Comparison of Thin Epitaxial Film Silicon Photovoltaics Fabricated on Monocrystalline and Polycrystalline Seed Layers on Glass

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We fabricate thin epitaxial crystal silicon solar cells on display glass and fused silica substrates overcoated with a silicon seed layer. To confirm the quality of hot-wire chemical vapor deposition epitaxy, we grow a 2- μ m-thick absorber on a (100) monocrystalline Si layer transfer seed on display glass and achieve 6.5% efficiency with an open circuit voltage (V_{OC}) of 586 mV without light trapping features. This device enables the evaluation of seed layers on display glass. Using polycrystalline seeds formed from amorphous silicon by laser-induced mixed-phase solidification (MPS) and electron beam crystallization (EBC), we demonstrate 2.9 %, 476 mV (MPS) and 4.1 %, 551 mV (EBC) solar cells. Grain boundaries likely limit the solar cell grown on the MPS seed layer and we establish an upper bound for the grain boundary recombination velocity (S_{GB}) of 1.6x10⁴ cm/s.

1. Introduction

In efforts to reduce the photovoltaic (PV) costs, research groups worldwide are fabricating solar cells using thin (< 20 µm thick) crystalline silicon (c-Si) layers. Thin crystal silicon solar cells are more material- and energy-efficient route to c-Si than wafer fabrication and could, therefore, reduce PV module costs if comparable efficiencies could be achieved. There may also be markets that benefit from multi-crystal silicon PV deposited on transparent display glass with a thermal expansion matched to Si. To date, however, thin polycrystalline photovoltaics have significantly lower efficiencies than wafer Si. While somewhat reduced short circuit currents (J_{SC}) are expected in thin c-Si PV [1], [2], the primary problem has been low open circuit voltage (V_{OC}) compared with \approx 630 mV commercial multicrystalline wafer PV. The company CSG Solar achieved 10.5 % efficiency with $\approx 1 \, \mu m$ to 3 μm grained thin films formed by annealing amorphous Si films [3]. Higher efficiencies were limited by the low V_{OC} of about 500 mV. Using Al-induced crystallization of amorphous silicon and subsequent >1000 °C epitaxy, IMEC fabricated $V_{OC} = 520 \text{ mV}$, 8.5% efficient solar cells on ceramic substrates.[4]-[6] Van Gestel et al. reported a 5.4 %, 494 mV Voc device on MPS seeds, also using high temperature epitaxy and ceramic substrates [7]. Recently, Dore et al. and Haschke et al. [8] have made significant V_{OC} improvements using laser or e-beam crystallized silicon featuring ≈100 µm wide, elongated grains about 1 cm long. Dore et al. report 11.7% efficient, 585 mV solar cells [9]. Using ≈ 100 nm grained CaF₂ seed layers on glass formed by ion beam assisted deposition (IBAD), we previously fabricated a 0.8 %, 375 mV device [10]. In all of these reports, the authors attribute low V_{OC} primarily to insufficient c-Si quality in the absorber and junction regions.

Here, we report new insights into thin c-Si PV obtained from analysis of three solar cells on glass grown by seed and epitaxy[11], [12] In the seed and epitaxy process, we initially fabricate a thin c-Si seed on display glass and then epitaxially thicken the layer to form the PV absorber [6], [11]-[17]. Epitaxial layers are grown by hot-wire chemical vapor deposition (HWCVD) [18], [19]. Compared to other thin, c-Si PV approaches, seed and epitaxy has two advantages. First, there is more flexibility in the seed fabrication process because the seed can be made thin (<1 μ m). Second, epitaxy allows more flexible dopant profiles. Solar cells without light trapping fabricated on layer-transfer seeds on display glass (6.5 % efficient), MPS seeds on silica (2.9 % efficient) and electron beam crystallized seeds on display glass (4.1 % efficient). The layer transfer process used here is too expensive for commercial PV, but since the seed is monocrystalline and (100)-oriented, it likely provides the best seed that could be achieved. The polycrystalline seeds are fabricated with amorphous-silicon-based processes that could be manufactured at the extremely low costs that are needed for economically-competitive PV. We correlate solar cell quality with the seed grain size, structure and orientation and compare the devices to other reported thin c-Si PV devices in order to better understand the limits to efficiency. The analysis reveals two main insights: 1) the potential of HWCVD silicon epitaxy on display glass and 2) new insights into how seed grain size and orientation impact the performance of thin c-Si solar cells.

