

# Laser and Electrical Current Induced Phase Transformation of $\text{In}_2\text{Se}_3$ Semiconductor Thin Film on Si(111)

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**Abstract** Phase transformation of thin film ( $\sim 30$  nm)  $\text{In}_2\text{Se}_3$ /Si(111) (amorphous  $\rightarrow$  crystalline) was performed by resistive annealing and the reverse transformation (crystalline  $\rightarrow$  amorphous) was performed by nanosecond laser annealing. As an intrinsic-vacancy, binary chalcogenide semiconductor,  $\text{In}_2\text{Se}_3$  is of interest for non-volatile phase-change memory. Amorphous  $\text{In}_x\text{Se}_y$  was deposited at room temperature on Si(111) after pre-deposition of a crystalline  $\text{In}_2\text{Se}_3$  buffer layer (0.64 nm). Upon resistive annealing to  $380^\circ\text{C}$ , the film was transformed into a  $\gamma$ - $\text{In}_2\text{Se}_3$  single crystal with its  $\{0001\}$  planes parallel to the Si (111) substrate and  $(11\bar{2}0)$  parallel to Si  $(\bar{1}10)$ , as evidenced by scanning tunneling microscopy, low energy electron diffraction, and x-ray diffraction. Laser annealing with 20 nanosecond pulses (0.1 millijoules/pulse, fluence  $\leq 50$  mJ/cm<sup>2</sup>) re-amorphized the region exposed to the laser beam, as observed with photoemission electron microscopy (PEEM). The amorphous phase in PEEM appears dark, likely due to abundant defect levels inhibiting electron emission from the amorphous  $\text{In}_x\text{Se}_y$  film.

films on Si and the energetic requirements for phase change have not been fully explored.

$\text{In}_2\text{Se}_3$  exhibits a complex phase system in which various structures with different arrangements of inherent defects can form, in contrast to the simpler phase systems of  $\text{Al}_2\text{Se}_3$  and  $\text{Ga}_2\text{Se}_3$  [5,6]. This complexity originates from many energetically similar routes by which trivalent and divalent atoms can combine to satisfy their bonding requirements, and leads to several crystalline phases with the same  $\text{In}_2\text{Se}_3$  stoichiometry ( $\alpha$ ,  $\beta$ ,  $\gamma$ ,  $\delta$  and  $\kappa$ ) [7,8].

This paper reports on the crystallization and amorphization of  $\text{In}_2\text{Se}_3$  thin films on Si(111) ( $\sim 30$  nm thick) via resistive and laser annealing, respectively. The initially amorphous films were crystallized using resistive heating (time-scale of several seconds), while a 150 micron diameter spot was re-amorphized using nanosecond laser annealing. The film morphology and conductivity were probed with scanning tunneling microscopy (STM) and photoelectron emission microscopy (PEEM); the crystallinity was probed with X-ray diffraction (XRD) and low energy electron diffraction (LEED), and the stoichiometry with Rutherford back-scattering spectroscopy (RBS). The stoichiometry, morphology and crystal structure of the crystallized  $\text{In}_2\text{Se}_3$  film and its orientation with respect to the substrate are determined.

