

UC Santa Cruz

UC Santa Cruz Electronic Theses and Dissertations

Title

Investigation of Amorphous Selenium Based Detectors

Permalink

<https://escholarship.org/uc/item/07f1d9rg>

ISBN

9798184189802

Author

McGrath, Molly

Publication Date

2026-06-29

Peer reviewed|Thesis/dissertation

UNIVERSITY OF CALIFORNIA
SANTA CRUZ

**INVESTIGATION OF AMORPHOUS SELENIUM BASED
DETECTORS**

A thesis submitted in partial satisfaction of the
requirements for the degree of

MASTERS OF SCIENCE

in

ELECTRICAL ENGINEERING

by

Molly McGrath

June 2026

The Thesis of Molly McGrath
is approved:

Professor Shiva Abbaszadeh, Chair

Professor Ahmet Yanik

Professor Nobuhiko Kobayashi

Peter Biehl
Vice Provost and Dean of Graduate Studies

Table of Contents

List of Figures	v
Abstract	vii
Dedication	ix
Acknowledgments	x
1 Introduction	1
2 a-Se Investigation	6
2.1 Fabrication and Characterization of a-Se Detectors	9
2.2 Effect of Breaking Vacuum on a-Se performance	14
2.3 Discussion and Future work	19
3 Selenium Tellurium alloy photodetectors and Hole Blocking Layers	21
3.1 Se-Te Fabrication	22
3.2 Hole blocking materials	23
3.2.1 Parylene	24
3.2.2 Cerium Oxide Quantum Dots	26
3.2.3 Silicon Dioxide	27
3.2.4 Aluminum Oxide	27
3.2.5 Tin Oxide	28
3.3 Comparison HBLs	28
4 Oxidation in Se-Te alloys	35
4.1 Materials and Fabrication	37
4.2 SEM-EDS	38
4.3 XRD	44
4.4 XPS	46
4.5 Further characterization	48

5 Conclusion	53
Bibliography	56

List of Figures

1.1	Diagram of a dual-layer flat-panel detector [1].	2
1.2	Left: Indirect Detector architecture. Right: Direct Detector architecture [2].	2
2.1	The quantum efficiency of holes in a-Se versus applied electric field with different light wavelengths [3].	8
2.2	Left side: Au E-beam evaporator. Right side: a-Se thermal Evaporator.	10
2.3	(a) Image of the ToF, Dark and photocurrent measurement setup. (b) Schematic demonstrating the deposition process for a single deposition (left) and the samples in which vacuum was broken to add additional Se to the evaporation crucible (right). (c) Picture of sample S3, in which both layers of a-Se can be noted and the 14 devices can be seen.	13
2.4	(a) Hole mobility for the 4 samples of breaking vacuum study. (b) Dark current density for the 4 samples up to 10 V/ μm . (c) QE found from ToF measurements at a wavelength of 355 nm. (d) QE found from dark and photocurrent measurements at a wavelength of 470 nm.	15
2.5	(a) Current output for samples S1 and S3 for dark and photocurrent measurements of a 1 second pulse of 470 nm light while biased at 10 V/ μm . (b) Lag for samples S1 and S3 after being exposed for 1 second by 470 nm at 10 V/ μm	17
2.6	a) Hole mobility samples with a-Se thicknesses ranging from 15 μm to 245 μm with applied fields of 0-35 V/ μm . (b) QE of samples calculated from ToF measurements. Black line shows the efficiency predicted by the Onsager model. (c) Hole mobility for additional 2 samples in breaking vacuum study, compared to range reported in (a) and in other works. b) QE for the same samples as (c), compared to range reported in (b) and in other works.	18
3.1	Flat Panel Detector Architecture [4]	22
3.2	Band gap diagram for potential Hole Blocking Layers.	23
3.3	Data for parylene samples S90-S440 [4]	25

3.4	Comparison of Lag in Parylene samples [4]	26
3.5	Current Density of samples with each hole blocking layer. (a) SiO_2 , (b) Al_2O_3 , (c) $\text{SnO}_2/\text{SiO}_2$, (d) $\text{SnO}_2/\text{Al}_2\text{O}_3$, (e) SnO_2	29
3.6	Comparison of the top HBL contenders for Dark current and EQE	30
3.7	Flat Panel Detectors with zoom in of red square.	31
3.8	(a) Linearity: Photocurrent density vs Intensity. (b) Pulse shape: Photocurrent density vs time. (c) Lag vs time. (d) Ghosting: Multiple exposures of light.	33
4.1	(a) Photo of an SeTe device, lit from behind. (b) Microscopic image of a film rupture. (c) SEM image of another hole in the surface of a Se-Te film.	36
4.2	SEM imaging of pellets from batch A (a-d) and B (e-g)	39
4.3	Cross-section SEM of Batch A pellets	40
4.4	SEM images of Batch B pellet in same position to showcase aging over multiple weeks. Yellow box in (b) highlights area of focus in Fig. 4.5.	41
4.5	SEM of blemish in yellow box in Fig. 4.4b	41
4.6	EDS spectra of Batch A, B and B aged. Grey vertical lines mark transitions assigned to each peak.	42
4.7	SEM of Se-Te film surfaces for devices made from Batch A and B.	44
4.8	XRD spectra for crushed pellets. Dashed blue lines highlight peaks formed in Batch A and B (aged) samples. Reflections for t-Se (ICSD-40018), t-Te (ICSD-161690), TeO_2 (ICSD-56004), and SeO_2 (ICSD-72366) are shown in the bottom quarter of the plot.	45
4.9	XPS spectra for all batches, with noted peaks marked and dashed blue lines for associated binding energies.	47
4.10	(a) Dark current and (b) external quantum efficiency of devices made from Batches A and B, with and without an SiO_2 blocking layer.	50

Abstract

Investigation of Amorphous Selenium based detectors

by

Molly McGrath

Photoconductive materials have become increasingly important for high-resolution radiation detectors because of their ability to directly or indirectly convert ionizing radiation into electrical signals. Optimizing material properties, device architectures, and fabrication processes is essential for developing next-generation imaging detectors with improved sensitivity, image quality, and reliability.

This thesis investigates the fabrication, characterization, and optimization of photoconductive detector materials for radiation imaging applications. First, the effect of interrupting the deposition process during thick-film fabrication is systematically evaluated through electrical and optical characterization, including carrier transport, dark current, quantum conversion efficiency, and temporal response measurements. The results demonstrate that interrupting the deposition process does not significantly degrade detector performance, providing a practical approach for fabricating thicker photoconductive layers required for high-efficiency radiation detection.

The second part of this work explores interface engineering through the investigation of multiple carrier-blocking layer materials to improve detector performance. Their effects on leakage current, charge collection efficiency, temporal response, and large-area device fabrication are systematically compared. The study identifies inter-

face designs that effectively suppress charge injection while maintaining high photoconductive performance and demonstrates their compatibility with large-area detector fabrication.

Finally, the influence of material oxidation on structural, chemical, and optoelectronic properties is investigated using complementary materials characterization techniques and correlated with device performance. The results provide insight into degradation mechanisms affecting detector stability and long-term reliability.

Overall, this work advances the understanding of photoconductive radiation detectors by establishing practical fabrication strategies, optimizing interface engineering, and identifying material degradation mechanisms. These findings provide important design guidelines for the development of next-generation high-resolution imaging detectors for medical imaging and related radiation detection applications.

To my family, who I'm sorry I stressed out completely, but thank you for your
love and support. You pushed me to go far.

Acknowledgments

I want to thank Professor Shiva Abbaszadeh for this opportunity, Dr. Katie Hellier for her support and guidance, and the rest of the Radiological Instrumentation Lab.

Chapter 1

Introduction

Photoconductors are semiconductors whose electrical conductivity increases upon absorption of photons with energy greater than the bandgap through the generation of mobile charge carriers. The incident photons from a light source exposed to the material generate free charge carriers in the material increasing electrical conductivity under an applied electric field. When illumination ceases, the generation of new carriers stops, and the excess carriers gradually decay through charge collection, trapping, detrapping, and recombination until the material returns to its equilibrium dark conductivity.

Recent advances in medical imaging have driven significant interest in dual-layer detector architectures, which enable simultaneous acquisition of low- and high-energy X-ray information from a single exposure. Compared with conventional single-layer detectors, dual-layer detectors provide improved material discrimination and quantitative imaging while reducing motion artifacts associated with sequential dual-energy

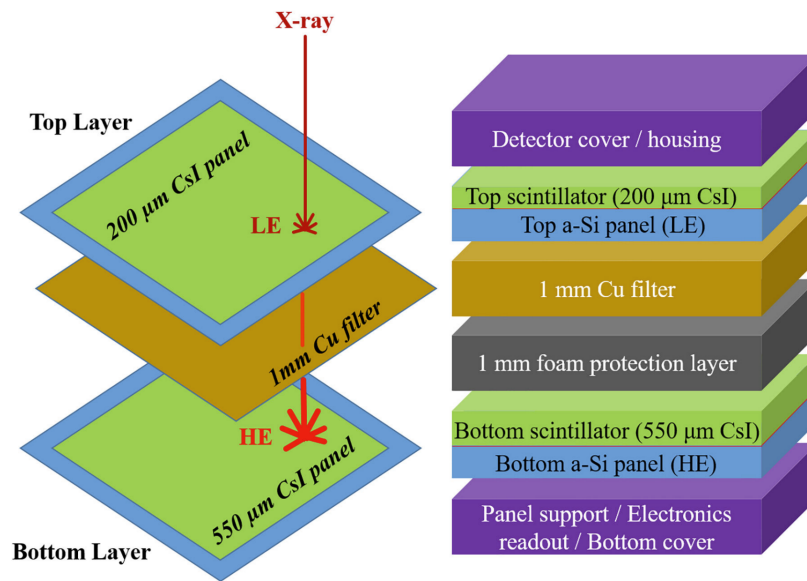


Figure 1.1: Diagram of a dual-layer flat-panel detector [1].

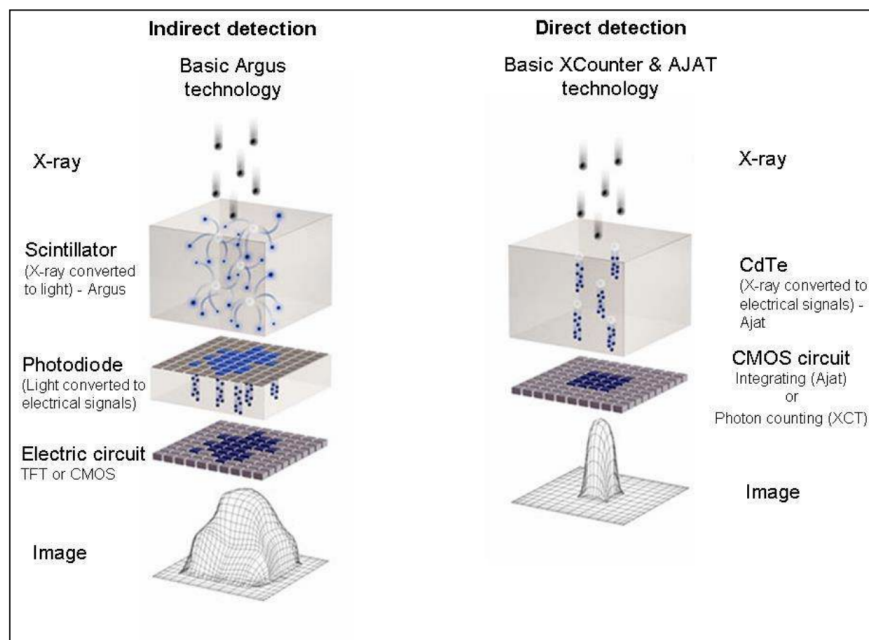


Figure 1.2: Left: Indirect Detector architecture. Right: Direct Detector architecture [2].

acquisitions [5, 6, 1]. Furthermore, hybrid direct-indirect dual-layer detector designs have the potential to combine the high spatial resolution of direct-conversion detectors with the high detection efficiency of indirect-conversion detectors, enabling improved image quality and spectral performance. A schematic of a CsI-based dual-layer flat panel detector can be seen in Fig. 1.1. Using a single shot from a polyenergetic X-ray source, two images would be produced - one from each layer, which attenuate low energies (top layer) and high energies (bottom layer) - that can be combined to provide a clear material decomposition without blurring due to motion [7]. The Radiological Instrumentation Laboratory (RIL) has proposed a hybrid direct-indirect conversion dual layer detector. The top layer would be a-Se based direct detector for low energy attenuation ($< \sim 35$ keV) that produces a high-resolution image. The bottom layer uses an indirect Se-Te alloy detector for high energy attenuation ($> \sim 35$ keV), collects the remaining X-rays, and has the potential to reach impact ionization.

Fig. 1.2 shows architecture for both direct and indirect detectors. A direct detector converts incident rays directly into an electrical signal by generating charge carriers in the detector material. Conversely, in an indirect detector, X-rays are first converted into visible photons by a scintillator before being detected by a photodiode.

Amorphous Selenium (a-Se) is a well known photoconductor that supports high-resolution and high-sensitivity imaging over a large range of energies in direct conversion detectors [8]. Direct conversion detectors use a photoconductor to absorb, convert, and collect carriers from incident X-rays. Compared to other semiconductor detector materials such as PbO, CdTe, and CdZnTe, a-Se offers unique advantages

such as avalanche multiplication through impact ionization at electric fields exceeding 65 V/ μm , and low-cost, large-area deposition using established fabrication processes [9, 10, 11, 12, 13, 14, 15, 16].

The energy spectrums that are attenuated and pass through the top layer are controlled by the thickness of the semiconductor, so we pursue an investigation into quantum conversion efficiency and hole mobility dependence on the thickness of a-Se. For the top, direct conversion layer, in order to achieve high attenuation of the low energy spectrum, a thickness of greater than 300 μm is necessary [17, 18]. RIL has a dedicated thermal evaporator for a-Se that can deposit layers up to 250 μm in a single fabrication [19]. To deposit thicknesses larger than that requires a break in vacuum to refill the crucible with additional selenium material. Thus, research into air exposure mid-fabrication for multiple depositions of the semiconductor is necessary.

Selenium–tellurium (Se–Te) alloys are a promising material for a broad range of optoelectronic and energy conversion applications and have attracted considerable research interest because of their combination of tunable photoconductive, thermoelectric, and phase-change properties. Incorporating Te into Se is known to reduce the material bandgap, thereby extending its photoconductive response into the near-infrared region, while increasing conductivity [20, 21, 22, 23, 24, 25]. Bulk films with Te concentrations above 30% tend to promote crystallization. However, recent studies of thin film devices have demonstrated amorphous structures with Te contents as high as 40% [26]. Within amorphous Se-based films, the addition of small amounts of Te has been shown to improve structural stability and increase the crystallization temperature, enhancing

their suitability for optoelectronic applications [27, 28]. Despite these advantages, increasing Te concentration can degrade the transport properties due to additional defect states that appear with Te bonding [29]. Nevertheless, Se–Te alloys have been successfully used in broadband photodetectors, solar cells, and high-gain avalanche rushing photodetectors (HARP) [30].

