-like” path (Figure 6c), i.e. the middle SL block shifted from B position to C position, and the top SL block shifted from C position to another B position. No energy barrier was observed for the continuous block shifting process. If two SL blocks moved towards each other and swapped their position, which is known as an “overhead” path [69, 70], a much higher energy barrier would appear. This favorable “snake-like” path is confirmed by a simple block shifting calculation presented in Figure S2.

As shown in Figure 4, the vacancy disordering process can continue upon extensive electron beam irradiation until vacancy-rich layers vanish, indicative of further atomic diffusion into the vacancy-rich layers. Once this process is completed, the vacancy concentration of all “cation” layers gets close to 25%, and the vacancy-rich layers are no longer distinguishable from other “cation” layers. This cub-50% $\rightarrow$ cub-25% transition process represents Stage III of the whole vacancy disordering process. Similar to Stage I, the transition is mainly dominated by out-of-plane atomic diffusion with a barrier of ~ 1.7 eV (Figure 7). Overall, the transition barriers during the whole tri-to-cub structural transition are not very large, and can be triggered by extensive exposure to high-energy electron beams.

In contrast to the vacancy ordering induced structural stabilization process upon thermal annealing, the irradiation-induced structural metastabilization is driven by the kinetic collision effects of electron beams [71], which generate displacement forces to trigger vacancy diffusion. As noted above, to generate tri-to-cub transition using pure thermal effects, the thin films should undergo firstly a melt-quenched amorphization and then recrystallization into the cub-phase. But clearly, in the irradiation experiments, the change in TEM images and corresponding diffraction was continuous, and no sign of abrupt melting was observed.

Starting from the cub-phase, if the thin films were exposed further to the electron beams, a progressive and non-thermal amorphization process would eventually occur [72]. The higher the accelerating voltage, the stronger the kinetic effects of the electron beam, thereby the faster the amorphization transitions [72]. As indicated by Raman spectroscopy monitoring, similar structural metastabilization from the tri-phase to the cub-phase and finally to the amorphous phase were also observed in GST thin films under Ar^+ ions bombardment [40, 73], in which the kinetic effects of ion beams were stronger than those of electron beams. The vacancy ordering and disordering process in GST crystals is schematically summarized in Figure 8.

In addition to kinetic effects, electron beams could also induce specimen heating [74-76] and radiolysis effects [71]. It was shown that amorphous Ge-Sb-Te and other PCMs thin films could be crystallized using electron beams, where the specimen heating played an important role [77-80]. The heating effects could get stronger at a lower accelerating voltage [71]. However, the estimated temperature rise here is well below the melting point of GST ($T_m \sim 650$ °C), which is insufficient to trigger melt-quenching amorphization and then recrystallization into the cub-phase. Indeed, no abrupt structural change was observed during in situ recording of the irradiation-induced metastabilization process, indicating a continuous and non-thermal structural transition from the tri-phase to the cub-phase, in stark contrast with the thermal-induced transitions in GST.

Besides, electron beam irradiation was also shown to tune the size of atomic blocks via vacancy disordering in GeTe-Sb₂Te₃ superlattices [81], or to induce the motion of swapped bilayers, and, thereby, dynamical reconfiguration of structural gaps, in layer-structured Ge-Sb-Te and related materials[82-85]. Due to the multiple effects of electron beams, we therefore note that for the measurement of structural details of a particular solid-state phase of PCMs, the three critical TEM parameters, namely, accelerating voltage, beam intensity and recording time, should be reduced as much as possible to avoid structural transitions during the measurement. However, in order to study the dynamical process of structural transitions in PCMs, these parameters can be tuned at higher values accordingly.

Conclusion

In summary, we have demonstrated a progressive vacancy disordering process that induced a structural transition from the tri-phase to the cub-phase in GST thin films under extensive electron beam irradiation. The in situ TEM recordings provided a comprehensive real-time and real-space view of the vacancy disordering process, while the DFT NEB calculations elucidated the atomic details involved in this transition. In contrast to heating effects that trigger vacancy ordering and structural transition to the stable tri-phase, the electron beam irradiation induces displacement forces driving a reverse metastabilization transition in crystalline GST thin films. It was demonstrated that intra-grain structural changes, in particular, the statistical distribution of atomic vacancies, play a major role in shaping the localization behavior of electrons. Our electron irradiation experiment provides an alternative approach to tune the vacancy distribution that is crucial for the disorder-driven metal-insulator transition of GST crystals.