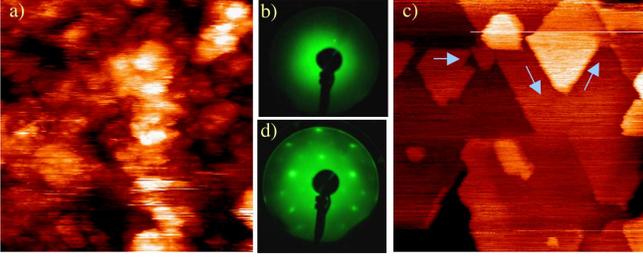
## 1 Introduction

Materials with rapid and reversible switching between their amorphous and crystalline phases, and which exhibit significant electrical resistivity and/or optical reflectivity change between these phases, are promising candidates for phase-change non-volatile memory. The strong change in optical reflectivity during the amorphous-crystalline phase transition in ternary chalcogenides is the basis of re-writable compact disc and digital video disk technology, with  $\text{Ge}_2\text{Sb}_2\text{Te}_5$  (GST) as the most commonly used material [1]. Over the past decade, binary chalcogenide semiconductors such as  $\text{InSe}$ / $\text{GaSe}$  and  $\text{In}_2\text{Se}_3$  have been investigated as potential phase-change media in proposed data storage and memory devices [2–4]. The electrical resistivity of  $\text{In}_2\text{Se}_3$  increases by a factor of  $10^5$  from the crystalline to the amorphous state [3,4], about 100 times larger than the resistivity change in GST [3]. However, the structural details of phase-change in  $\text{In}_2\text{Se}_3$

## 2 Experimental Details

Experiments were carried out both at the Pacific Northwest National Laboratory, for PEEM and RBS, and at the University of Washington, Seattle, for ultrahigh vacuum (UHV) STM and LEED and *ex-situ* two-dimensional X-ray diffraction (2D-XRD). The film was grown and crystallized in Seattle under UHV, and then transferred in atmosphere for 2D-XRD, RBS, and PEEM.

Silicon substrates were cut from commercial *p*-type Si(111) wafers, ( $\rho \sim 1\Omega\text{-cm}$ ) and prepared by repeated annealing of a chemically etched [9], outgassed sample to  $1100^\circ\text{C}$  for 5–10 s, followed by 20 min at  $800^\circ\text{C}$  and then a slow cool to room temperature to obtain a low-defect  $7\times 7$  reconstruction. Samples were heated by a direct current; the temperature was monitored using an optical pyrometer.



**Fig. 1** Scanning tunneling microscopy (a,c,  $200 \times 200 \text{ nm}^2$ ) and low energy electron diffraction (b,d) of  $\text{In}_2\text{Se}_3/\text{Si}(111)$  deposited at room temperature on a crystalline  $\text{In}_2\text{Se}_3$  buffer layer before (a,b) and after (c,d) annealing to  $380^\circ\text{C}$  (a) STM of an as-deposited film ( $-10 \text{ V}$ ,  $0.2 \text{ nA}$ , color z-scale  $0\text{-}5 \text{ nm}$ ); (b) LEED pattern of the surface at electron energy  $E_{el} = 47 \text{ eV}$ ; (c) STM of the same film, imaged at  $380^\circ\text{C}$  ( $-9.8 \text{ V}$ ,  $0.13 \text{ nA}$ , color z-scale  $0\text{-}1.2 \text{ nm}$ ); (d) room temperature LEED pattern at electron energy  $E_{el} = 47 \text{ eV}$  after annealing to  $380^\circ\text{C}$ .

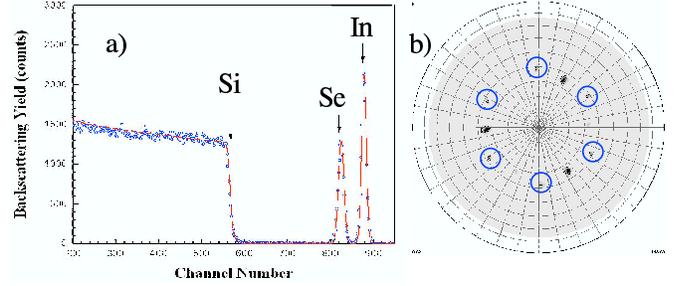
Indium selenide was evaporated from  $\text{In}_2\text{Se}_3$  granules (purity 99.999%) heated in a PBN crucible to temperatures between  $900$  and  $1000^\circ\text{C}$ , yielding a flux of  $2\text{-}5 \text{ \AA}/\text{min}$ , as measured by a quartz crystal microbalance (QCM). An epitaxial  $\text{In}_2\text{Se}_3$  layer of 2 bilayers ( $6.4 \text{ \AA}$ ) was deposited on the  $\text{Si}(111)$  substrate at  $555^\circ\text{C}$  to serve as a buffer layer prior to amorphous film deposition. Other experiments have shown this reduces the temperature needed to crystallize the film, as well as increases its crystallinity [10]. The amorphous  $\text{In}_x\text{Se}_y$  film was deposited atop the buffer layer at room temperature and subsequently annealed for crystallization (via electrical current through the substrate). The nominal deposition thickness of the film for which data are shown in Figs. 1–3 was  $47 \text{ nm}$  (flux measured before growth on QCM  $\times$  total deposition time); after annealing, RBS showed the film to be  $27.5 \text{ nm}$  thick.