For the indirect detector, we investigate the tunability of optoelectronic properties that occur when varying the percentage of Te in the Se-Te alloy to find the optimized concentration for our purpose. To counteract the additional defect states introduced, we investigate different hole blocking layers to promote carrier mobility.

Chapter 2

a-Se Investigation

Amorphous selenium has been studied extensively starting in the 1940s for its efficient X-ray photoconductivity [31]. It was widely used in the photocopy industry as the photoconductive layer in xerographic photoreceptors due to its excellent spatial resolution until the 1980s, when it was largely replaced by lower-cost organic photoconductors [11]. Interest in a-Se was revived in the 1990s with the advent of thin-film transistor (TFT)-based active matrix arrays (AMAs), enabling the development of large-area direct-conversion flat-panel X-ray detectors for medical imaging [32].

The resurgence of interest in a-Se is due to the appealing physical properties for use in both direct and indirect photoconversion detectors. The material has a bandgap of 2.2 eV and the ability to reach high photodetection efficiency at wavelengths less than 400 nm [33]. Due to its high resistivity, a-Se has a low dark current and thermal noise making it attractive for medical imaging usage [12]. The charge transport is also non-dispersive over time which creates predictable charge behavior in a-Se based devices

[17].

Fig. 2.1 shows the quantum efficiency of a-Se over a range of applied E-fields with different wavelengths. It was found that the optical quantum or photogeneration efficiency in a-Se depends on the incoming light wavelength, the applied electric field and temperature [34]. At longer wavelengths, the initial separation between carriers is smaller, resulting in lower dissociation efficiency and reduced quantum efficiency. However, when an electric field is applied, the likelihood of electron-hole pair dissociation increases, leading to higher photogeneration efficiency at stronger electric fields [3].

When biased with electric fields exceeding approximately $10 \text{ V}/\mu\text{m}$, a-Se exhibits efficient X-ray photoconductivity due to improved charge collection efficiency, making it well suited for direct-conversion X-ray imaging detectors, particularly in mammography [35]. At fields even higher ($> 65 \text{ V}/\mu\text{m}$), impact ionization can occur in the sample. To avoid joule heating that can cause thermal crystallization of the material, an insulating layer is included in the device; RIL commonly uses polyimide (PI). This allows for a-Se to remain stable at higher electric fields and allow for low leakage current and high conversion efficiency [36, 37]. Previous efforts that utilized hole blocking layers have led to the stable application of high fields for achieving avalanche multiplication [38, 39].

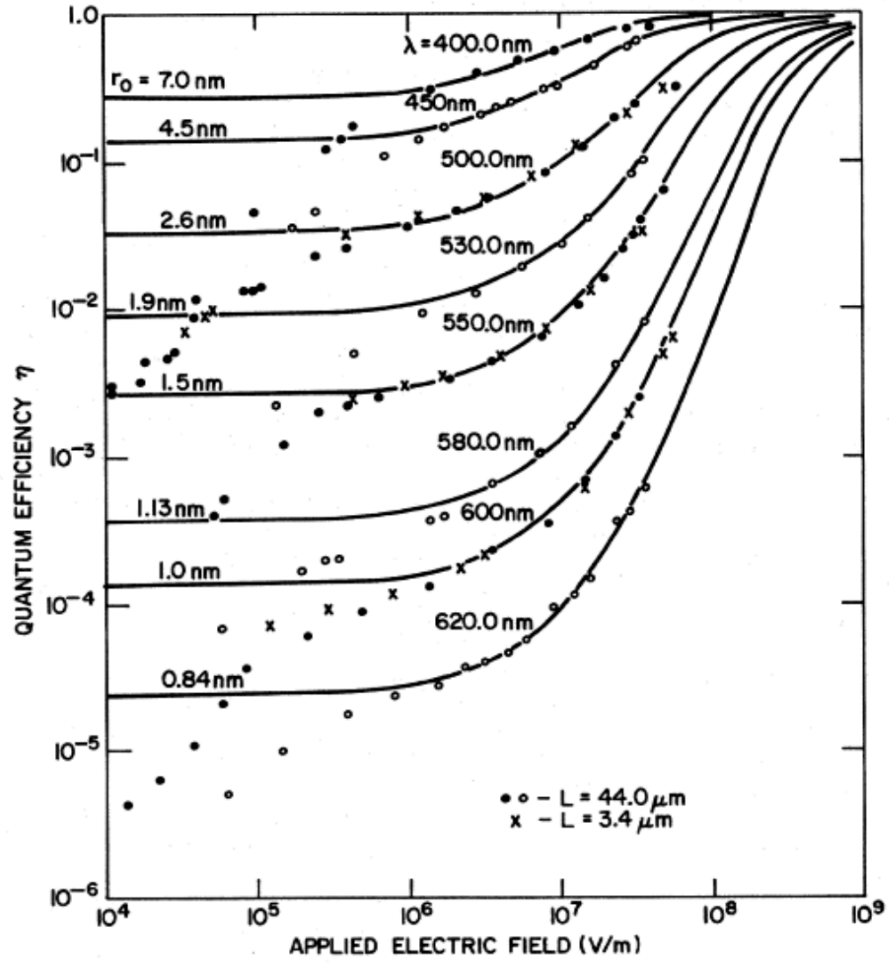


Figure 2.1: The quantum efficiency of holes in a-Se versus applied electric field with different light wavelengths [3].

2.1 Fabrication and Characterization of a-Se Detectors

The fabrication process for the samples begins with a thorough cleaning of a 3"x1" glass slide coated with Indium Tin Oxide (ITO). The cleaning procedure involves placing three slides side by side in a petri dish containing acetone and subjecting them to ultrasonic waves for cleaning for 10 minutes. Afterward, the slides are rinsed over the first petri dish using isopropyl alcohol (IPA) and are then transferred to a fresh petri dish filled with IPA. This is followed by another 10-minute ultrasonic treatment. Finally, the slides are rinsed with deionized water and dried using nitrogen gas.

For samples with the PI insulating layer, the preparation of the PI begins by warming the high-concentration solution to room temperature for approximately 30 minutes. The weight of a clean liquid squeeze dropper bottle is first measured, followed by the addition of the warmed PI to fill the bottle about halfway. The new weight is recorded, and for every 2 grams of PI in the bottle, 1 mL of N-Methyl-2-pyrrolidone (NMP) is added using a micropipette. As PI is highly viscous, the components are gently mixed by slowly rolling the bottle between the hands to avoid the formation of air bubbles as they can cause streaking during the deposition process.

To deposit PI onto the samples, we put Kapton tape along the edges of the cleaned slides, leaving a few millimeters exposed on each side. The slides are then placed in a spin coater which pulls vacuum on the sample to secure them in position. With a single, continuous pour of the previously prepared diluted PI solution, the entire surface of the slide is covered. Afterward, the top of the spin coater is covered, and the slide is

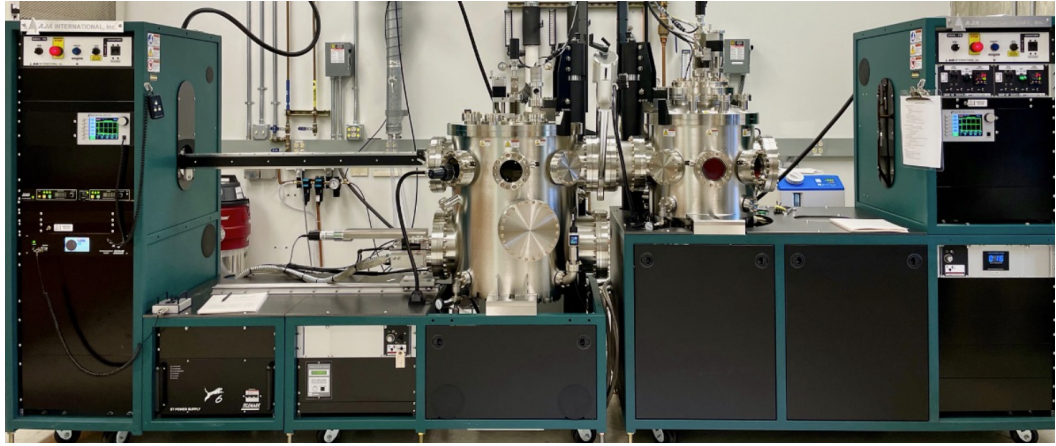


Figure 2.2: Left side: Au E-beam evaporator. Right side: a-Se thermal Evaporator.

left to rest for at least 30 seconds to allow initial bonding of the PI to the ITO layer. A preset spin-coating program is used, which ramps up to 1450 rpm (rotations per minute) over 5 seconds and maintains that speed for 180 seconds, then takes another 5 seconds to ramp down. Once spinning stops, the slide is quickly removed and placed on a hot plate preheated to 90°C. After all samples have been coated, they are left on the hot plate for at least 90 seconds. The temperature is then increased to 150°C and samples rest for another 90 seconds. Finally, the temperature is further raised to 350°C at a ramp rate of 240°C per hour. Once the temperature reaches 350°C, the samples are held at this temperature for 30 minutes before the heat is turned off, which allows them to cool gradually back to room temperature.

Once cleaning and PI deposition is complete, the a-Se deposition process can begin. Deposition chambers for a-Se and Au (as a top contact) are shown in Fig. 2.2. The procedure begins by venting the chamber of the thermal evaporator, ensuring

that all pumps are sealed, and filling it with nitrogen (N₂). When the pressure gauge indicates that the chamber has reached atmospheric pressure (760 torr), the chamber lid is lifted to access the interior. Three substrates are then loaded on side by side to the platen. A shadow mask is positioned on top of the substrates and secured with screws. The plate is then reattached to the chamber lid.

Next, the condition of the quartz crystal monitor (QCM) is checked — if the crystal life is below 20%, it is replaced with a new one. Additional a-Se is prepared by first zeroing the weight of a plastic petri dish, pouring the desired amount of a-Se pellets into the dish, and measuring the weight. The a-Se is then transferred into the Radak crucible using a funnel, ensuring proper loading for the deposition process.

Once all components are loaded into the thermal evaporation chamber the chamber lid is then closed, and the chamber is evacuated. Initially, the roughing pump is used until a pressure of approximately 1e-2 Torr is reached. At this point, the roughing pump is closed, and the cryo pump is engaged to further reduce the pressure to around 1e-6 Torr.

To begin the deposition process, the heater components are activated, and the system is set to remote control for automation. Once the substrate reaches the target temperature, the desired thickness of the a-Se layer to be deposited is selected. The computer system regulates the power to the crucible heater and monitors the deposition rate of the a-Se using the QCM. The temperature is gradually increased until a deposition rate of 105 Å/s is achieved. At this point, the shutter to the substrate is opened, allowing deposition onto the slides to begin. The shutter remains open until the

computer determines that the desired thickness has been achieved, based on the QCM readout. Once deposition is complete, the shutter closes, and the heater is turned off. The system undergoes a prolonged cooling period of approximately 6 hours before the chamber can be vented, and the substrates removed.

For some samples, the a-Se deposition process was repeated in order to have two layers of the photo-conductive material.

Once the a-Se depositions is complete, an electron beam evaporator is used to deposit gold (Au) top contacts onto the samples. This process begins similarly by venting the evaporator chamber, placing the substrates on the platen, and securing them with a shadow mask containing 3 mm, 4 mm, and 5 mm holes to define the top contact areas. After sealing the chamber, it is evacuated until the pressure reaches approximately $1\text{e-}6$ Torr. Once this level is achieved, the system is set to automation, targeting a deposition thickness of 100 nm of Au. During the deposition, consistent monitoring is required to ensure that the electron beam remains correctly aligned with the crucible. Upon completion, the system and substrates are allowed to cool before venting the chamber and removing the samples. An illustration of the fabricated device architecture can be found in Fig. 2.3b. A fully fabricated device, specifically the breaking vacuum sample S3, can be seen in Fig. 2.3c.

The devices were characterized using transient photocurrent time-of-flight (TOF) measurements to determine the hole carrier transit time and calculations of hole mobility. This process was done by applying a bias for each sample based on its final thickness, limited by either the maximum voltage supplied by the power source or de-

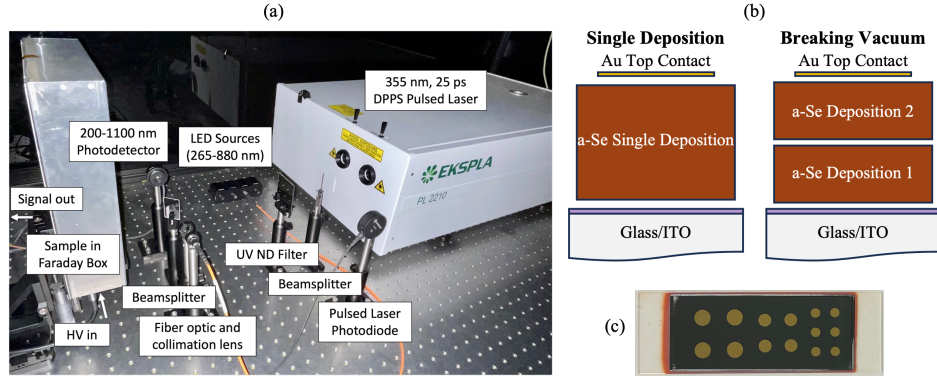


Figure 2.3: (a) Image of the ToF, Dark and photocurrent measurement setup. (b) Schematic demonstrating the deposition process for a single deposition (left) and the samples in which vacuum was broken to add additional Se to the evaporation crucible (right). (c) Picture of sample S3, in which both layers of a-Se can be noted and the 14 devices can be seen.

vice breakdown due to crystallization, then flashing a 355 nm high intensity laser at the sample and observing the resulting voltage signal output on an oscilloscope. The transit time was identified as the interval from the initial pulse to the point where the signal plateau decreased by 50% [40].

The devices were further characterized by dark and photocurrent measurements to determine leakage current, quantum conversion efficiency (QE), and lag - the time taken to return to a percentage of the signal [41]. Leakage or dark current is the current output with the absence of light and gives important information to the performance of the detector to optimize the signal to noise ratio and to prevent breakdown due to Joule heating [42]. In this process, the samples were again biased to an appropriate

E-field based on their thickness and flashing with an LED at 470 nm wavelength to see the current output on an picoammeter. The optical setup for sample measurements of ToF, dark and photocurrent measurements can be seen in Fig. 2.3a.

2.2 Effect of Breaking Vacuum on a-Se performance

The final thicknesses for the single deposition samples, S1 and S2, were 223 μm and 171 μm , respectively. For the samples where vacuum was broken, the final thicknesses were 223 μm for S3 and 200 μm for S4. The deviations from the target thickness are attributed to fluctuations in the deposition rate, which was approximately 105 $\text{\AA}/\text{s}$ when the QCM was shuttered. Although the power and temperature were kept constant, the deposition rate could still vary by ± 8 $\text{\AA}/\text{s}$ leading to a difference between final and expected thickness over the duration of the shuttered period.