Acknowledgements

W.Z. thanks the support of National Natural Science Foundation of China (61774123), 111 Project 2.0 (BP2018008), the Science and Technology Department of Jiangsu Province (BK20170414), and the Young Talent Support Plan and of Xi'an Jiaotong University. C.-L.J., R.M., and M.W. acknowledge funding from Deutsche Forschungsgemeinschaft within SFB 917 "Nanoswitches." The authors also acknowledge the computational resources provided by the HPC platform of Xi'an Jiaotong University. The authors also acknowledge the support by the Materials Studio for Neuro-inspired

Computing (mSonic) and the International Joint Laboratory for Micro/Nano Manufacturing and Measurement Technologies of Xi'an Jiaotong University.

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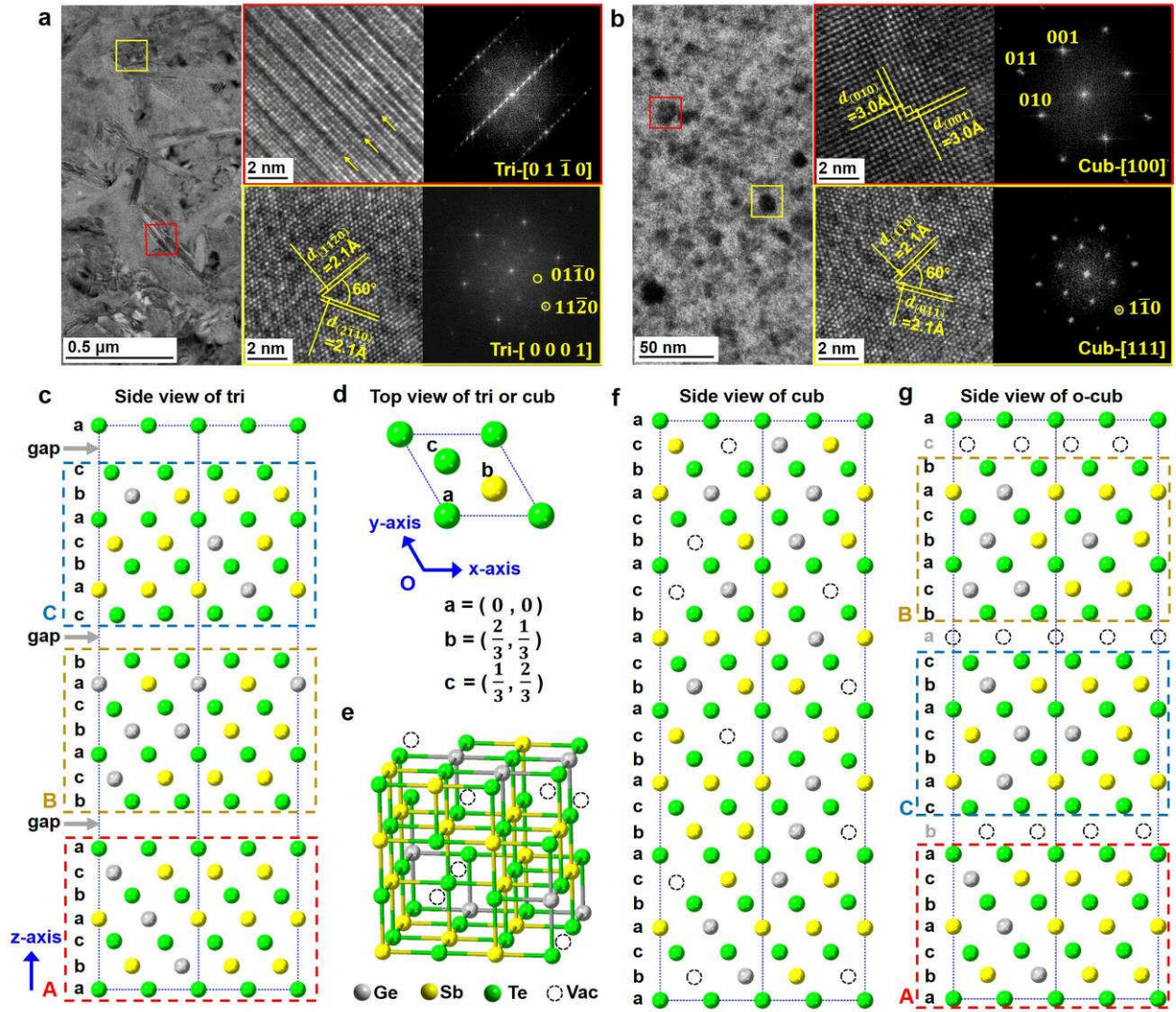