X-ray diffraction was performed on a Bruker D-8 diffractometer, with the incident  $\text{Cu K}\alpha$  excitation passing through a single Göbel mirror and a  $0.5 \text{ mm}$  pinhole collimator. Diffracted radiation was captured with a multiwire area-detector. The beam was rastered across a sample area of  $0.7 \text{ mm} \times 0.7 \text{ mm}$ .

The PEEM system was equipped with both  $\text{Hg}$  ( $h\nu = 5.1 \text{ eV}$ ) and  $\text{D}_2$  ( $h\nu = 6.0 \text{ eV}$ ) sources for imaging a  $150 \mu\text{m}$  diameter region of the sample. A quadrupled Nd:YAG laser ( $\lambda = 266 \text{ nm}$ ,  $h\nu = 4.7 \text{ eV}$ , FWHM pulse width  $20 \text{ ns}$ , spot diameter  $100 \mu\text{m}$ ,  $0.1 \text{ mJ}/\text{pulse}$  at  $1 \text{ Hz}$  repetition rate), incident through a UV-grade fused silica window was used to amorphize the crystalline film *in situ*.

### 3 Results

Figure 1 shows STM and LEED of a  $\text{In}_2\text{Se}_3/\text{Si}(111)$  film before [Figs. 1(a) and (b)] and after [Figs. 1(c) and (d)] resistive heating to  $380^\circ\text{C}$ . The morphology in Fig. 1 (a) shows no distinctive structures, with an rms roughness of  $1.2 \text{ nm}$  on a  $100 \times 100 \text{ nm}^2$  area, showing the film exhibits no nanoscale surface crystallinity. LEED [Fig. 1(b)] likewise shows no long-range order. The sample was then slowly heated in the STM chamber during continuous scanning until crystallization was observed. The amorphous  $\text{In}_x\text{Se}_y$  film fully crystallized at



**Fig. 2** (a) Rutherford backscattering spectrum (dots) of the phase-change  $\text{In}_2\text{Se}_3$  thin film shown in Fig. 1, after annealing to  $380^\circ\text{C}$ . The solid line shows a fit to  $\text{In}:\text{Se} = 2:3$  and thickness =  $27.5 \text{ nm}$ . (b) Pole figure showing six-fold symmetry of the  $\gamma\text{-In}_2\text{Se}_3(0001)$  film and three-fold symmetry of the underlying  $\text{Si}(111)$  substrate.  $2\theta = 46.9^\circ\text{-}48.3^\circ$  [ $\text{Si}(220)$  at  $35^\circ$  and  $\gamma\text{-In}_2\text{Se}_3(20\bar{2}8)$  at  $38^\circ$ ]