From the ToF measurements, hole carrier mobility is calculated as $\mu = L^2/(V \cdot t_T)$, where L is the thickness, V the applied voltage, and t_T is the carrier transit time determined from the measurement [43]. The hole mobility for the four samples can be seen in Fig. 2.4a. When comparing the single deposition to the samples where vacuum was broken, the hole mobility for all fall within the range of 0.11-0.17 $\text{cm}^2/\text{V}\cdot\text{s}$ at 10 $\text{V}/\mu\text{m}$, as previously reported [43, 44, 45, 46, 47, 48]. The process of breaking the vacuum during deposition seems to reduce mobility compared to a single deposition. This reduction is likely due to the samples' exposure to changes in humidity, temperature, and airborne particles when the vacuum is broken. These environmental factors may

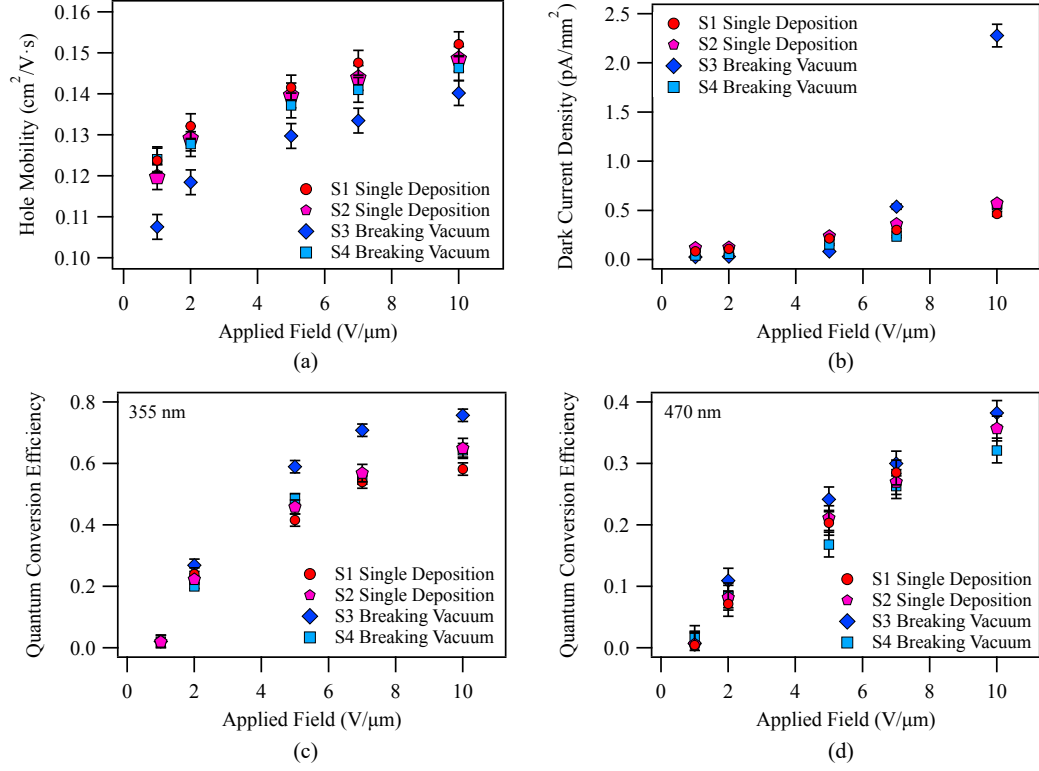


Figure 2.4: (a) Hole mobility for the 4 samples of breaking vacuum study. (b) Dark current density for the 4 samples up to $10 V/\mu m$. (c) QE found from ToF measurements at a wavelength of 355 nm. (d) QE found from dark and photocurrent measurements at a wavelength of 470 nm.

cause a slight decrease in mobility by allowing molecules to interact with and fill shallow traps at the interface.

Fig. 2.4b displays the dark current density of the samples, all of which exhibited a leakage current below 1 pA/mm^2 and were consistent within the margin of error, except for sample S3, which showed an unexpected spike at $10 \text{ V/}\mu\text{m}$. Sample S2 had a slightly higher leakage current compared to the others, likely due to being thinner, which contributes to increased dark current. There is no evidence from the data to suggest any issues arising from breaking the vacuum.

Calculated from our ToF measurements Fig. 2.4c shows the QE of the samples at 355 nm . The samples fall within error of each other at lower fields and follow the same trend across samples. The recorded efficiencies match up with previously reported values [34]. S3 had a slightly higher QE value than others at higher applied fields.

Fig. 2.4d shows the QE at 470 nm from dark and photocurrent measurements. This follows previous works reported values though S4 falls slightly below other samples. Again S3 shows a higher QE than the other samples at a larger wavelength at all applied fields. At this wavelength, stabilized a-Se has a reduced response as it falls along the edge of the mobility gap.

Since S1 and S3 came out to almost the same thickness of $223 \text{ }\mu\text{m}$, we directly compare their behavior in dark and photocurrent measurements. The samples were biased at $10 \text{ V/}\mu\text{m}$ for 10 minutes with Fig. 2.5a showing the photocurrent measurement taken of the samples being exposed with a 1 second, 470 nm LED flash then recovering in dark. S3 has a higher dark current but both samples reach the sample photocurrent

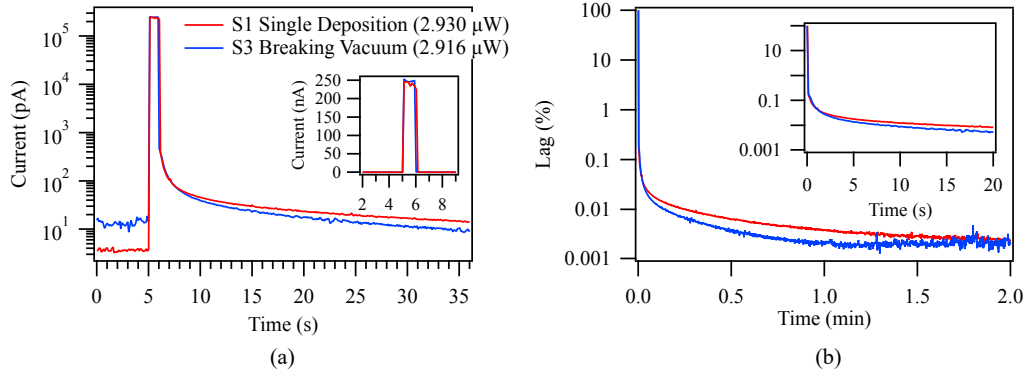


Figure 2.5: (a) Current output for samples S1 and S3 for dark and photocurrent measurements of a 1 second pulse of 470 nm light while biased at 10 V/ μ m. (b) Lag for samples S1 and S3 after being exposed for 1 second by 470 nm at 10 V/ μ m.

with a relatively flat signal that can be seen in the inset. The difference between dark currents is minimal as they differ by roughly 10 pA and the recovery curves only differ by 1 pA.

To gain more information about the recovery period in Fig. 2.5b, we calculated the lag of the samples over 2 minutes post exposure, with the inset highlighting the first 20 seconds. S3 appears to return to its original state quicker than S1, but both recover to less than 0.1% of exposure signal within a few seconds and 0.01% within 10 seconds. Thus both effectively recover quickly and fluctuations can be attributed to instrument noise.

To further explore the variability of stabilized a-Se devices, eight samples with thicknesses ranging from 15 μ m to 245 μ m were characterized. Fig. 2.6a shows the hole mobility of these samples for E-fields between 0-35 V/ μ m. As expected, mobility

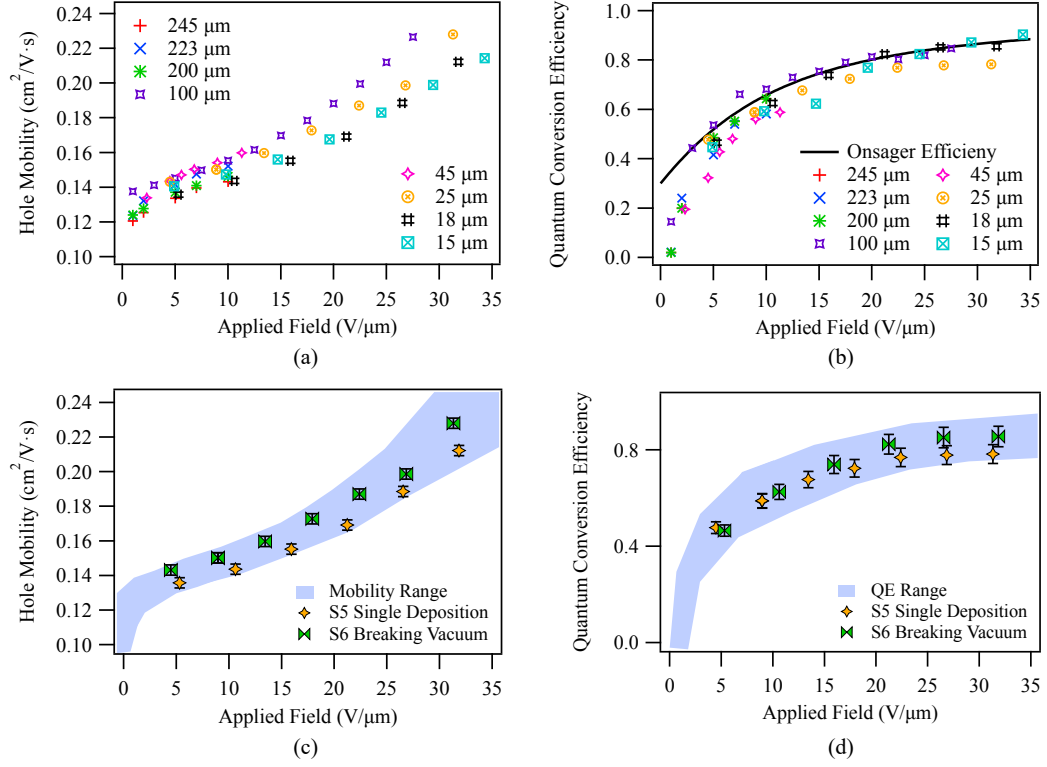


Figure 2.6: a) Hole mobility samples with a-Se thicknesses ranging from 15 μm to 245 μm with applied fields of 0-35 $\text{V}/\mu\text{m}$. (b) QE of samples calculated from ToF measurements. Black line shows the efficiency predicted by the Onsager model. (c) Hole mobility for additional 2 samples in breaking vacuum study, compared to range reported in (a) and in other works. (d) QE for the same samples as (c), compared to range reported in (b) and in other works.

increases with the applied field with a general trend of $\pm 0.02 \text{ cm}^2/\text{V}\cdot\text{s}$ from the average values in line from previous works [43, 44, 45, 46, 47, 48]. There is no clear dependence of hole mobility on thickness of stabilized a-Se [49].

Calculated from ToF measurements, Fig. 2.6b shows the QE of samples at 355 nm. As the field is increased over the sample, we see a rapid increase of QE. The black line representing the Onsager model for 355 nm, shows our devices following the trend at fields above $3 \text{ V}/\mu\text{m}$ [43, 34].

To continue further testing, two additional samples were fabricated; S5, a single deposition of $19.9 \mu\text{m}$, and S6, fabricated from two depositions reaching $22.4 \mu\text{m}$. Fig. 2.6c shows the hole mobility for these two samples, which were able to be biased at higher fields due to the thinner layers. They fall within the approximate range of hole mobility, shaded in light blue [44, 45, 46, 47, 48]. Fig. 2.6d compares QE taken from ToF measurements, and show that each sample fall within the range of QE expected [50, 51, 34, 11].

2.3 Discussion and Future work

The work focusing on the effects of breaking vacuum during the deposition process shows that while there is some variability between samples, there was no significant impact on the direct conversion detector performance in QE, dark current, or recovery from pulsing in the specified applied fields. We see that variations in the hole mobility and QE is not due to a-Se layer thickness and appears consistent as the thickness in-

creases between 15 and 245 μm for samples fabricated in our lab. These discrepancies may be attributed to measurement errors of thickness or minor differences in sample fabrication. For thicker samples, our crucible would need to be refilled before each sample and during breaking vacuum with stabilized selenium pellets. However, thinner samples would not use most of the material in the crucible, meaning that following deposition process would use previously melted stabilized selenium. During fabrication of samples, exposure to air particles should always be limited, nevertheless it does not appear to introduce negative effects on the performance of the devices from 15 μm to 245 μm . This will allow further research to be conducted into thicker layers of selenium and bypass current crucible limitations.

Chapter 3

Selenium Tellurium alloy photodetectors and Hole Blocking Layers

Building upon the investigation of a-Se, this work next explores selenium-tellurium (Se-Te) alloys to tailor the optoelectronic properties of photoconductive materials for indirect-conversion radiation detectors. Selenium-tellurium alloys have attracted considerable interest because their composition can be tuned to modify key material properties, including bandgap, photoconductivity, carrier transport, and crystallization behavior. These characteristics make Se-Te alloys promising candidates for indirect detector applications, where optimizing the balance between sensitivity, charge transport, and device stability is essential.

Our primary interest in the Se-Te alloy is to alter its properties to increase detector responsiveness, improve charge transport, and modify the impact ionization field. The incorporation of Te has been shown to reduce the bandgap and increase

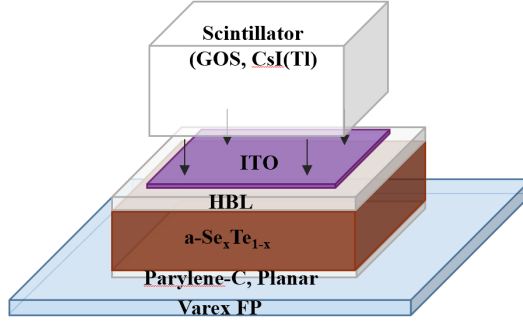


Figure 3.1: Flat Panel Detector Architecture [4]

the film's conductivity, leading to higher leakage currents [23]. Additionally, Te raises the crystallization temperature of the amorphous film, improving its tolerance to Joule heating [27, 28]. However, Te also introduces both shallow and deep trap states, which can reduce charge carrier mobility and shorten carrier lifetimes [29].

3.1 Se-Te Fabrication

Deposition of Se-Te films follows the same procedure as a-Se deposition as described in section 2.1 with slight changes in crucible temperature, deposition rates, and material. Thin films were deposited on ITO/glass slides with hole blocking layers, by evaporation of the Se-Te pellets in a dedicated Se thermal evaporator onto substrates. The substrates were rotated at 40 rpm during deposition and either maintained at room temperature or heated to 60 °C to control film morphology. Deposition rates were maintained between 90 and 105 Å/s at base pressures below 1.5×10^{-6} Torr. Gold top contacts of 100 nm were deposited by e-beam evaporation, utilizing a shadow mask.