Figure 1. a-b The TEM images and the corresponding FFT of the images, showing tri-phase and cub-phase GeSb_2Te_4 . Grains with typical crystallographic orientations for the two crystalline phases are highlighted. c-g The atomic models of the two crystalline phases. c The side view of the tri-phase, showing three septuple-layer blocks with three structural gaps. Compositional disorder on the cation-like layers is considered. d The top-view of the three typical in-plane atomic arrangements. e-f The atomic models of the cub-phase built in a conventional cubic box and in an equivalent orthorhombic box. g The atomic model of an o-cub-phase with full vacancy layers.

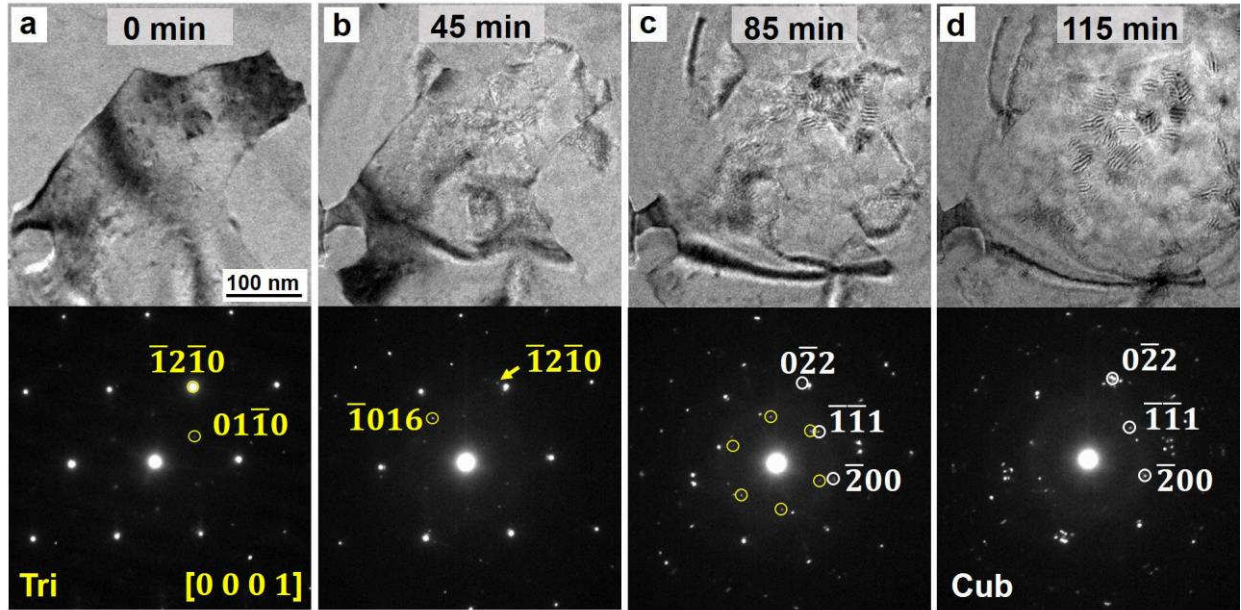


Figure 2. **a** Bright-field TEM image of a thermally-induced tri-phase grain viewed along the $[0001]_t$ direction, as confirmed by the SAED pattern. **b** The snapshot and the corresponding SAED pattern of the tri-phase grain after 45 min irradiation. Yellow arrows in the SAED pattern indicate another set of reflection spots that does not belong to the original $[0001]_t$ zone axis, revealing the re-orientation of part of the grain. The lattice is slightly distorted due to strain effects during the re-orientation process. **c** shows a transition from tri- to cub-GST. Both tri- and cub-GST reflection spots can be observed in the SAED pattern, which are marked with yellow and white circles, respectively. **d** All the tri-phase grains transformed into cub-phase.