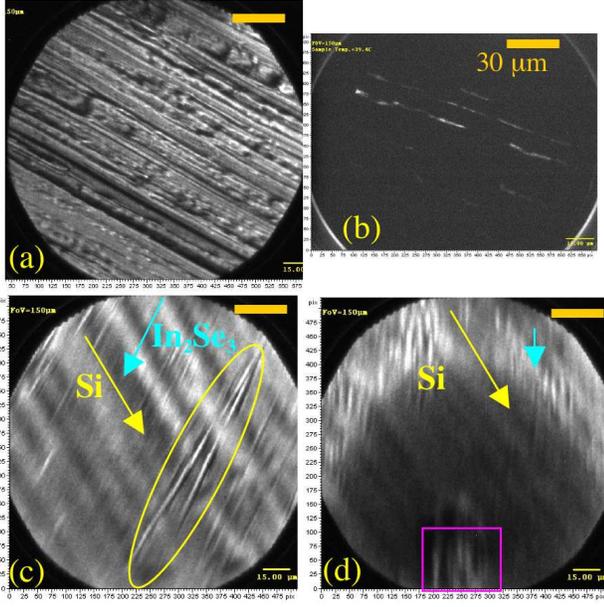
$380^\circ\text{C}$ , well below the melting temperature. The image in Fig. 1(c) was taken at  $380^\circ\text{C}$ . The post-annealed morphology [Fig. 1(c)] shows the edges of  $\text{In}_2\text{Se}_3$  terraces and the edges of triangular overlayer islands on the terraces (height  $0.32 \text{ nm}$ ) are aligned with directions of  $\text{Si}(111)$  (arrows); the rms roughness was  $0.14 \text{ nm}$  on a  $100 \times 100 \text{ nm}^2$  area. Subsequent, room temperature LEED patterns exhibited both  $1 \times 1$  and  $(\sqrt{3} \times \sqrt{3})\text{R}30^\circ$  spots [Fig. 1(d)].

The composition and thickness of the post-annealed  $\text{In}_x\text{Se}_y$  film were measured by Rutherford backscattering. In Fig. 2(a), the dotted line is the experimental data and the solid line is a fit for a 2 to 3 ratio of In to Se (held fixed) and a thickness of  $27.5 \text{ nm}$  (varied in the fit). The minimum yield ( $\chi_{min}$ ) value is 40%.

The 2D XRD measurements show that the annealed  $\gamma\text{-In}_2\text{Se}_3$  film, which is clearly oriented on a  $200 \text{ nm}$  length scale in Fig. 1(b), is predominantly a single crystal over a millimeter length scale. Other measurements without the crystalline buffer layer showed polycrystalline  $\gamma\text{-In}_2\text{Se}_3$  after annealing [10]. The only  $\text{In}_x\text{Se}_y$  phase observed for the film shown in Figs. 1 and 2 was  $\gamma\text{-In}_2\text{Se}_3$  with fully relaxed lattice parameters: the (0006) peak at  $2\theta = 27.65^\circ$  ( $d_{0006} = 3.23 \text{ \AA}$ ), (000 12) peak at  $2\theta = 56.97^\circ$  ( $d_{00012} = 1.61 \text{ \AA}$ ), and (10 $\bar{1}8$ ) at  $2\theta = 40.0^\circ$  ( $d_{10\bar{1}8} = 2.25 \text{ \AA}$ ). A broad peak at  $2\theta \sim 22.6^\circ$  was also observed that may result from an oxide (e.g.,  $\text{In}_2\text{O}_3$  or  $\text{In}_2(\text{SeO}_4)_3$ ).

The in-plane texture alignment of the  $\gamma\text{-In}_2\text{Se}_3$  film was probed via XRD pole figure measurements. The pole figure relative to the  $\text{Si}(111)$  surface normal for  $\gamma\text{-In}_2\text{Se}_3(20\bar{2}8)$  is shown in Fig 2(b). The  $\gamma\text{-In}_2\text{Se}_3(20\bar{2}8)$  planes have spacing  $d_{20\bar{2}8} = 1.90 \text{ \AA}$  and make an angle of  $38^\circ$  towards  $[10\bar{1}0]$  with  $\gamma\text{-In}_2\text{Se}_3(0001)$ , while  $\text{Si}(220)$  planes have spacing  $d_{220} = 1.91 \text{ \AA}$  and make an angle of  $35^\circ$  towards  $[11\bar{2}]$  with  $\text{Si}(111)$ . Because of the similarity in plane spacing, both peaks may be acquired in the same pole figure. The three-fold symmetric spots in Fig. 2(b) are thus attributed to the  $\text{Si}$  substrate, while the six-fold pattern is from well-oriented,  $\gamma\text{-In}_2\text{Se}_3(20\bar{2}8)$ . The  $30^\circ$  rotation between the  $\text{Si}(220)$  and  $\gamma\text{-In}_2\text{Se}_3$  scattering shows that the  $\gamma\text{-In}_2\text{Se}_3[11\bar{2}0]$  is aligned with  $\text{Si}[1\bar{1}0]$ .