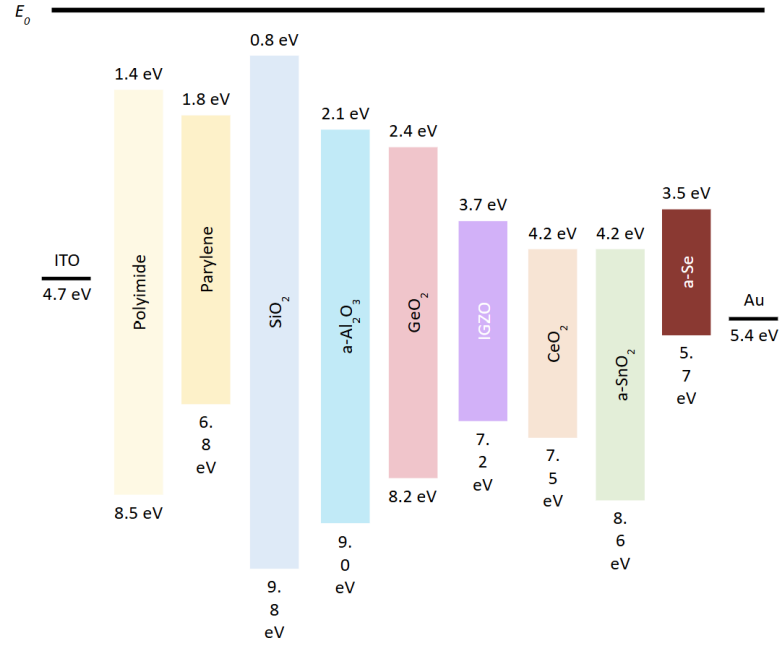


Figure 3.2: Band gap diagram for potential Hole Blocking Layers.

3.2 Hole blocking materials

While Se-Te has some ideal properties, the addition of Te into the system induces a higher leakage current and introduces shallow and deep trap states which decrease the carrier mobility and lifetime of the charge carriers. In order to counteract the effects of high dark current and lag due to trapping states, we can incorporate a hole blocking layer on our samples to increase the mobility of certain charge carriers. We investigated numerous potential candidates each with previous research in similar applications.

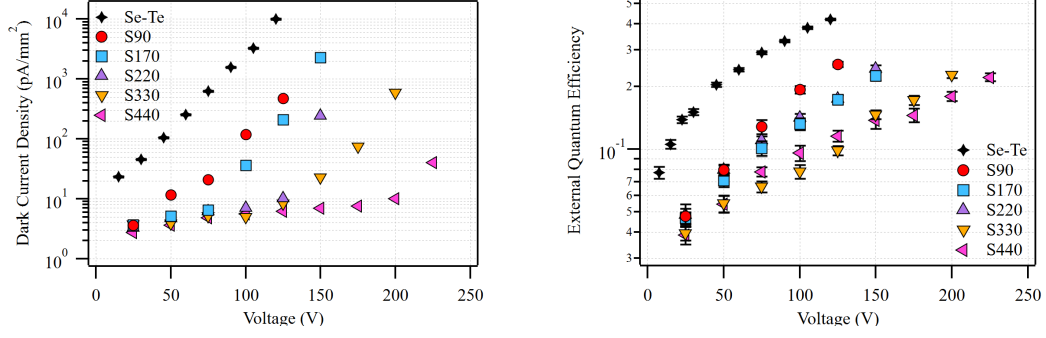
3.2.1 Parylene

Parylene-C and Parylene-N have been investigated as hole-blocking layers for direct detectors of a-Se thicknesses greater than 20 μm , and more commonly exceeding 50 μm . Studies have shown that thick parylene layers ($>5 \mu\text{m}$) lead to charge buildup, reduced signal, and increased lag and ghosting due to their highly insulating nature at these thicknesses [52]. In contrast, thinner parylene layers ($<5 \mu\text{m}$) have been successfully used as HBLs without reported negative impacts on thermal performance or signal response [53, 54]. Our group has demonstrated that submicron ($<1 \mu\text{m}$) parylene layers can be effectively used in indirect detectors with 15 μm a-Se [4]. At these smaller thicknesses, charge buildup is minimal, but there is an increased leakage current instability. Based on these characteristics, parylene was selected as the leading candidate for the hole-blocking layer.

To fabricate samples incorporating a parylene layer, the glass slides are first cleaned following the procedure described in Section 2.1. The parylene coater is then prepared by ensuring the chiller is clean and removing any residual material from the vaporizer (furnace port). A large piece of ultra high vacuum (UHV) aluminum foil is shaped into a “boat” to hold the parylene-C, and approximately 0.45-0.50 g of material is weighed into the boat to typically produce a film thickness of 100–200 nm. Larger amounts of parylene-C produce correspondingly thicker films. The loaded boat is placed into the furnace port, and the cleaned substrates are mounted on the rotating sample stage inside the deposition chamber. Once the chamber reaches the required vacuum

	S90	S170	S220	S330	S440	S780
Parylene thickness	$90 \pm 20nm$	$170 \pm 15nm$	$220 \pm 10nm$	$330 \pm 10nm$	$440 \pm 10nm$	$780 \pm 10nm$
SeTe thickness	$3.2 \pm 0.05\mu m$	$3.2 \pm 0.05\mu m$	$2.85 \pm 0.05\mu m$	$2.85 \pm 0.05\mu m$	$2.85 \pm 0.05\mu m$	$13.35 \pm 0.05\mu m$
SeTe:HBL ratio	35.6:1	18.8:1	13.0:1	8.6:1	6.5:1	17.1:1

Table 3.1: Sample information with varying thickness of Parylene and SeTe layers [4]



(a) Dark Current Density vs Voltage (b) External Quantum Efficiency vs Voltage

Figure 3.3: Data for parylene samples S90-S440 [4]

pressure, the automated deposition process begins. After deposition is complete, the parylene film thickness is measured before the samples are transferred to the thermal evaporator for Se-Te deposition.

A series of samples with varying parylene thicknesses were fabricated to determine the optimal hole-blocking layer thickness for indirect detection applications. Details of the fabricated samples are provided in Table 3.1. With the exception of sample S780, all samples contained a photoconductive layer between 2.85 and 3.2 μm thick, allowing the effects of parylene thickness to be evaluated independently of photoconductor thickness.

Samples were characterized using dark and photo-characterization methods as

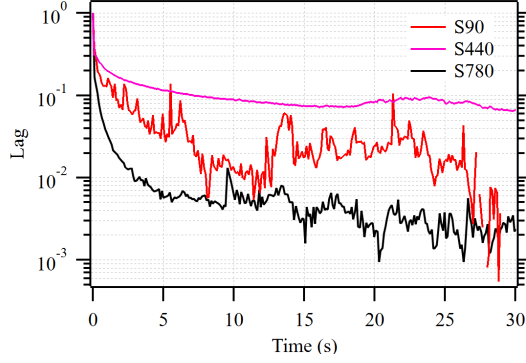


Figure 3.4: Comparison of Lag in Parylene samples [4]

described previously in Section 2.1 with an LED producing a wavelength of 533 nm.

In Fig. 3.3a, it can be seen that the leakage current reduces with increasing parylene layer thickness and higher voltages can be achieved. In Fig. 3.3b, EQE reduces with increasing parylene thickness at each voltage (thus max EQE reduced). Even though the parylene layer reduces the overall leakage current, it decreases the photocurrent detection as well, impacting the EQE.

In comparing the thinner and thicker samples of parylene in Fig. 3.4, we see that thin parylene has minimal charge build-up, but increased leakage instability. As the thickness increases so does a build-up of charges.

3.2.2 Cerium Oxide Quantum Dots

Cerium oxide (CeO_2) Quantum Dots (QDs), provided a interesting option for a HBL [55, 56, 57]. Previous work shows ultralow dark current in a-Se photodetectors using the solution processed CeO_2 QDs [58]. The application occurs at room temperature onto 15 μm a-Se with a 100 nm SiO_2 buffer layer. They exhibit extremely low

dark current and the ability to reach avalanche gain. While extremely promising, the deposition of the QDs is done through spin coating which is very difficult for large area applications.

3.2.3 Silicon Dioxide

Silicon Dioxide (SiO_2) has a very large bandgap that creates a strong energy barrier for holes, effectively suppressing unwanted carrier injection and minimizing dark current. While typically used as an insulating layer, at nanometer thicknesses, electrons can still tunnel through the layer while holes remain blocked [59, 60]. This combination of strong hole blocking, low leakage current, excellent stability, and easy fabrication make it a promising candidate.

SiO_2 films were deposited using the same e-beam evaporator and deposition procedure described for Au deposition in Sec. 2.1. The primary differences were the use of SiO_2 as the source material and a shadow mask designed to deposit the film across the active device area while leaving the sample edges free to maintain electrical contact with the ITO. Film thickness was monitored throughout the deposition using the quartz crystal microbalance (QCM) integrated into the e-beam evaporator, providing an estimate of the deposited SiO_2 thickness.

3.2.4 Aluminum Oxide

Aluminum Oxide (Al_2O_3) is an excellent insulating and passivation layer with a very wide bandgap, so it can strongly suppress leakage current and surface recombina-

tion [61, 62]. However, because it is highly insulating, charge transport through Al_2O_3 can be limited unless the layer is made extremely thin. Deposition of Al_2O_3 films was done in same process as SiO_2 .

3.2.5 Tin Oxide

Tin oxide (SnO_2) has a high electron mobility compared to more insulating materials such as SiO_2 and Al_2O_3 . SnO_2 is a strong hole blocking layer because its wide bandgap and deep valence band create a large energetic barrier for holes, preventing them from reaching the electrode and reducing interfacial recombination and dark current [63]. Due to the mismatch of conduction bands between SnO_2 and a-Se, an additional HBL buffer was included in some samples to provide better charge transport. Deposition of SnO_2 films were done in the same process as SiO_2 .

3.3 Comparison HBLs

Taking dark and photo current measurements (as described in Sec. 2.1) of SiO_2 , Al_2O_3 , SnO_2 and combination SnO_2 with both SiO_2 and Al_2O_3 , we can observe the behavior of the HBLs.

Testing photo current density for each type of hole blocking layer at the same applied electric field (with the exception of SnO_2), we can compare the potential hole blocking layers to give a clearer picture as to the best one for our applications. SnO_2 combined with both Al_2O_3 and SiO_2 in Fig. 3.5cd, reduced the dark current by a wide margin, but also reduced the photocurrent. This decrease in the photocurrent detected

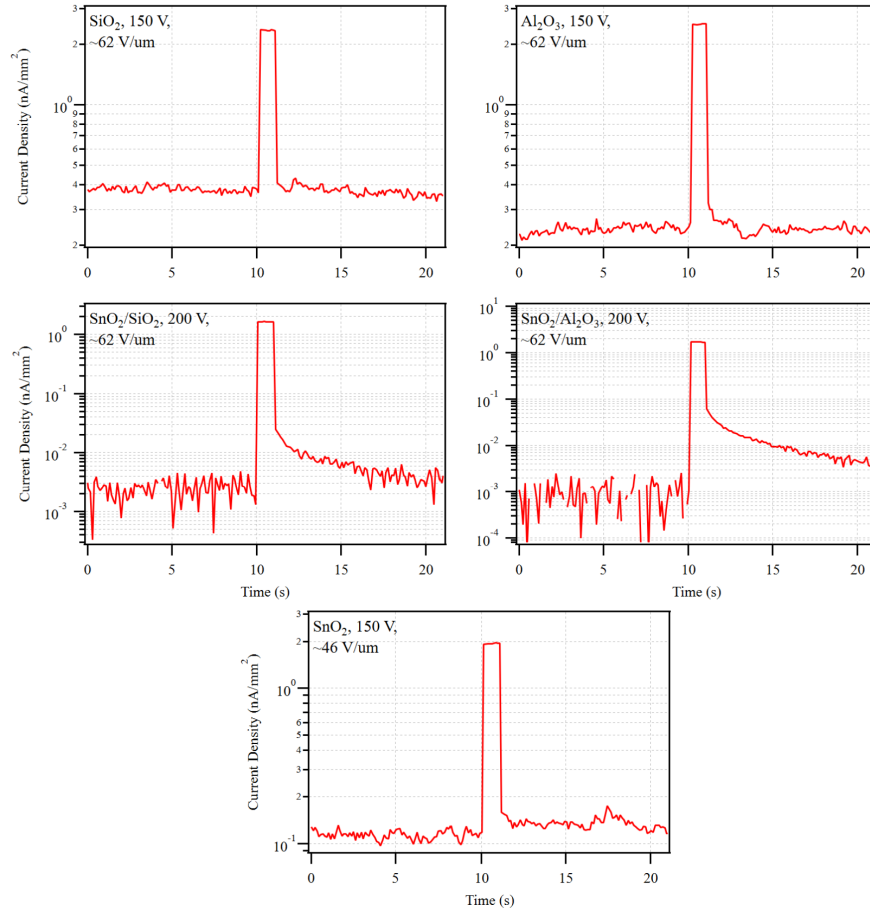


Figure 3.5: Current Density of samples with each hole blocking layer. (a) SiO₂, (b)

Al₂O₃, (c) SnO₂/SiO₂, (d) SnO₂/Al₂O₃, (e) SnO₂

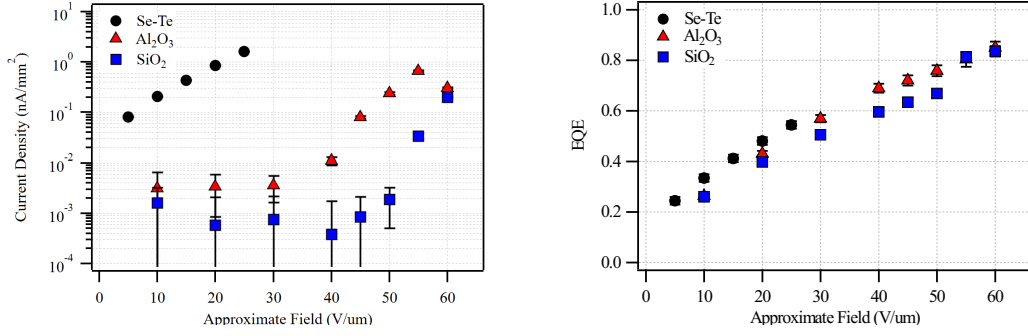
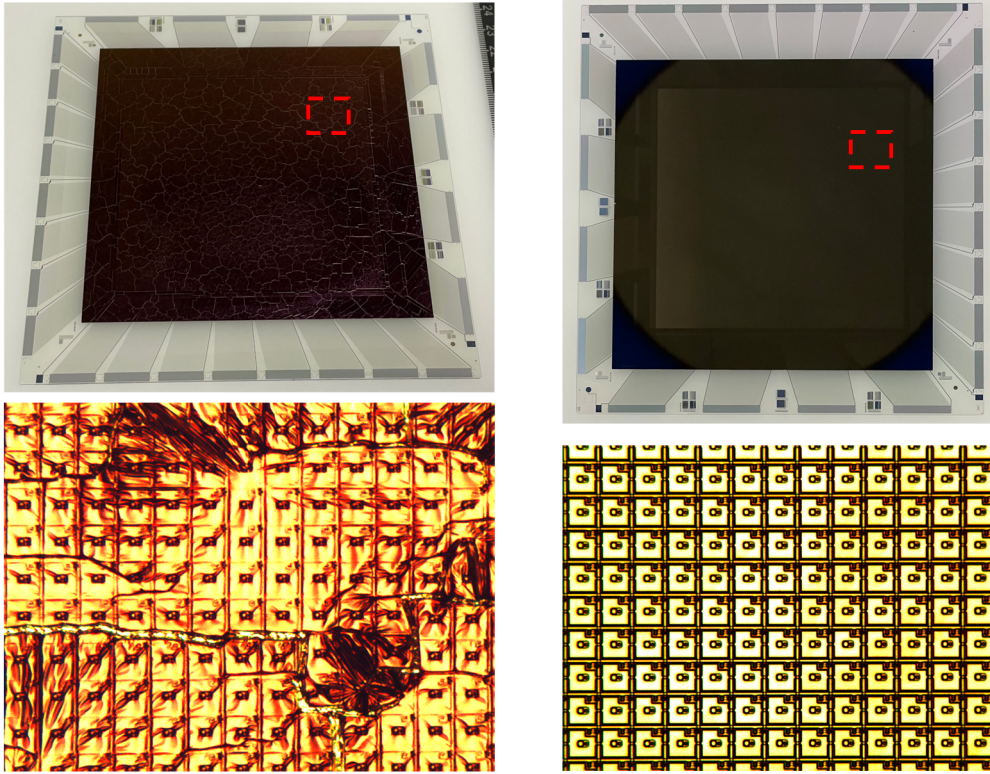