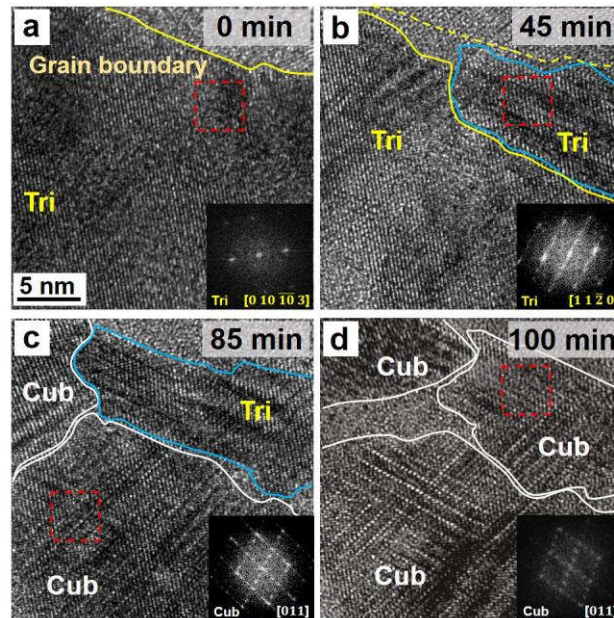


Figure 3. **a** The HRTEM image of a tri-phase grain prior to the extensive irradiation. The FFT pattern corresponding to the area inside the red dashed box indicates the $[0\ 10\ \bar{1}0\ 3]_t$ orientation of this grain. **b-d** The snapshots of the metastabilization process after 45, 85 and 100 min irradiation. The dashed yellow line in **b** marks the initial grain boundary. Under irradiation, the tri-phase grain in **a** was divided into two parts

firstly, as marked with the yellow and blue lines in **b**. In **c**, two cub-phase grains with different orientations appeared, as confirmed by the FFT pattern. **d** All tri-phase grains transformed into the cub-phase with intersecting vacancy-rich layers after 100 min irradiation.