While a slow anneal below the melting temperature can lead to crystallization of an amorphous film, returning to the amorphous state requires rapid heating and cooling to



**Fig. 3** PEEM images (field of view  $150\ \mu\text{m}$ , (yellow) scale bar  $30\ \mu\text{m}$ ) for (a) clean Si(111); (b) an amorphous  $\text{In}_2\text{Se}_3$  film on Si(111); (c) annealed  $\text{In}_2\text{Se}_3/\text{Si}(111)$  (same film as in Figs. 1 and 2; (d) after irradiation with 266 nm photon pulse. Images presented here are acquired with Hg lamp illumination ( $h\nu = 5.1\ \text{eV}$ )

quench in the disordered state. This was accomplished with pulsed laser annealing. The sample was monitored with photoelectron emission microscopy (PEEM), which distinguishes structures on the surface through a combination of their work function, electrical conductivity, and density of states near the Fermi level [11–13].

Fig. 3(a) shows a PEEM image of a clean Si(111) substrate after being Ar-sputtered and then annealed to  $\sim 850^\circ\text{C}$ . An amorphous  $\text{In}_2\text{Se}_3/\text{Si}(111)$  film deposited under similar conditions to those for the film in Figs. 1 and 2, but without a crystalline buffer layer, is shown in Fig. 3(b). The room-temperature-deposited  $\text{In}_2\text{Se}_3$  film presents a completely dark PEEM image with either Hg (5.1eV) or  $\text{D}_2$  (6.0 eV) illumination, except for defect lines running through the center of the image. The work function of  $\text{In}_2\text{Se}_3$  is reported to be 4.35 eV [14], which should yield emission with either source.

PEEM images of the same film as for Figs. 1 and 2 are shown before [Fig. 3(c)] and after [Fig. 3(d)] laser annealing. Similar results were obtained with both Hg and  $\text{D}_2$  illumination, although only the Hg results are shown. Comparison to bare Si [Fig. 3(a)] indicates that the wider stripe structure running from upper left to lower right arises from the underlying Si terrace structure (red arrows), while the narrower lines (blue arrows) correspond to the texture of the  $\text{In}_2\text{Se}_3$  films. Two crossed lines marked in the oval are defects on the surface.

The 266 nm laser was used to anneal a spot on the sample approximately 150 microns in diameter near the place where the image in Fig. 5(c) was taken. Single pulses of 0.1 mJ and 20 ns duration were measured at the laser head. Subsequent aperture-reduction and focusing culminated in final pulse fluence of  $\sim 50\ \text{mJ}/\text{cm}^2$ . The resulting annealed

region is shown in Fig. 3(d), where a dark circular region may be observed surrounded by features characteristic of the crystalline  $\text{In}_2\text{Se}_3$  film. The darker region still shows (though with less contrast) the structure characteristic of the underlying Si substrate, with  $\text{In}_2\text{Se}_3$  features less distinct.

## 4 Discussion

Room temperature deposition of  $\text{In}_2\text{Se}_3$  on Si(111) results in an amorphous film, even when a crystalline template is first deposited. The nominal thickness from the crystal monitor was over 40% larger than that measured with RBS after annealing. The In:Se ratio measured by RBS after annealing (2:3) was equal to that in the incident deposition flux, indicating that any evaporation must have been congruent. However, rather than re-evaporation at  $380^\circ\text{C}$ , it is more likely either that the flux decreased during deposition or that the geometry was slightly different from when the QCM was calibrated. The rms height variation on a  $100 \times 100\ \text{nm}^2$  area is reduced from 1.2 nm to 0.14 nm ( $\sim 1$  bilayer) upon annealing. On similar  $100 \times 100\ \text{nm}^2$  areas, the original Si(111) substrate had an rms roughness of 0.37 nm, while the crystalline buffer layer (0.7 nm thick) had an rms roughness of 0.23 nm. After further annealing than the images in Fig. 1, the surface showed 8 nm deep triangular pits with a density  $\sim 10$  per square micron, and 0.22 nm rms roughness between the pits.