Figure 3.6: Comparison of the top HBL contenders for Dark current and EQE

lowers the EQE and increases the lag, similar to the issues seen with parylene. SnO_2 in Fig. 3.5e had trouble maintaining a field higher than $46 \text{ V}/\mu\text{m}$ before joule heating caused crystallization in the sample. Electrical characterization identified Al_2O_3 and SiO_2 as the most promising HBLs. Subsequent large-area fabrication studies demonstrated the superior mechanical compatibility of SiO_2 with Se-Te films, leading to its selection for further device development.

After seeing that Al_2O_3 and SiO_2 are the best options thus far, we took more data to compare the current density and external quantum efficiency for each type. As can be seen in Fig. 3.6a, SiO_2 appears to have lower leakage current at all applied biases compared to Al_2O_3 . Both show a significant reduction compared to Se-Te alone and show a rapid increase in dark current density as applied e-field increases. While looking at the EQE in Fig. 3.6b, there is a slight reduction from Se-Te to samples with HBL. Al_2O_3 and SiO_2 exhibit similar trends in the graph with Al_2O_3 overall higher until the highest voltages.

Fabrication of large flat panel detectors (FPD) with the additional hole block-



(a) Flat Panel Detector made with Al_2O_3 (b) Flat Panel Detector made with SiO_2

Figure 3.7: Flat Panel Detectors with zoom in of red square.

ing layers of Al_2O_3 and SiO_2 were fabricated based on the structure from Fig. 3.1. Fabrication of the large area flat panels was done in top-down order as opposed to the bottom-up sample deposition. Gold contacts are embedded in the FPD, with subsequent depositions of the photodetector layer, HBL, and conductive ITO layer. Due to this difference in deposition order, issues arose due to the mismatch in thermal expansion between Se-Te and Al_2O_3 . This caused cracking across the surface of the Se-Te seen in Fig. 3.7a. Conversely, in Fig. 3.7b SiO_2 had no visible defects after deposition with a clear surface. As this fabrication method demonstrated that SiO_2 is a suitable hole blocking layer for large-area deposition, subsequent work focused on SiO_2 as the leading HBL candidate.

Further experimentation with SiO_2 as the hole blocking layer, shown in Fig. 3.8a, revealed linear photocurrent behavior from intensities $0.5 - 4.5 \mu\text{W}/\text{cm}^2$ in regions below impact ionization, with the higher applied fields increasing at a faster rate. The pulse shape when exposed to light shows fast rise and fall times with a steady signal across all applied E-fields in Fig. 3.8b. At $60 \text{ V}/\mu\text{m}$, the dark current is $0.2 \text{ nA}/\text{mm}^2$ higher than the other applied fields, but shows a $0.2 \text{ nA}/\text{mm}^2$ increase when irradiated compared to $55 \text{ V}/\mu\text{m}$. The recovery time or lag after light exposure in Fig. 3.8c, is reduced with increasing E-field. At $50 \text{ V}/\mu\text{m}$, there appears to be a small residual charge that remains after short period, but higher applied fields demonstrate a faster recovery time. Looking at repeated exposure of light on the sample in Fig. 3.8d, there appears to be minimal ghosting effects from charge buildup. While there is no discernible difference for lower applied fields, the dark and photo current for $60 \text{ V}/\mu\text{m}$ increases by

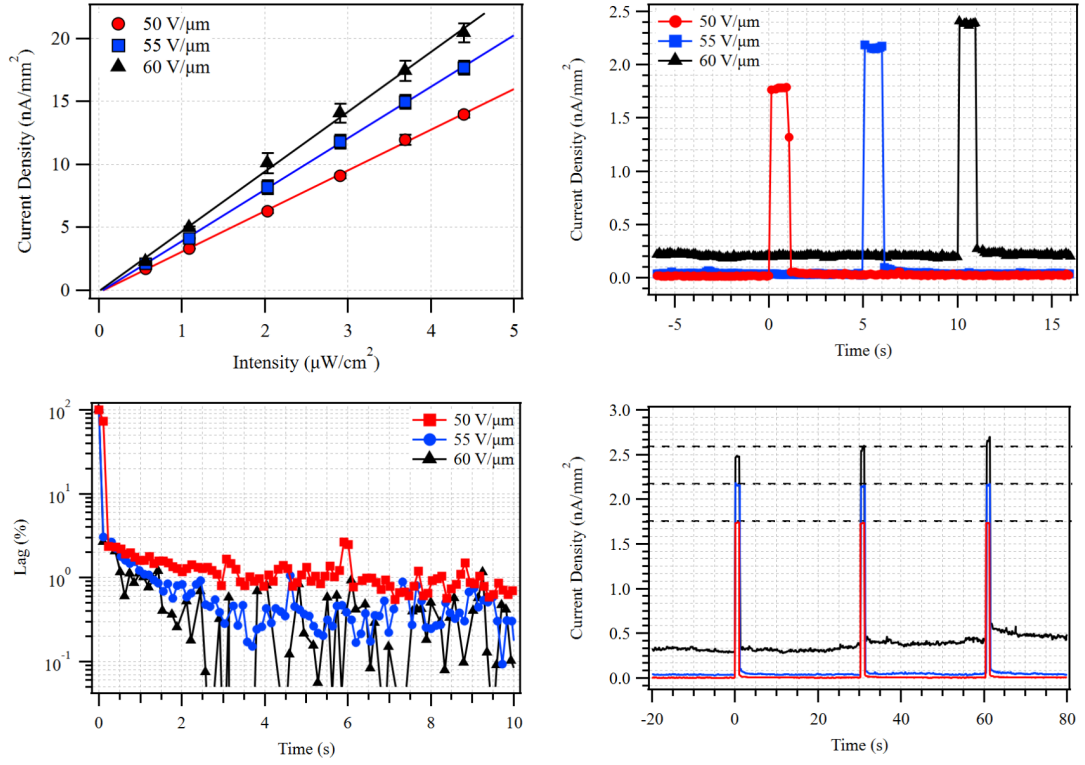


Figure 3.8: (a) Linearity: Photocurrent density vs Intensity. (b) Pulse shape: Photocurrent density vs time. (c) Lag vs time. (d) Ghosting: Multiple exposures of light.

approximately 0.2 nA/mm^2 after three exposures over the course of 60 seconds. This small increase shows trace amounts of charge build up but overall has little increased instability.

SiO_2 met the primary requirements for a top layer HBL, exhibiting high EQE, acceptable leakage current, and minimal lag and ghosting. The devices maintained high sensitivity while achieving a dark current density below 2.9 nA/mm^2 . In addition to its electrical performance, SiO_2 is transparent at interested wavelengths, compatible with large-area deposition, low fabrication cost, and low temperature processing. The main limitation is a higher leakage current. However, this can potentially be minimized by improving the quality of the Se-Te layer.

Further investigation can be done into the other HBLs by means of other fabrication methods. In comparison with other works, it can be seen that e-beam deposition appears to introduce many defects with high levels of trapping and increased lag. Using sputtering for SnO_2 may allow for a reduction in defects that promote conductivity, allowing for higher E-fields to be achieved before thermal crystallization occurs. The same may be said for Al_2O_3 as this new method of fabrication has the potential to sidestep the thermal expansion mismatch with the Se-Te layer. Sputtering CeO_2 would also avoid the large area deposition issue making it a potential candidate again.

Chapter 4

Oxidation in Se-Te alloys

SeTe devices fabricated in our lab using pre-alloyed evaporation pellets initially demonstrated performance consistent with previously reported results for similar alloy compositions [43, 64]. Over time, device performance progressively degraded, observed by increased failure rates, short-circuiting upon bias application, dark current instabilities at electric fields above 20 V/ μm , and current surges between 40 and 60 V/ μm . These current spikes indicated field induced electrical breakdown rather than photogenerated charge transport.

Post fabrication of the samples, degradation of the samples were visible by eye as we were able to see surface holes going through the sample by shining a light directly underneath the sample. Once noticed we began looking into SeTe degradation and oxidation pre and post fabrication. These holes can be seen in Fig. 4.1.

Subsequent optical and scanning electron microscopy (SEM) revealed material gaps, raised edges, and surface roughening consistent with gas evolution and ruptures.

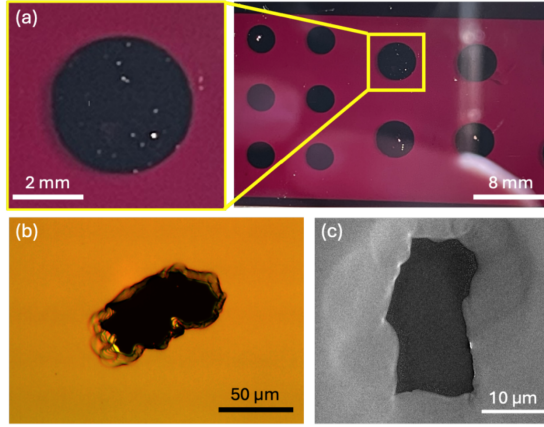


Figure 4.1: (a) Photo of an SeTe device, lit from behind. (b) Microscopic image of a film rupture. (c) SEM image of another hole in the surface of a Se-Te film.

Visual inspection of previously fabricated devices also showed that the amount of these defects increased with sample age. No correlation was observed between defect formation and fabrication parameters such as substrate temperature, deposition environment, or film thickness. Any other process modifications such as substrate cleaning, deposition rate variations or extended pre-deposition evaporation did not appear to mitigate the formation of defects. These defects were observed exclusively in Te containing samples. Since the only consistent variables to all affected devices was prolonged environmental exposure, we expect that to be the primary reason responsible for defect formation.

The original material (Batch A) was stored in large airtight containers containing ambient air, exposing the remaining material each time material was removed for deposition. Over time, condensation was observed inside the containers, raising concerns and prompting the purchase of new source material. The new material was supplied in smaller, individually sealed packages under an inert gas to better preserve

its condition and allow the effects of storage to be evaluated. To determine the role of storage conditions and how long-term storage had degraded the original material, a comparative study was conducted using devices fabricated from both the original and newly acquired materials.

4.1 Materials and Fabrication

The new Se-Te evaporation pellets (Batch B) were purchased from William Rowland with a composition of 70% Se and 30% Te. According to the manufacturer, the pellets were fabricated in an argon glovebox by first producing a high-purity molten Se-Te alloy, granulating it through a sieve into deionized water and vacuum dried at low temperature to prevent crystallization. The original batch of source material was supplied in a large plastic container filled with ambient air and was stored in the dark when not in use. The new batch was packaged into 100 g argon sealed bags and was opened only when used. Upon opening, the pellets were transferred directly to the thermal evaporator crucible for deposition. Any remaining material was left in the crucible under vacuum and only exposed during preparation for the next deposition.

Thin films of Se-Te were prepared similarly to previous depositions described in section 3.1, yielding films of approximately 3 μm thickness. Devices were fabricated on ITO/glass substrates with and without a 30 nm SiO_2 hole blocking layer, deposited by electron beam evaporation. Gold top contacts of 100 nm were deposited by electron beam evaporation, utilizing a shadow mask to pattern 3 or 4 mm diameter devices.

For imaging in later sections, pellets of Se-Te were cut using a razorblade to initiate fractures, allowing natural crack propagation through the material to minimize mechanical damage. Images that were taken were in areas unaffected by direct razorblade contact.

Accelerated aging of the newer material Batch B, was done under high-humidity and elevated temperature conditions. Pellets were mounted on SEM stubs using carbon adhesive tape, imaged, and then placed in a polystyrene sample box. Deionized (DI) water was introduced to maintain humidity without direct liquid contact with the samples. The box was closed and kept on a hot plate at 50 °C for three weeks, besides the time taken for testing, to accelerate environmental interaction. The stub was removed and pellets imaged by SEM in the same location each week after the first imaging was performed, and DI water was added as necessary to maintain a humid environment. During the third week, no additional water was introduced to evaluate the effect of reduced humidity on surface modification.

4.2 SEM-EDS

Surface and cross sectional images of the pellets and films were taken using a Thermo Scientific Scios 2 DualBeam SEM operated at 10-20 kV and 25-80 pA. Elemental analysis was performed using Oxford Instruments AZtecLive Ultim Max 100 energy dispersive spectroscopy (EDS) detector with the electron beam operating at 10 kV and 0.8 mA beam current. Transition energy levels were reported from provided analysis

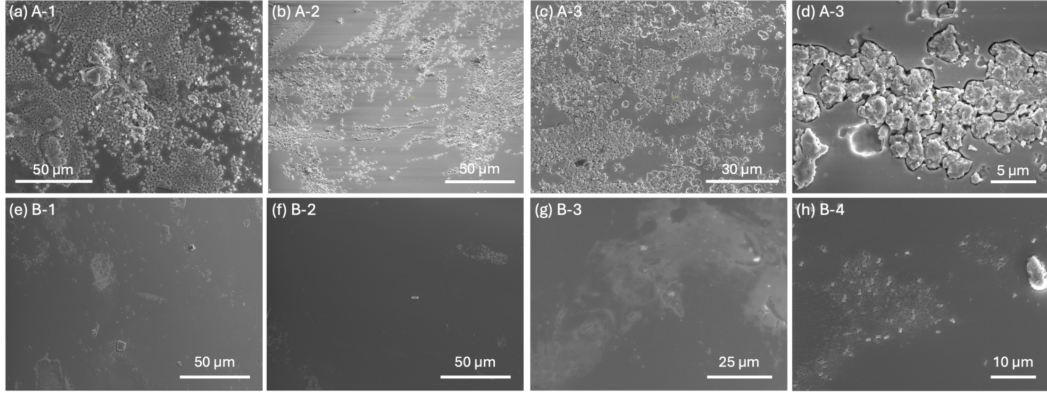


Figure 4.2: SEM imaging of pellets from batch A (a-d) and B (e-g)

procedure and verified from database values [65, 66].