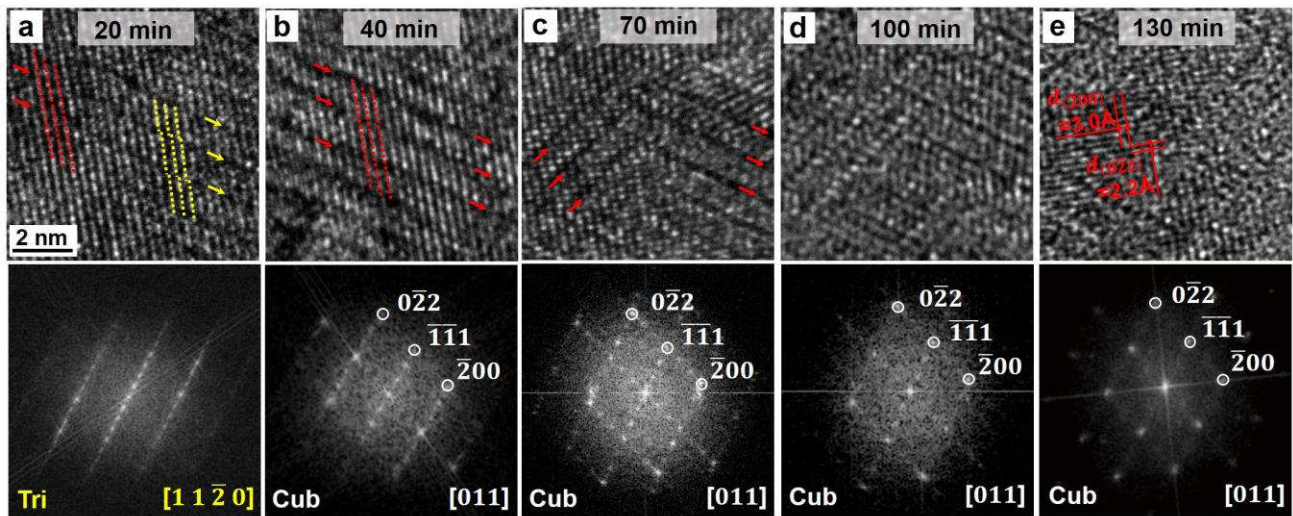


Figure 4. Series of images of a GST grain and corresponding FFT patterns at different irradiation time. **a** The GST grain with mixture of tri-phase (yellow dots) and cub-phase (red dots) stackings formed after 20 min irradiation. Structural gaps and vacancy-rich layers are marked by yellow and red arrows, respectively. **b** The shift of SL blocks occurred after 40 min irradiation. **c** Multiple $(\bar{1}\bar{1}1)_c$ and $(1\bar{1}1)_c$ vacancy-rich layers were still present after 70 min irradiation. **d** The vacancy-rich layers were mostly filled after 100 min irradiation. **e** The image of the grain after 130 min irradiation. The vacancy-rich layers fully vanished, and atomic vacancies became randomly distributed as in the conventional cub-phase.

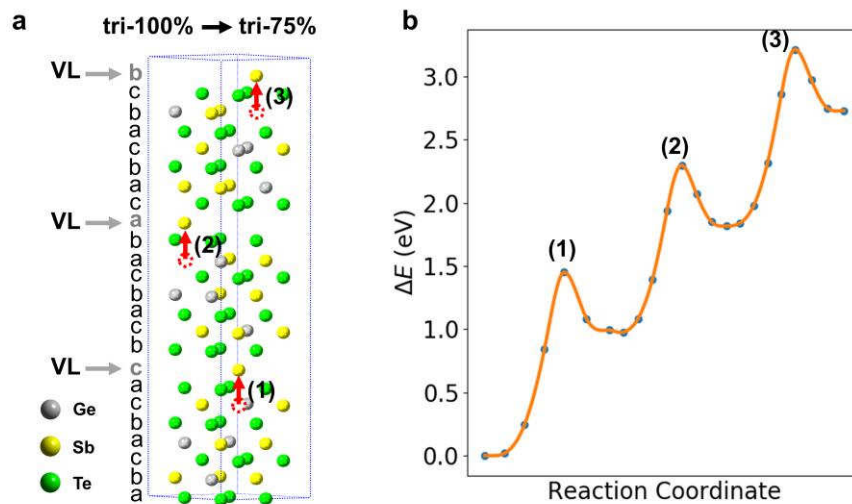


Figure 5. **a** A typical transition path from tri-100% to tri-75% with three Sb atoms moved into their adjacent structural gap regions, respectively, forming three VLs. Dotted circles represent the original positions of the Sb atoms. **b** The NEB energy profile of the three atomic diffusion processes described in **a**. Each process raised the total energy of the system by ~ 1 eV, with an energy barrier of ~ 1.5 eV.