2. Experiment

We fabricate solar cells on three seed layers, as shown schematically in Fig. 1. The first seed layer is a 500 nm thick (100) n^+ -doped single crystal that was cleaved from a wafer and bonded to Corning Eagle glass with a layer transfer process ('LT seed') described elsewhere [20]. The second seed is created by laser-induced mixed-phase solidification of amorphous silicon grown on silica substrates ('MPS seed'). The third seed is electron-beam crystallized amorphous silicon grown on glass substrates ('EBC seed').

MPS seeds were fabricated by 1) depositing \approx 130 nm thick heavily phosphorus doped amorphous Si (a-Si:H) onto fused silica substrate by plasma enhanced chemical vapor deposition, 2) annealing the deposited a-Si:H films at 600 °C to remove the excess hydrogen and 3) crystallizing the thin silicon layers via mixed phase solidification (MPS), as described in detail elsewhere [21]. Briefly, we scanned the silicon film with a 750 µm by 30 µm beam from a continuous-wave 532 nm solid-state laser at room temperature. The laser power at the sample was 5.2 W and the sample was scanned through the beam at 15 mm/s. By overlapping subsequent scans (5 to 7 times), we obtain c-Si grains a few microns wide and with {100} surface texture (see below). Hall measurement of the MPS seed indicates a carrier concentration of $4x10^{20}$ cm⁻³ and a mobility of 23 cm²/V-s (corresponding to 41 Ω/\Box).

EBC seeds are prepared as follows: 1) we sputter an intermediate layer stack (200 nm SiOX and 10 nm SiCX) onto a 1.1 mm thick glass substrate. 2) We deposit a 10 μ m thick Si layer with a B concentration of $4x10^{16}$ cm⁻³ by electron-beam co-evaporation of silicon and boron. 3) We crystallize the Si layer with a line-shaped electron-beam that is scanned once

across the sample surface. The resulting $\approx 10 \ \mu m$ thick poly-Si layer has light p-type doping and was intended for use as an absorber, as described elsewhere [8]. Here, we use the entire layer as a 'seed' for our seed and epitaxy. In an optimized geometry, the crystallized seed layer would be thinner and far more heavily doped so it could act as a back contact.

Solar cells (see Fig. 1) are fabricated by 1) epitaxially growing silicon layers at ≈ 200 nm/min by hot-wire chemical vapor deposition (HWCVD) at ≈750 °C from SiH₄ and PH₃ precursors, 2) hydrogenating the silicon with a remote plasma [22], 3) growing an i/p a-Si:H heterojunction by HWCVD, 4) evaporating a 70 nm thick ITO anti-reflection layer, 5) lithographically isolating 2.5 mm diameter mesas and 6) evaporating a top metal grid and bottom contact to the (underlying n^+ Si). Before the epitaxial growth, each seed was solvent-cleaned to remove organics and then etched with 4 % HF to remove the native oxide. Under these conditions, we do not observe any deformation of the display glass substrates. We described this process in more detail previously [22]. We intentionally grow a $\approx 2 \mu m$ thick epitaxial HWCVD n^+ layer on the EBC seed to make the back contact. For the MPS and LT seeds, the n^+ seed itself acts as the back contact and no epitaxial n^+ layer is needed. We grew 1 µm thick epitaxial absorbers for both the MPS and EBC seed devices and a 2 µm epitaxial absorber on the LT seed. After device fabrication, we measure the solar cell dark and AM1.5 current-voltage (J-V) response, total reflection and quantum efficiency (QE). We characterize the crystallographic orientation with x-ray diffraction (XRD), using a Bruker D2000 system with a two-dimensional (2D) detector that measures a stereographic projection of the x-rays diffracting from the sample (i.e., a range of 2θ and χ angles). We examine the local EBC solar cell collection efficiency using electron beam induced current (EBIC). Cross-sectional transmission electron microscopy (TEM) reveals the silicon seed and epitaxy microstructure.