Pellets are roughly 1-4 mm across with variations in size and appearance. Se-Te pellets with a relatively flat surface were chosen for SEM imaging. Fig. 4.2 a-c show images from three different pellets from Batch A. While there are some smoother areas, defects are clearly visible across the surface. Upon closer inspection the same pellet from Fig. 4.2 c as seen in Fig. 4.2d, there is a more structured morphology, potentially indicating a crystal formation at the surface of the material. This behavior was noticed across all Batch A sample pellets. Fig. 4.2 e-h showcase Batch B across four different pellets. The surfaces are smoother with occasional small areas of scratches, semi fused patches or slightly raised areas on the surface, likely where original Se and Te material didn't fully combine. These features lack the roughness seen in Batch A and are relatively uniform and smooth.

In order to better observe the defects occurring at the surface of the pellets in Batch A, the pellets were cleaved to get a cross-sectional image to see the depth

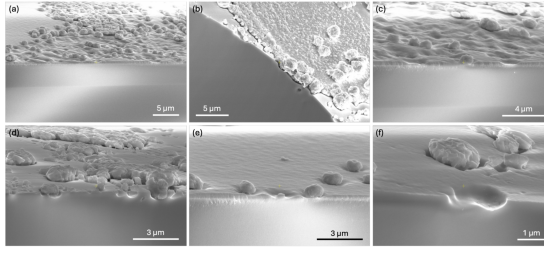


Figure 4.3: Cross-section SEM of Batch A pellets

of protrusions. Fig. 4.3 shows images captured of multiple pellets. In all cases, any distortions appear to occur only on the surface of the pellets. As can be clearly seen in Fig. 4.3 d-f, the protrusions have a stub in which the rough material forms, separating from the surface. In Fig. 4.3f, the indent at the edge shows where the cleaving removed the top of the floret. This implies that the contamination at the surface is extruding elements from the pellet to form new oxidation compounds.

As atmospheric water is the most common contaminant that the pellets were exposed to in Batch A, we accelerated the aging process of Batch B using a high-humidity, high temperature environment. After an initial SEM imaging of fresh material, pellets were aged using methods described in Sec. 4.1, and measured in the same location, each week for three weeks as seen in Fig. 4.4. There appear to be some slight protrusions before the aging cycle begins, likely due to semi-fused patches or slightly raised areas on the surface, likely where original Se and Te material didn't fully combine. After one week in a high humidity environment, blemishes begin to appear on the surface of the pellet as seen by the discoloration in the image. The surface appears slightly rougher in these areas. After a second week, a large increase in protrusions appear

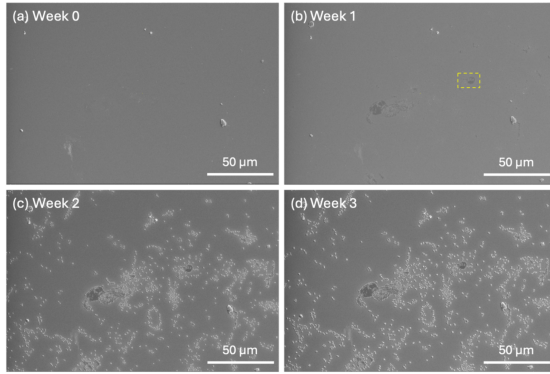


Figure 4.4: SEM images of Batch B pellet in same position to showcase aging over multiple weeks. Yellow box in (b) highlights area of focus in Fig. 4.5.

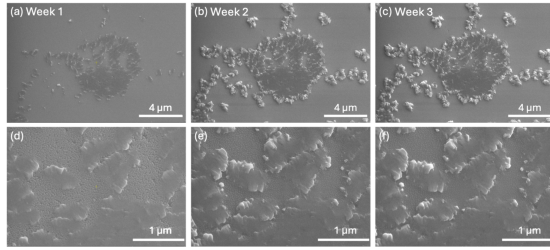


Figure 4.5: SEM of blemish in yellow box in Fig. 4.4b

across the surface of the pellet. They appear to cluster together or follow an existing protrusion or blemish. After the third week with a reduced humidity, the surface doesn't appear to have gained any further defects but has made the current ones more defined.

We take a closer look at a blemish in Fig. 4.5, that appeared after a week of aging highlighted in the yellow box in Fig. 4.4b. The bottom images focus on the region at the centers of the top images. From week 1 to week 2 of aging, we see the protrusions rise more from the surface and become much sharper along the edges. The third week, we don't see much development with slightly more defined edges of the protrusions.

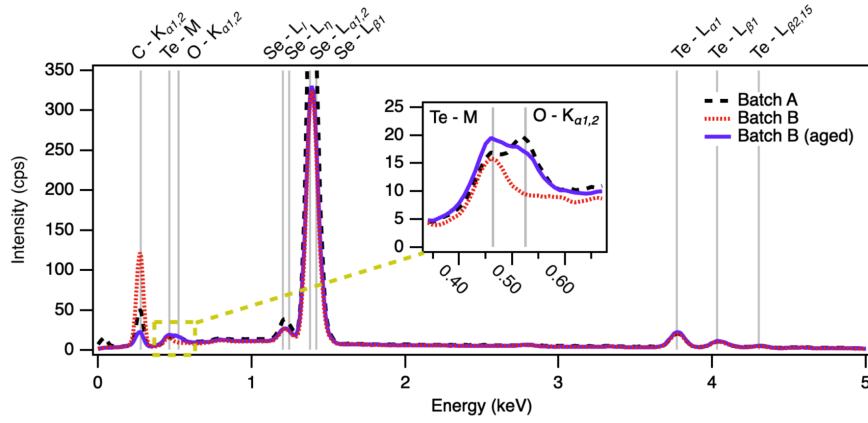


Figure 4.6: EDS spectra of Batch A, B and B aged. Grey vertical lines mark transitions assigned to each peak.

In comparing the images of Batch A to Batch B, we can see that there are less defects appearing on the aged Batch B with more jagged and smaller protrusions. They appear to sit on top of the surface, as opposed to being slightly embedded in the surface as they are in Batch A and lack the floret appearance. This is likely due to our aged Batch B, being in the early stages of degradation or accelerated interactions induced compared to the slow degradation of Batch A.

Elemental compositions determined by EDS for each batch can be seen in Table 4.1. Fig. 4.6 shows a typical spectrum taken from each batch. Peak identification shows clear spectral lines from Se, Te and carbon (C). The latter occurs from contamination during mounting to the carbon adhesive tape. All three Batches show a subtle peak from Te-M transitions, highlighted in the inset. Batch A and B (aged) pellets show a shoulder encompassing 0.525 keV, the $K_{\alpha 1,2}$ transition for oxygen, unlike the spectra of Batch B. The composition places this value $0.6 - 2.0wt.\%$ depending on area scanned

Sample	Se	Te	O
Batch A	75 ± 2	14.7 ± 2	0.4 ± 0.1
Batch A (cross-section)	69.2 ± 0.4	29.8 ± 0.4	1.1 ± 0.3
Batch B	70.2 ± 1.4	19.8 ± 1.4	-
Batch B (aged)	69.9 ± 0.8	28.2 ± 0.8	1.9 ± 0.4

Table 4.1: Weight concentration of elements detected in each pellet by EDS listed as percentage of total values. Values averaged over minimum of 4 measurements, with varying points on sample and across multiple pellets.

and definition of the oxygen shoulder in fitting. Obtaining a definitive value for any sample was difficult, given the limited number of peaks distinguishable in the EDS for oxygen and the inhomogeneous oxidation of the pellets. However it is clear that oxygen is present in Batch B (aged) and not present in Batch B.

To compare Batch A and B in the setting we typically use to characterize our samples, we fabricated films of both batches and conducted an SEM of both a week after fabrication in Fig. 4.7 a and b. The film from Batch A showed ruptures in the film, and over the course of a few months, developed crystal branches from the rupture along the surface. The formation of crystalline t-Se nanoparticles with exposure to oxygen presents as needle-like crystals [67, 68]. After six months, the film surface was covered in a variety of patterns. After one week, no ruptures were found on the Batch B film. Imaging five months after fabrication yielded a smooth surface with some particulates and scuff marks on the surface, likely due to handling and repeated testing.

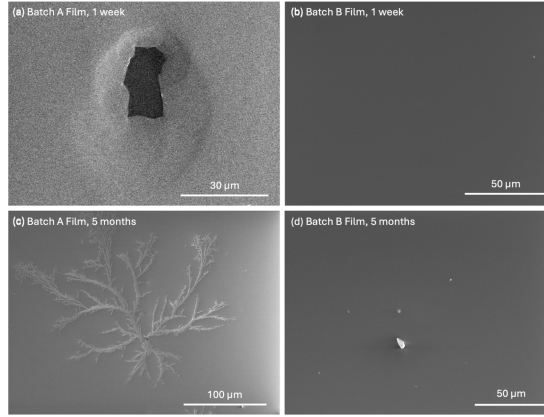


Figure 4.7: SEM of Se-Te film surfaces for devices made from Batch A and B.

4.3 XRD

X-ray diffraction (XRD) was performed on Batch A, B, and B aged pellets to compare chemical composition. Pellets were ground into coarse grains and mounted on a rough quartz substrate designed for powder diffraction. Measurements were taken by a Rigaku SmartLab X-ray diffractometer over a 2θ range of 10° to 70° , with a scan rate of $1^\circ/\text{min}$, step size of 0.02° , and source parameters of 45 kV and 200 mA.

The powder XRD of the ground pellets for all batches can be seen in Fig. 4.8. Batch B shows an amorphous structure with broader peaks at 27° and 51° , similar to that seen in a-Se, and no sharp peaks. Batch A shows two small peaks among the amorphous behavior at 23.3° and 29.1° . Batch B (aged) matches these peaks with the one at 29.1° having a sharper appearance and higher differentiation. In both cases we expect any peaks apparent to be small as the defects were on the surface of the material, and thus the amount of crystalline material will be small in comparison to the rest of

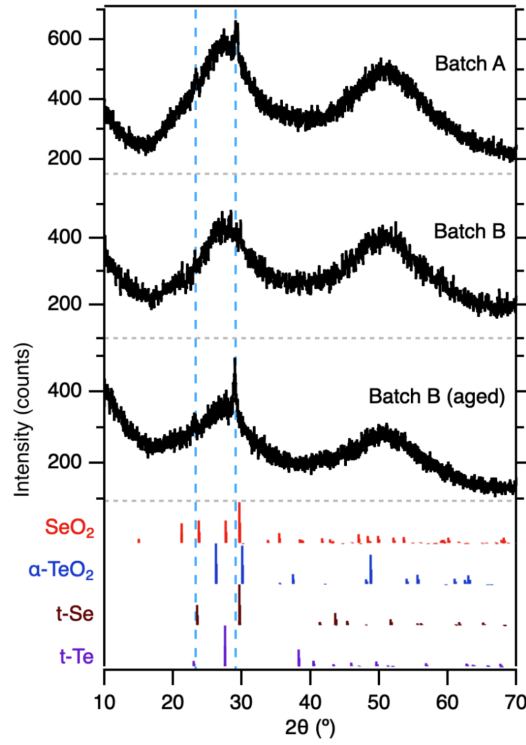


Figure 4.8: XRD spectra for crushed pellets. Dashed blue lines highlight peaks formed in Batch A and B (aged) samples. Reflections for t-Se (ICSD-40018), t-Te (ICSD-161690), TeO₂ (ICSD-56004), and SeO₂ (ICSD-72366) are shown in the bottom quarter of the plot.

the amorphous pellets.

The crystalline compounds involving Se, Te, O, and H were explored through peak matching in the ICSD database. Reflections for the most common structures of crystalline t-Se, t-Te, TeO₂ and SeO₂ [69, 70, 71, 72] are shown at the bottom of the plot in Fig. 4.8. The peaks show consistent alignment with t-Se. For SeO₂, the (201) reflections around 27.7° have similar intensity to (210) reflections around 23.8°, but don't

appear in the scans. Oxygen may be present in an amorphous form of SeO_2 or TeO_2 , inducing crystallization and extrusion of Se at the surface.

4.4 XPS

X-ray photoelectron spectroscopy (XPS) was performed by Eurofins EAG Laboratories on Batch A, B, and aged B pellets. Characterization was done on a Thermo K-Alpha Plus XPS spectrometer with a monochromated Al $K\alpha$ X-ray source. Survey scans were collected over a 400 μm spot size from 1350 to 0 eV binding energy at a pass energy of 200 eV, with 1 eV step size and a total of 4 scans with a scan rate of 20 eV per second. High resolution scans were collected with a pass energy of 50 eV and step size of 0.1 eV for Se 3d, Te 3d_{5/2}, O 1s, and C 1s + Se LMM spectral ranges. A charge correction using the Se 3d_{5/2} energy of 55.1 eV was implemented and the data was analyzed using a nonlinear least squares curve fitting to the high resolution spectra. Characterization only occurs within the top 5-10 nm of the surface.

Samples from all three batches were sent for XPS analysis to gain a more accurate concentration of oxygen and bonding structures on the surface of the pellets. Fig. 4.9 shows the full scan of the three batches. Table 4.2 shows the atomic concentrations of detected elements in each batch, seen to primarily be Se, Te, SeO_x and TeO_x . There appears to be a preference of oxygen bonding with Te over Se. Moderate levels of carbon were found on Batch A and some on Batch B (aged), likely due to organic contamination from air exposure as Batch B has none. However all batches showed the

Sample	Se, in total % Se		Te, in total % Te	
	Se ⁰	SeO _x	Te ⁰	TeO _x
Batch A	99	1	80	20
Batch B	98	2	83	17
Batch B (aged)	85	15	3	97

Table 4.3: Chemical states of each element.

Se and Te were predominantly found as Se⁰ and Te⁰. Oxygen did not show presence in any states besides that of metal oxides, with a large majority bonded to Te atoms. While some oxide formation occurs with Se in Batches A and B, it is a very small amount. Batch B (aged) shows a much higher level of oxidation, with a majority of Te states, almost fully converting the surface Te atoms to oxides. A higher level of SeO_x bonds are present in Batch B (aged) likely due to a more developed defect space and opportunities for interactions. These results show a clear picture of what oxides are formed in the pellets and how the structure is altered.

In XPS, oxygen did not show presence in any states besides that of metal oxides, and the majority of bonds were with Te atoms across all samples. The lack of visibility in the XRD data of Fig. 4.8 implies that these oxides are amorphous.