There are several known crystal structures of  $\text{In}_2\text{Se}_3$ , each with a different arrangement of the intrinsic vacancies.  $\alpha$ - $\text{In}_2\text{Se}_3$  has a layered hexagonal structure with  $a = 4.03\text{\AA}$  and  $c = 19\text{\AA}$ . Each layer is built up from tetrahedral covalent bonds in the sequence Se-In-Se-In-Se (a penta layer) along the  $c$  axis [15], and the In vacancies are localized in (0001) planes. The  $\beta$  phase is also hexagonal ( $a = 4.03\text{\AA}$  and  $c = 28\text{\AA}$ ) and metastable. It can be obtained by heating  $\alpha$ - $\text{In}_2\text{Se}_3$  to  $523\ \text{K}$  [16].  $\gamma$ - $\text{In}_2\text{Se}_3$  is the stable room temperature phase, a defected wurtzite structure with  $a = 7.13\text{\AA}$ ,  $c = 19.38\text{\AA}$  [7,17,18]. The unit cell of the  $\gamma$ - $\text{In}_2\text{Se}_3$  contains six bilayers. Each In layer has vacancies arranged in a  $(\sqrt{3} \times \sqrt{3})R30^\circ$  pattern, with a helical alignment from layer to layer [7,17,18]. The  $\delta$  phase exists at very high temperature (973 K) and could be a product of dissociation of  $\text{In}_2\text{Se}_3$ . A new phase,  $\kappa$ - $\text{In}_2\text{Se}_3$ , with the unit cell symmetry similar to  $\alpha$ - $\text{In}_2\text{Se}_3$  was recently discovered, with Zn doping necessary to obtain and stabilize  $\kappa$ - $\text{In}_2\text{Se}_3$  thin films at room temperature [19,20].

Annealing amorphous  $\text{In}_2\text{Se}_3/\text{Si}$  resulted in formation of single crystal  $\gamma$ - $\text{In}_2\text{Se}_3$ , which is supported by LEED, STM and XRD. The hexagonal  $1 \times 1$  LEED spots would arise from the (oriented) hexagonal lattice in 0001 planes of any wurtzite-based phase of  $\text{In}_2\text{Se}_3$ , but the  $(\sqrt{3} \times \sqrt{3})R30^\circ$  spots are consistent with intrinsic vacancy ordering in  $\gamma$ - $\text{In}_2\text{Se}_3$  continuing to the surface.  $\alpha$ - $\text{In}_2\text{Se}_3$ , in contrast, has no reason for surface reconstruction, since the (0001) planes between penta-layers have no dangling bonds. The LEED is thus excellent evidence for long range order in the intrinsic vacancies, and shows the surface of the 30 nm thick film is still oriented relative to the underlying Si substrate. The STM data show this oriented alignment as well, with the step edges of the  $\text{In}_2\text{Se}_3$  film well aligned with those of the underlying sub-

strate. The STM step height of 0.32 nm is also consistent with the  $\gamma$ -phase (it would be 0.95 nm in the  $\alpha$ -phase).

The XRD data further confirms the assignment of single-crystal  $\gamma$ - $\text{In}_2\text{Se}_3$ . The plane spacings are slightly larger in the  $\gamma$ -phase than in the  $\alpha$ -phase. The off-axis planes such as (20 $\bar{2}$ 8), (10 $\bar{1}$ 8) and (11 $\bar{2}$ )6 that we observe would also be at slightly different orientation angles and  $2\theta$ -values. The pole figure in Fig. 2(b) clearly shows the relative orientation of the Si and  $\gamma$ - $\text{In}_2\text{Se}_3$  lattices. The Si(220) spots lie in an azimuth 30 from the  $\gamma$ - $\text{In}_2\text{Se}_3$ (20 $\bar{2}$ 8). Note the underlying wurtzite lattice is not rotated relative to the diamond structure, however, since the ( $\sqrt{3} \times \sqrt{3}$ )R30 $^\circ$  vacancy ordering that determines the  $\gamma$ - $\text{In}_2\text{Se}_3$  unit cell yields unit vectors rotated 30 $^\circ$  from those of the underlying wurtzite structure.