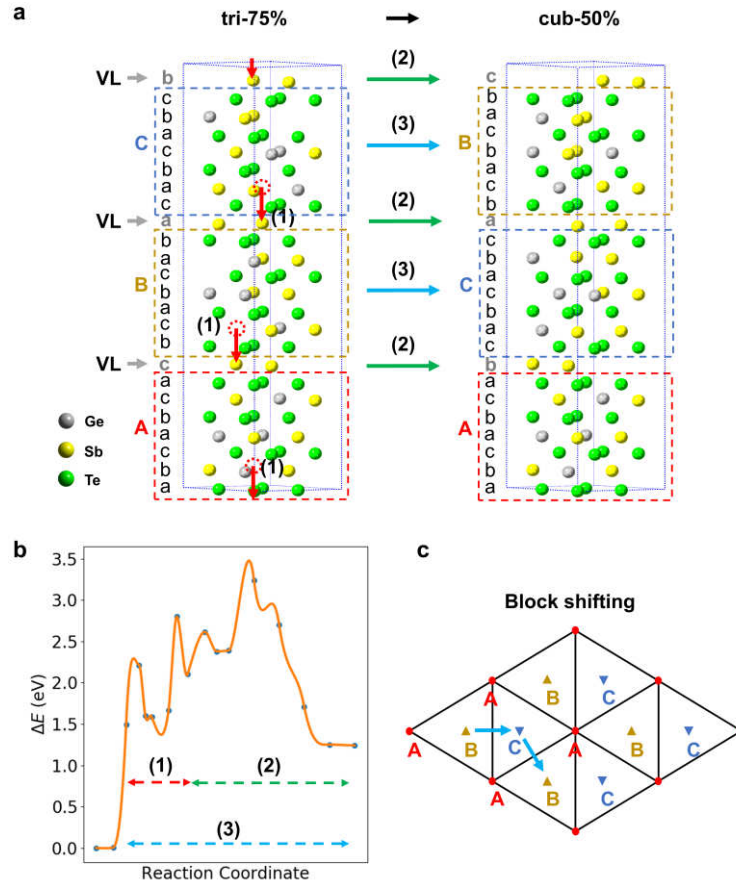


Figure 6. **a** A typical transition path from tri-75% to cub-50% in which three sub-processes occurred simultaneously: (1) out-of-plane "cation" diffusion, (2) in-plane "cation" diffusion, (3) SL block shifting. **b** The corresponding NEB energy profile. The first two peaks correspond to out-of-plane "cation" diffusion and the latter peaks correspond to in-plane "cation" diffusion, while the SL blocks shifted continuously during the transition with no obvious energy peak. **c** The shifting path of the top two SL blocks.

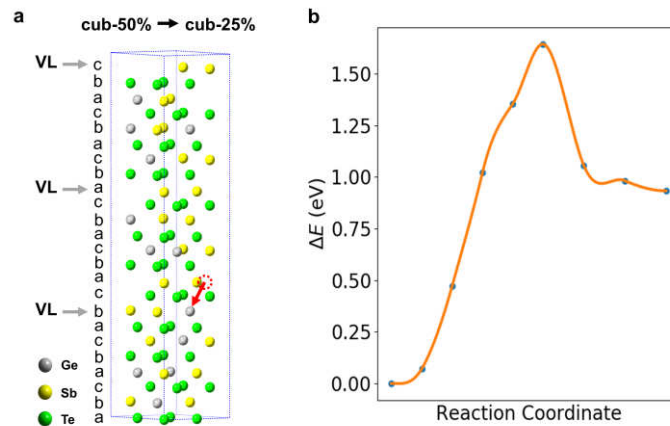


Figure 7. **a** A typical transition path from cub-50% to cub-25%. **b** The corresponding NEB energy profile. The energy barrier for the out-of-plane diffusion is ~ 1.7 eV.

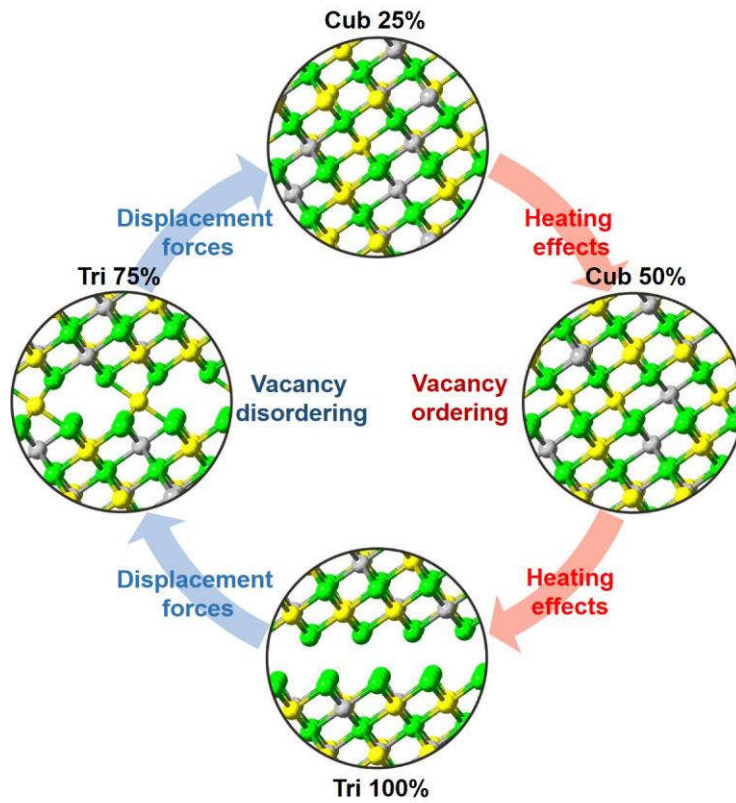


Figure 8 The schematic of the phase transitions in GST crystals. Vacancy ordering and disordering could be triggered by either heating effects or displacement forces, driving the reverse structural transition between the cub-phase and the tri-phase of GST.