4.5 Further characterization

In order to understand the effects of oxygen contamination in photodetectors, devices were fabricated from both batches. Se has a melting point of 221°C and Te has a melting point of 449.5°C, their alloy falls somewhere in between. To ensure amorphous and high quality films, the deposition rate was held around 90 Å/s in temperatures be-

tween 290-330°C for evaporation. This temperature range encompassed the sublimation temperature for SeO_2 at 315°C. It avoids the sublimation temperature for TeO_2 though. This resulted in films with a Te concentration of 11-16 wt.%, and a small concentration of O in Batch A 0.1-0.6 wt. %. The devices fabricated were of uniform thickness with Batch A having a Te concentration of 11.6 wt.% and Batch B had a concentration of 16.3 wt.%. The variation in Te concentration arises from the difference in Se and Te melting temperatures, as Se will evaporate at a faster rate than Te. Thus the material remaining in the crucible will have a higher Te concentration. Examination by eye of the remaining material in the crucible from batch A once cooled showed a dull, yellow coloring on the surface, whereas the batch B material remained a dull, dark silver. This is consistent with the presence of oxides in Batch A, as SeO_2 and TeO_2 are reported as white solids with some yellow discoloration [73, 74, 19].

Looking at the SEM images of the films in Fig. 4.7, the crystallization of Batch A after a period of 5 months from fabrication is further indicative of the formation of crystalline Se material. Se has been shown to present as needle-like crystals when formed as t-Se nanoparticles or h-Se with exposure to oxides [67, 68]. The underside of the films as seen through the transparent substrate, had no observable changes, while the top, air exposed surface went from a glassy film to patches of dull, rough areas until that coated the full film. This shows crystallization on the air exposed surface and not through the depth of the film. The presence of defects in the pellets in Batch A likely contributes to the formation of SeO_2 crystals.

The presence of oxides in small quantities of the bulk of the film may not have

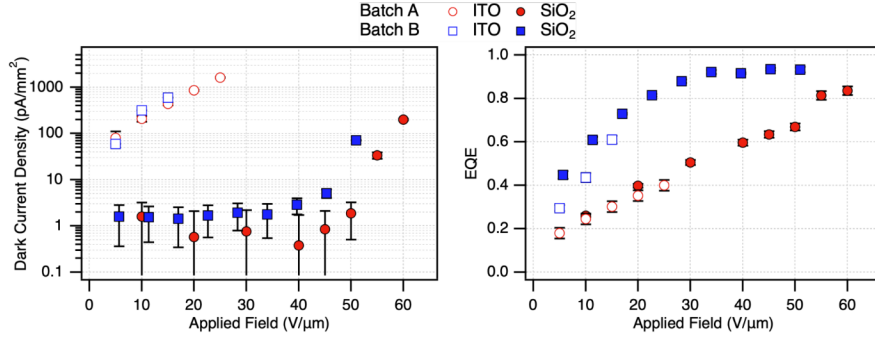


Figure 4.10: (a) Dark current and (b) external quantum efficiency of devices made from Batches A and B, with and without an SiO₂ blocking layer.

a strong effect on the performance of a photodetector. Devices fabricated from Batch A and B were characterized for dark and photocurrent at a range of applied fields using the same methods from previous sections to test this. The external quantum efficiency was calculated from $\eta = \frac{(I_P - I_D)/e}{P_\lambda/E_\lambda \cdot N_A}$ where I_P is average photocurrent, I_D is the average dark current, e is elementary charge, P_λ is the power incident on the device, E_λ is the energy of the wavelength incident on the device, and N_A is Avogadro's number. Devices were illuminated with monochromatic light of $\lambda = 533\text{nm}$ at intensities $0.4\text{-}0.7\mu\text{W}/\text{cm}^2$ for 1 s. Each measurement cycle consisted of a voltage bias ramping followed by a dark settling period, then the illumination and another dark recovery period. Three devices were characterized to ensure consistency in performance with their respective thicknesses found by stylus profilometry (Dektak 3).

Fig. 4.10a shows the dark current density from Batch A and B devices with and without a 30 nm SiO₂ hole blocking layer. Devices made from batch A show a slightly lower dark current, likely showing an increased effective resistivity. As ox-

oxides are typically insulators, we expect there to be an increased resistivity in the SeTe films and suppression of the leakage current. In SeTe alloys, Te incorporation increases conductivity and modifies the electronic structure of a-Se [75]. However oxide defective regions would locally disrupt the conductive network by introducing an insulating phase and defect states that hinder carrier transport. Thus despite the addition of Te, there is an increase in effective resistivity and reduces leakage current. The devices without a hole blocking layer are highly conductive as expected though less so when oxidized. Batch B was only able to maintain an applied field of 15 V/ μm before failure. This is due to increased joule heating that comes with high conductivity, which crystallizes the alloy and causes a short in the system. Batch A was able to reach fields up to 25 V/ μm , consistent with reduced conductivity and suppressed joule heating from the presence of oxides [76].

The EQE for each device under illumination from a 533 nm LED is shown in Fig. 4.10b. The difference in Te concentration percentages between Batch A (11.6 wt.%) and Batch B (16.3 wt.%) would suggest a 3-5% reduction in EQE at fields between 10-30 V/ μm . However as can be seen, Batch A had a substantial decrease in performance in this applied field range and showed a linear ECE response. Batch B exhibited behavior characteristic of a semiconductor photoconductor with the exponential rise and saturation behavior. Batch B achieves peak efficiency near 35 V/ μm , while Batch A shows no saturation even at 60 V/ μm .

These results suggest that the oxidation occurring in Batch A introduces defect states and insulating regions that have little effect on dark conduction but strongly

impair photocarrier generation and collection. The presence of Se and Te oxides and associated oxide-amorphous interfaces likely introduces trap states that capture carriers, increase recombination, and disrupt transport pathways. While dark current can still be supported through remaining conductive Se-Te pathways and is suppressed with the presence of oxides, efficient photocurrent generation requires carrier transport across the full active volume, making it far more sensitive to localized defects. As a result, oxidation produces a relatively small reduction in dark current but a pronounced suppression in EQE. Overall, these findings demonstrate that degradation of Se-Te source materials leads to defect-mediated transport limitations that significantly impact device sensitivity and the fields required to achieve high quantum efficiency.

Chapter 5

Conclusion

This work investigated key material and fabrication factors affecting the performance of direct conversion amorphous selenium and indirect conversion selenium-tellurium x-ray detectors. The primary objective was to improve detector sensitivity while maintaining low leakage current, minimal lag, and compatibility with scalable fabrication processes. The research focused on three primary areas: the effects of interrupting vacuum during a-Se deposition, optimization of hole-blocking layers (HBLs), and the impact of oxidation on Se-Te detector performance.

The investigation of breaking vacuum during a-Se deposition demonstrated that, while some variability was observed between fabricated samples, interrupting the deposition process did not produce a significant decrease in detector performance. Quantum efficiency, dark current, and recovery from irradiation remained consistent over the investigated electric field range, while variations in hole mobility and QE were found to be independent of a-Se thickness between 15 and 245 μm . These results indicate

that brief air exposure during fabrication does not adversely affect device performance. These findings demonstrate that thick photoconductive layers can be fabricated using multiple deposition steps without significantly compromising device performance, thereby overcoming the practical limitations imposed by the crucible capacity.

Among the investigated HBL materials, SiO_2 emerged as the most promising candidate, satisfying the primary performance requirements by providing high EQE, low lag and ghosting, and dark current densities below 2.9 nA/mm^2 . In addition to its electrical performance, SiO_2 offers several practical advantages, including optical transparency over the wavelengths of interest, compatibility with large area deposition techniques, low fabrication cost, and low temperature processing. While the measured leakage current remained higher than ideal, further studies into the quality of Se-Te could help. Taking another look at the remaining HBL candidates may be significantly improved through alternative deposition methods such as sputtering or ALD, which could reduce defect formation, improve thermal compatibility with Se-Te, and enhance scalability for large-area detector fabrication.

The investigation of degraded Se-Te material demonstrated that oxidation introduces defect states and insulating oxide regions that have little effect on leakage current but substantially reduce photocarrier generation. The formation of Se and Te oxides creates trap assisted recombination areas and disrupts charge transport, leading to large reductions in EQE. These findings emphasize that maintaining high purity, oxide free material is important for maximizing detector sensitivity and device lifetime while minimizing defect charge limitations.

Overall, this work demonstrates that detector performance depends not only on optimized device architecture but also on the fabrication processes and material quality. The results establish that vacuum interruptions during deposition is not a significant fabrication constraint, identifies SiO_2 as an effective and practical HBL for Se-Te detectors, and highlights the importance of preventing oxidation of Se-Te alloys. Together, these findings support the continued development of selenium based direct and indirect conversion x-ray detectors. This also identifies future directions, including the fabrication of thicker photoconductors, optimization of alternative HBL deposition techniques, and further improvements in Se-Te material purity to reduce trapping, leakage current, and performance variability.

Bibliography

- [1] L. Shi, M. Lu, N. R. Bennett, E. Shapiro, J. Zhang, R. Colbeth, J. Star-Lack, and A. S. Wang, “Characterization and potential applications of a dual-layer flat-panel detector,” *Medical physics*, vol. 47, p. 3332—3343, August 2020.
- [2] D. Walter, U. Zscherpel, and U. Ewert, “Photon counting and energy discriminating x-ray detectors - benefits and applications,” *e-Journal of Nondestructive Testing*, 2016.
- [3] S. Abbaszadeh, “Indirect conversion amorphous selenium photodetectors for medical imaging applications,” 2014.
- [4] K. Hellier, H. Mirzanezhad, M. McGrath, P. Pryor, I. Mollov, and S. Abbaszadeh, “Optimizing parylene and photoconductor thickness in indirect conversion amorphous selenium detectors,” *Proceedings of SPIE—the International Society for Optical Engineering*, vol. 13405, p. 134050E, Feb. 2025.
- [5] X. Duan, A. Howansky, H. Huang, and W. Zhao, “Evaluation of a direct-indirect dual layer detector for contrast enhanced breast imaging: Monte Carlo simulations

- and physical experiments,” in *Medical Imaging 2022: Physics of Medical Imaging*, vol. 12031, pp. 63–70, SPIE, Apr. 2022.
- [6] A. Swaby, S. Abbaszadeh, A. S. Wang, and M. J. Willemink, “Dual-layer flat panel detector with a-Se top layer for opportunistic screening of coronary artery calcium: A simulation study,” in *Medical Imaging 2022: Physics of Medical Imaging* (W. Zhao and L. Yu, eds.), (San Diego, United States), p. 143, SPIE, Apr. 2022.
- [7] W. Wang, Y. Ma, M. Tivnan, J. Li, G. J. Gang, W. Zbijewski, M. Lu, J. Zhang, J. Star-Lack, R. E. Colbeth, and J. W. Stayman, “High-resolution model-based material decomposition in dual-layer flat-panel CBCT,” *Medical Physics*, vol. 48, no. 10, pp. 6375–6387, 2021.
- [8] S. O. Kasap, M. Zahangir Kabir, and J. A. Rowlands, “Recent advances in X-ray photoconductors for direct conversion X-ray image detectors,” *Current Applied Physics*, vol. 6, pp. 288–292, June 2006.
- [9] N. Hijazi, D. Panneerselvam, and M. Z. Kabir, “Electron–hole pair creation energy in amorphous selenium for high energy photon excitation,” *Journal of Materials Science: Materials in Electronics*, vol. 29, pp. 486–490, Jan. 2018.
- [10] G. Juška and K. Arlauskas, “Impact ionization and mobilities of charge carriers at high electric fields in amorphous selenium,” *physica status solidi (a)*, vol. 59, no. 1, pp. 389–393, 1980.
- [11] S. Kasap, J. B. Frey, G. Belev, O. Tousignant, H. Mani, L. Laperriere, A. Reznik,

- and J. A. Rowlands, “Amorphous selenium and its alloys from early xeroradiography to high resolution X-ray image detectors and ultrasensitive imaging tubes,” *physica status solidi (b)*, vol. 246, no. 8, pp. 1794–1805, 2009.
- [12] L. B. Schein, “Charge generation from band-gap states in amorphous selenium films,” *Physical Review B*, vol. 10, pp. 3451–3457, Oct. 1974.
- [13] G. Pellegrini, M. Chmeissani, M. Maiorino, G. Blanchot, J. Garcia, M. Lozano, R. Martinez, C. Puigdemonges, M. Ullan, and P. Casado, “Performance limits of a 55-Mm pixel CdTe detector,” *IEEE Transactions on Nuclear Science*, vol. 53, pp. 361–366, Feb. 2006.
- [14] W. C. Barber, E. Nygard, J. C. Wessel, N. Malakhov, G. Wawrzyniak, N. E. Hartsough, T. Gandhi, and J. S. Iwanczyk, “Fast photon counting CdTe detectors for diagnostic clinical CT: Dynamic range, stability, and temporal response,” in *Medical Imaging 2010: Physics of Medical Imaging*, vol. 7622, pp. 465–472, SPIE, Mar. 2010.
- [15] S. Kasap, J. B. Frey, G. Belev, O. Tousignant, H. Mani, J. Greenspan, L. Laperriere, O. Bubon, A. Reznik, G. DeCrescenzo, K. S. Karim, and J. A. Rowlands, “Amorphous and Polycrystalline Photoconductors for Direct Conversion Flat Panel X-Ray Image Sensors,” *Sensors (Basel, Switzerland)*, vol. 11, pp. 5112–5157, May 2011.
- [16] O. Grynko, T. Thibault, E. Pineau, and A. Reznik, “Engineering of a Blocking

- Layer Structure for Low-Lag Operation of the a-PbO-Based X-Ray Detector,” *IEEE Transactions on Electron Devices*, vol. 68, pp. 2335–2341, May 2021.
- [17] Y.-H. Hu and W. Zhao, “The effect of amorphous selenium detector thickness on dual-energy digital breast imaging,” *Medical Physics*, vol. 41, no. 11, p. 111904, 2014.
- [18] D. A. Scaduto, O. Tousignant, and W. Zhao, “Experimental characterization of a direct conversion amorphous selenium detector with thicker conversion layer for dual-energy contrast-enhanced breast imaging,” *Medical Physics*, vol. 44, no. 8, pp. 3965–3977, 2017.
- [19] M. Farahmandzadeh, S. Marcinko, C. Jaramillo, M. K. Cheng, D. Curreli, and S. Abbaszadeh, “Simulation and Experimental Validation of the Uniformity of Thermally Evaporated Amorphous Selenium Films for Large-Area Imaging and Radiation Detection Applications,” *IEEE Transactions on Electron Devices*, vol. 68, pp. 626–631, Feb. 2021.
- [20] I. Hadar, X. Hu, Z.-Z. Luo, V. P. Dravid, and M. G. Kanatzidis, “Nonlinear Band Gap Tunability in Selenium–Tellurium Alloys and Its Utilization in Solar Cells,” *ACS Energy Letters*, vol. 4, pp. 2137–2143, Sept. 2019.
- [21] K. V. Reddy and A. K. Bhatnagar, “Electrical and optical studies on amorphous Se-Te alloys,” *Journal of Physics D: Applied Physics*, vol. 25, pp. 1810–1816, Dec. 1992.