3. Results

Fig. 2 shows optical microscope images of the completed solar cells on MPS and EBC seeds, revealing the grain size and shapes. Fig. 2a and 2c are low-magnification images showing nearly the entirety of one circular solar cell mesa. At low magnification, the MPS grains are too small to distinguish (Fig. 2a); however, at higher magnification (Fig. 2b), the 1 μ m to 3 μ m-wide grains are apparent. The MPS grains are very uniform in size, with essentially no grains with dimension below 1 μ m. The much larger grains in the EBC solar cell are hundreds of microns

wide and easily visible even at low magnification (Fig. 2c). Optical images of the LT seed solar cell (not shown) are largely featureless because the surface is smooth and there are no grains.

Fig. 3 shows XRD measurements that reveal the crystallographic orientation of the grains in the two devices. In Fig. 3a-d, we show images of the XRD detector at different 2 θ ranges for the MPS and EBC solar cells. The χ range (vertical axis) of the projection is approximately -20 to +20°. Peaks appearing at $\chi \neq 0$ correspond to grains oriented away from the surface normal. Randomly oriented polycrystalline silicon samples with small grains (so many grains are sampled by the $\approx 1 \text{ mm}^2 \text{ x-ray beam}$) would produce rings on the detector image corresponding to the constant 2 θ values for each Si plane spacing. Polycrystalline Si samples with uniaxial (100) texture produce a (400) diffraction streak that is most intense at $\chi=0$ and become less intense at higher χ angles. To obtain the customary 2 θ dependence, shown in e), we integrate the x-ray intensity over approximately $\chi = -10$ to $+10^\circ$. Both samples show diffraction from the polycrystalline ITO coating (vertical streaks with weak χ dependence in the two-dimensional detector images. We denote the ITO peaks from the EB solar cell with asterisks (panel e); in the other solar cells, the ITO peaks appear at the same 2 θ positions.

The detector frames from the EBC seed device have bright spots along the arcs corresponding to the 2 θ values for Si crystallographic plane spacings (panel f). In the EBC device, the diffraction creates spots, instead of streaks, because a total of only about 10 grains are sampled in the x-ray beam area. Thus, each bright spot at Si 2 θ positions in the frame likely arises from a single grain. Because the Si(400) and Si(331) Si grains in this particular sample location happen are oriented with $|\chi|>10^{\circ}$ (dotted arrows in Fig 3b), these spots are not observed in the 2 θ integration and no peak corresponding to these spots appears in Figure 3e. When we repeat the XRD measurement on different physical locations on the EBC sample, we observe similar spotty images, but with the spots in different positions along the arcs, corresponding to constant 2 θ .

The primary Si diffraction observed from the MPS sample is the Si(400) signal indicated in by the arrow in panel d. The Si (400) is centered about $\chi=0^{\circ}$, with a full-width half maximum of 12°, indicating preferred (100) texture normal to the surface. This XRD measurement is consistent with previously-reported electron beam scattered diffraction measurements that found the MPS grains to be uniaxial textured (no preferential in-plane orientation) [21]. For the LT solar cell, only Si (400) diffraction is observed under the measurement conditions and the diffraction appears as a sharp spot on the 2D detector (not shown). We plot the integrated θ -2 θ data for the LT seed in Fig. 3e. The XRD confirms that the HWCVD epitaxy was successful in replicating the original seed crystal structure on all three devices. In contrast, if epitaxy fails at a 750 °C growth temperature, HWCVD produces a polycrystalline film with < 100 nm grains [18]. The resulting XRD would produce broad streaks at all Si 2 θ positions.

Fig. 4 shows the current-voltage (J-V) curves for each of the three solar cells, measured with a Ag metal back reflector (BR) behind the substrate (providing ≈ 95 % reflectivity). The EBC seed device is better in all respects than the MPS seed device, but neither is as good as the device on the LT seed. The LT seed J-V has an 'S-shape' near V_{OC}. In our experience, an S-shaped J-V curve typically indicates that the intrinsic amorphous silicon in the heterojunction layer was unintentionally grown too thick.