The crystallization process of amorphous  $\text{In}_2\text{Se}_3$  occurs at a temperature well below that of the melting temperature [21], and also occurs at a lower temperature (380 $^\circ\text{C}$ ) when a crystalline buffer layer is deposited before the amorphous layer than when the film is deposited directly on Si(111) at room temperature (450  $^\circ\text{C}$  for 17 nm thick film). Annealing the pure amorphous film resulted in a polycrystalline film, with random in-plane orientation, although the (0001) axis was still normal to the film [10].

The inverse process of amorphization likely requires local melting of the  $\text{In}_2\text{Se}_3$ . An order of magnitude estimate of the energy required for amorphization is  $3Nk_B * 1000$  K, [m.p.=1158 K,  $\gamma$ - $\text{In}_2\text{Se}_3$ ] where  $N$  is the number of atoms in the heated volume and  $k_b$  is Boltzmann constant. The volume of film exposed to the laser (27.5 nm thick, 100  $\mu\text{m}$  diameter) had about  $7 \times 10^{13}$  atoms, for which  $3Nk_b * 1000$  K is about 3  $\mu\text{J}$ , or about 1/30 the energy in each laser pulse. Since the absorption length of the 266 nm radiation is 10-100 nm ( $\alpha \sim 10^5 - 10^6 \text{ cm}^{-1}$ ) [16], some of the energy is deposited in the substrate; additional energy is also lost to the optics. It is thus not surprising that a few pulses were required to initially heat the area before a single pulse would carry a region over threshold for amorphization.

The contrast between the crystalline and amorphous regions in PEEM could arise from differences in work function, in densities of states near the Fermi level, or in film conductivity. The work function of crystalline  $\text{In}_2\text{Se}_3$  is 4.35 eV [14], so both Hg (5.1 eV) and D<sub>2</sub> (6.0 eV) illumination should result in electron emission. Our photoemission measurements on a 1 nm thick film [10] show an In 4d core level shift of  $\sim 0.7$  eV to higher binding energy (relative to the Fermi energy) from the amorphous to the annealed phase. If the In shift were completely due to a change in work function, then the Hg radiation would be close to threshold for photoemission, but there should still be significant emission with D<sub>2</sub> radiation. However, we find no emission with either source for the as-deposited film except at large scale defects [Fig. 3(b)]. It is thus likely that photoexcited electrons become trapped in the amorphous material and cannot travel to the surface for emission into vacuum. The laser-amorphized film [Fig. 3(d)] still shows a small amount of emission, mostly from the underlying Si substrate, indicating it has fewer trapping defects than the as-deposited film, but still no states that can emit electrons. Upon returning to the same point on the sample as Fig 3(c), the defect in the oval greatly increased in relative intensity, becoming twice as bright as its surroundings. This indicates impact of the laser beam through heating and/or

free carrier excitation in the substrate over a range of a few hundred microns.

## 5 Conclusions

The control of material phase (amorphous, or one of several possible crystalline or polycrystalline states) is essential to the properties of modern devices. The difference in properties associated with a phase change may be exploited for device operations, as in DVDs where the change in optical reflectivity between crystalline and amorphous states allows media rendition. Driven by the need for a new paradigm to continue the relentless increase in both density and speed that has characterized developments in computer memory for four decades, one promising avenue for non-volatile memory is phase-change memory (PCM), for which chalcogenides are prime candidates. Amorphous-crystalline phase transitions can be achieved by applying voltage pulses with different amplitudes and durations in the range of tenths of nanoseconds. In our study, we have shown that rapid, selective, and local crystal-amorphous transformation can be realized using pulsed, UV nanosecond lasers at low fluences. This is a necessary step in the development of high density/high-speed chalcogenides-based optical PCM.

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