- [22] H. P. D. Lanyon, “Optical and Electrical Properties of Selenium-Tellurium Alloys,” *Journal of Applied Physics*, vol. 35, pp. 1516–1523, May 1964.
- [23] H. Watanabe and K. C. Kao, “Structure and Properties of $\text{Se}_{1-x}\text{Te}_x$ Films,” *Japanese Journal of Applied Physics*, vol. 18, pp. 1849–1850, Sept. 1979.
- [24] A. L. Dawar, J. C. Joshi, and L. Narain, “Electrical and optical properties of $\text{Se}_{1-x}\text{Te}_x$ films,” *Thin Solid Films*, vol. 76, pp. 113–118, Feb. 1981.
- [25] C. Juhasz, M. Vaezi-Nejad, and S. O. Kasap, “Xerographic properties of single- and double-layer photoreceptors based on amorphous selenium-tellurium alloys,” *Journal of Materials Science*, vol. 22, pp. 2569–2582, July 1987.
- [26] G. C. Das, M. B. Bever, D. R. Uhlmann, and S. C. Moss, “Relaxation phenomena in amorphous selenium-tellurium alloys,” *Journal of Non-Crystalline Solids*, vol. 7, pp. 251–270, Apr. 1972.
- [27] S. O. Kasap and C. Juhasz, “Charge transport in chlorine doped amorphous Se:Te xerographic photoreceptor films,” *Journal of Non-Crystalline Solids*, vol. 72, pp. 23–37, June 1985.
- [28] G. Ghosh, R. C. Sharma, D. T. Li, and Y. A. Chang, “The Se-Te (Selenium-Tellurium) system,” *Journal of Phase Equilibria*, vol. 15, pp. 213–224, Apr. 1994.
- [29] T. Takahashi, “Drift mobilities in amorphous As-Se-Te ,” *Journal of Non-Crystalline Solids*, vol. 34, pp. 307–312, Dec. 1979.

- [30] W.-D. Park and K. Tanioka, “Lag and light-transfer characteristics of amorphous selenium photoconductive film with tellurium-doped layer,” *Japanese Journal of Applied Physics*, vol. 55, p. 071401, July 2016.
- [31] H. Huang and S. Abbaszadeh, “Recent developments of amorphous selenium-based x-ray detectors: A review,” *IEEE Sensors Journal*, vol. 20, pp. 1694–1704, Feb. 2013.
- [32] W. Zhao and J. A. Rowlands, “Digital radiology using active matrix readout of amorphous selenium: Theoretical analysis of detective quantum efficiency,” *Medical Physics*, vol. 24, no. 12, pp. 1819–1833, 1997.
- [33] K. Hellier, E. Benard, C. C. Scott, K. S. Karim, and S. Abbaszadeh, “Recent progress in the development of a-se/CMOS sensors for x-ray detection,” vol. 5, no. 4, p. 29. Number: 4 Publisher: Multidisciplinary Digital Publishing Institute.
- [34] D. M. Pai and R. C. Enck, “Onsager mechanism of photogeneration in amorphous selenium,” *Physical Review B*, vol. 11, pp. 5163–5174, June 1975.
- [35] B. Zhao and W. Zhao, “Temporal performance of amorphous selenium mammography detectors,” *Medical Physics*, vol. 32, no. 1, pp. 128–136, 2005.
- [36] K. Hellier, A. Swaby, J. Ott, and S. Abbaszadeh, “Performance evaluation of an amorphous selenium photodetector at high fields for application integration,” vol. 23, no. 12, pp. 12759–12766. Conference Name: IEEE Sensors Journal.
- [37] S. Abbaszadeh, C. C. Scott, O. Bubon, A. Reznik, and K. S. Karim, “Enhanced

- Detection Efficiency of Direct Conversion X-ray Detector Using Polyimide as Hole-Blocking Layer,” *Scientific Reports*, vol. 3, p. 3360, Nov. 2013.
- [38] S. Abbaszadeh, N. Allec, S. Ghanbarzadeh, U. Shafique, and K. S. Karim, “Investigation of hole-blocking contacts for high-conversion-gain amorphous selenium detectors for x-ray imaging,” *IEEE Transactions on Electron Devices*, vol. 59, no. 9, pp. 2403–2409, 2012.
- [39] S. Abbaszadeh, A. Tari, W. S. Wong, and K. S. Karim, “Enhanced dark current suppression of amorphous selenium detector with use of igzo hole blocking layer,” *IEEE Transactions on Electron Devices*, vol. 61, no. 9, pp. 3355–3357, 2014.
- [40] S. O. Kasap, “Doped and Stabilized Amorphous Selenium Single and Multilayer Photoconductive Layers for X-Ray Imaging Detector Applications,” in *Photoconductivity and Photoconductive Materials*, ch. 18, pp. 715–779, John Wiley & Sons, Ltd, 2022.
- [41] K. Hellier, A. Swaby, J. Ott, and S. Abbaszadeh, “Performance Evaluation of an Amorphous Selenium Photodetector at High Fields for Application Integration,” *IEEE Sensors Journal*, vol. 23, pp. 12759–12766, June 2023.
- [42] M. M. Wronski, W. Zhao, A. Reznik, K. Tanioka, G. DeCrescenzo, and J. A. Rowlands, “A solid-state amorphous selenium avalanche technology for low photon flux imaging applications,” *Medical Physics*, vol. 37, no. 9, pp. 4982–4985, 2010.
- [43] K. Hellier, D. A. Stewart, J. Read, R. Sfadia, and S. Abbaszadeh, “Tuning Amor-

- phous Selenium Composition with Tellurium to Improve Quantum Efficiency at Long Wavelengths and High Applied Fields,” *ACS Applied Electronic Materials*, vol. 5, pp. 2678–2685, May 2023.
- [44] W. E. Spear, “Transit Time Measurements of Charge Carriers in Amorphous Selenium Films,” *Proceedings of the Physical Society. Section B*, vol. 70, pp. 669–675, July 1957.
- [45] J. L. Hartke, “Drift Mobilities of Electrons and Holes and Space-Charge-Limited Currents in Amorphous Selenium Films,” *Physical Review*, vol. 125, pp. 1177–1192, Feb. 1962.
- [46] H. P. Grunwald and R. M. Blakney, “Electron and Hole Drift Mobilities in Vitreous Selenium,” *Physical Review*, vol. 165, pp. 1006–1010, Jan. 1968.
- [47] M. D. Tabak, “Temperature Dependence of Hole Transport in Vitreous Selenium,” *Physical Review B*, vol. 2, pp. 2104–2108, Sept. 1970.
- [48] S. O. Kasap, K. V. Koughia, B. Fogal, G. Belev, and R. E. Johanson, “The influence of deposition conditions and alloying on the electronic properties of amorphous selenium,” *Semiconductors*, vol. 37, pp. 789–794, July 2003.
- [49] S. Kasap, C. Koughia, J. Berashevich, R. Johanson, and A. Reznik, “Charge transport in pure and stabilized amorphous selenium: Re-examination of the density of states distribution in the mobility gap and the role of defects,” *Journal of Materials Science: Materials in Electronics*, vol. 26, pp. 4644–4658, July 2015.

- [50] J. L. Hartke and P. J. Regensburger, “Electronic States in Vitreous Selenium,” *Physical Review*, vol. 139, pp. A970–A980, Aug. 1965.
- [51] M. D. Tabak, “Photogeneration Effects in Amorphous Selenium,” *Applied Optics*, vol. 8, pp. 4–7, Jan. 1969.
- [52] J. Jakabovič, J. Kováč, M. Weis, D. Haško, R. Srnánek, P. Valent, and R. Resel, “Preparation and properties of thin parylene layers as the gate dielectrics for organic field effect transistors,” *Microelectronics Journal*, vol. 40, no. 3, pp. 595–597, 2009.
- [53] T. Wei, K. Fujiwara, T. Kanki, and H. Tanaka, “Impact of parylene-C thickness on performance of KTaO₃ field-effect transistors with high-k oxide/parylene-C hybrid gate dielectric,” *Journal of Applied Physics*, vol. 119, p. 034502, Jan. 2016.
- [54] I. Tszydel, T. Marszalek, J. Ulanski, A. Nosal, and M. Gazicki-Lipman, “Applications of parylene films in the manufacture of organic field-effect transistors,” *Surface and Coatings Technology*, vol. 290, pp. 21–27, Mar. 2016.
- [55] F.-C. Chiu, “Thickness and Temperature Dependence of Dielectric Reliability Characteristics in Cerium Dioxide Thin Film,” *IEEE Transactions on Electron Devices*, vol. 57, pp. 2719–2725, Oct. 2010. Conference Name: IEEE Transactions on Electron Devices.
- [56] C.-H. Chen, W.-c. Shih, H.-L. Huang, and F.-C. Chiu, “Temperature and Thickness

- Dependence of Cerium Oxide Dielectric Breakdown,” *ECS Transactions*, vol. 28, p. 421, Apr. 2010.
- [57] S. Abbaszadeh, N. Allec, S. Ghanbarzadeh, U. Shafique, and K. S. Karim, “Investigation of Hole-Blocking Contacts for High-Conversion-Gain Amorphous Selenium Detectors for X-Ray Imaging,” *IEEE Transactions on Electron Devices*, vol. 59, pp. 2403–2409, Sept. 2012. Conference Name: IEEE Transactions on Electron Devices.
- [58] H. Kannan, J. Stavro, A. Mukherjee, S. Léveillé, K. Kisslinger, L. Guan, W. Zhao, A. Sahu, and A. H. Goldan, “Ultralow Dark Currents in Avalanche Amorphous Selenium Photodetectors Using Solution-Processed Quantum Dot Blocking Layer,” *ACS Photonics*, vol. 7, pp. 1367–1374, June 2020.
- [59] J. M. Aitken, “Radiation-induced trapping centers in thin silicon dioxide films,” *Journal of Non-Crystalline Solids*, vol. 40, pp. 31–47, July 1980.
- [60] A. G. Revesz and H. L. Hughes, “The structural aspects of non-crystalline SiO₂ films on silicon: a review,” *Journal of Non-Crystalline Solids*, vol. 328, pp. 48–63, Oct. 2003.
- [61] A. Pil-Ali, G. Kaur, S. Adnani, M. A. Mohajerzadeh, and K. S. Karim, “Enhanced temporal and electrical stability of amorphous selenium photoconductor by an aluminum oxide thin-film hole-blocking layer,” in *Medical Imaging 2026: Physics of Medical Imaging*, vol. 13924, p. 20, SPIE, Apr. 2026.

- [62] C.-Y. Chang, F.-M. Pan, J.-S. Lin, T.-Y. Yu, Y.-M. Li, and C.-Y. Chen, “Lateral amorphous selenium metal-insulator-semiconductor-insulator-metal photodetectors using ultrathin dielectric blocking layers for dark current suppression,” *Journal of Applied Physics*, vol. 120, p. 234501, Dec. 2016.
- [63] R. Li, S. Bai, Z. Jia, X. Chen, Y. Liu, and Q. Lin, “Flexible Broadband Photodiodes Based on Amorphous $\text{Te}_{0.4}\text{Se}_{0.6}$ Thin Films,” *ACS Photonics*, vol. 11, pp. 2521–2527, June 2024.
- [64] H. Mirzanezhad, K. Hellier, M. Teicheira, and S. Abbaszadeh, “Spectral Performance of Multilayer Amorphous Selenium and Selenium–Tellurium Photodetectors,” *ACS Applied Optical Materials*, vol. 3, pp. 646–655, Mar. 2025.
- [65] P. I. L. d. B. E. L. J. A. J. C. D. S. J. C. R. S. K. O. R. D. Deslattes, E. G. Kessler Jr. and R. Dragoset, “X-Ray Transition Energies Database Version 1.2,” Aug. 2005.
- [66] A. Nicolellis and M. Newville, “X-ray Data for the Elements.”
- [67] I. V. Baimler, A. V. Simakin, A. O. Dikovskaya, V. V. Voronov, O. V. Uvarov, A. A. Smirnov, A. V. Sibirev, A. S. Dorokhov, and S. V. Gudkov, “Fabrication and growth mechanism of t-selenium nanorods during laser ablation and fragmentation in organic liquids,” *Frontiers in Chemistry*, vol. 12, p. 1449570, 2024.
- [68] K. Ikeda and T. Ohtani, “Peculiar crystallization of amorphous selenium in the presence of organic liquids: Effects of addition of titanium oxides,” *Journal of Alloys and Compounds*, vol. 434-435, pp. 275–278, May 2007.

- [69] D. Zagorac, H. Müller, S. Ruehl, J. Zagorac, and S. Rehme, “Recent developments in the Inorganic Crystal Structure Database: theoretical crystal structure data and related features,” *Journal of Applied Crystallography*, vol. 52, pp. 918–925, Oct. 2019.
- [70] R. Keller, W. B. Holzapfel, and H. Schulz, “Effect of pressure on the atom positions in Se and Te,” *Physical Review B*, vol. 16, pp. 4404–4412, Nov. 1977.
- [71] C. E. M. Campos, K. Ersching, J. C. de Lima, T. A. Grandi, H. Höhn, and P. S. Pizani, “Influence of minor oxidation of the precursor powders to form nanocrystalline CdTe by mechanical alloying,” *Journal of Alloys and Compounds*, vol. 466, pp. 80–86, Oct. 2008.
- [72] “The crystal structure of SeO₂ at 139 and 286 K,” *Zeitschrift für Kristallographie*, vol. 202, pp. 99–107, Jan. 1992.
- [73] J. E. Hoffmann and M. G. King, “Selenium and Selenium Compounds,” in *Kirk-Othmer Encyclopedia of Chemical Technology*, pp. 1–36, John Wiley & Sons, Ltd, 2010. _eprint: <https://onlinelibrary.wiley.com/doi/pdf/10.1002/0471238961.1905120508150606.a01.pub3>.
- [74] J. E. Hoffmann, M. G. King, S. C. Carapella, J. E. Oldfield, and R. D. Putnam, “Tellurium and Tellurium Compounds,” in *Kirk-Othmer Encyclopedia of Chemical Technology*, pp. 1–30, John Wiley & Sons, Ltd, 2011. _eprint: <https://onlinelibrary.wiley.com/doi/pdf/10.1002/0471238961.2005121208150606.a01.pub3>.

- [75] R. M. Mehra, P. C. Mathur, A. K. Kathuria, and R. Shyam, “Electrical transport in bulk amorphous selenium,” *Journal of Physics and Chemistry of Solids*, vol. 39, pp. 295–298, Jan. 1978.
- [76] D. Armitage and C. H. Champness, “Switching in amorphous selenium,” *Journal of Non-Crystalline Solids*, vol. 7, pp. 410–416, May 1972.