In Fig. 5, we plot the external quantum efficiency (QE) of the three devices and two QE calculations with the aim of comparing the absorber quality. The QE calculations, described previously [22], assume perfect collection in the absorber, but assume that no light absorbed by the a-Si:H, back contact or seed layers is collected. The model includes the optical effects of the ITO layer. Fig. 5a compares the LT seed QE, measured with a metal back reflector (BR) placed behind the display glass substrate, and a calculation for a 2 μ m absorber with an 0.5 μ m thick c-Si back contact (the seed) and a 95%-reflective BR. At wavelengths above 700 nm, the measured QE and calculated QE are similar. At wavelengths below 700 nm, the measured QE is lower than the calculation. The good long-wavelength QE suggests that the absorber grown on the LT seed has sufficient quality for full absorption. It is likely that the intrinsic amorphous silicon heterojunction layer was unintentionally grown unnecessarily thick, reducing the low-wavelength QE and also explaining the S-shaped J-V seen in Fig. 4.

Fig. 5b compares the QE of the EBC and MPS solar cells with a calculation for a 1 μ m absorber with no back reflection. In the EBC device, back reflection is ineffective because the thick seed and n⁺ layers absorb most of the light before it can re-enter the absorber. Therefore, we measure and plot the QE of both devices without a BR so that they both act approximately as comparable, 1- μ m-thick absorbers. The MPS solar cell does have higher QE at longer wavelengths when measured with BR, whereas the EBC solar cell QE is unaffected by the presence of a BR. The MPS solar cell QE is poorer than the calculation at most wavelengths,

suggesting that the c-Si quality is insufficient for adequate collection. At all wavelengths, the EBC solar cell QE is higher than both the MPS device and the calculated one pass QE. This is for two reasons. First, the EBC n- c-Si absorber quality is superior to the MPS absorber, and adequate for full collection from the 1 μ m thick absorber. Second, the epitaxial n⁺ layer [1] and possibly the seed itself, also contribute a small amount of current.

To estimate the diffusion length (L_D) of the absorber of the three devices, we replot the one-pass internal QE data versus absorption depth (data not shown) and calculate the inverse slope according to the method of Hirsch et al. [23]. The L_D calculation results are included in the table inset to Fig. 4; higher L_D values are consistent with the trends in V_{OC} and the approximation suggested by Alberi et al. [1] that complete collection demands that L_D be 3X the quasi-neutral layer thickness.

Fig. 6a shows a low-resolution EBIC image of the EBC solar cell mesa. Qualitatively, bright and dark grains indicate higher and lower carrier extraction efficiencies, respectively. Figs. 6b-e show higher resolution SEM and EBIC images acquired from dark (frames b and c) and bright grains (frames d and e) using identical electron currents. In f, we show the actual EBIC current for the line scans indicated in frames c and e. Note that the EBIC collection current is not expected to be linearly proportional to the 1 sun QE. Within the bright grain, diagonal borders separate slightly brighter and darker regions. These borders may indicate stacking faults in the absorber. In both the dark and bright grains, there are a number of circular defects (black spots in frames c and e) with reduced EBIC collection. These defects reduce collection within a diameter of about $\approx 2 \,\mu$ m to 5 μ m, corresponding roughly to L_D. The line scans in frame f show that EBIC collection is ≈ 1.6 X higher in the brighter grain and that the local defects reduce collection dramatically in both the bright and dark grains.

The large difference in EBIC collection efficiency between adjacent grains is striking. In order to examine the crystallographic quality, we examined multiple grains of the EBC solar cell with cross-sectional TEM (not shown). TEM confirms epitaxial growth and reveals a variety of defects in the epitaxial layers of the EBC solar cell, including dislocations and voids. However, the defects are present in all of the cross sections and we could not correlate differences in EBIC collection efficiency with structural differences.

In Fig. 7, we show a TEM image of the MPS solar cell. The a-Si:H and ITO layers of the sample were removed during the ion milling preparation. In the imaged grain, there are defects

present in both the seed and epitaxial layer. In other images (not shown), we often observe grains without defects in either the seed or epitaxy layers. In the Fig. 7 grain, some dislocations appear to nucleate at the seed/epitaxy interface.

4. Discussion

Unsurprisingly, the best solar cell was fabricated on the LT seed on display glass. This device establishes a baseline quality for the glass-compatible HWCVD epitaxy, hydrogenation, heterojunction and contacting processes. Analysis of the internal QE shows $L_D \approx 7 \mu m$ in the LT seed solar cell, sufficient for excellent collection from the 2 μm absorber (Fig. 5). The 586 mV V_{OC} is excellent for a c-Si solar cell on glass, but still lower than 620 mV epitaxial solar cells we fabricated on (inactive) Si wafer substrates by nearly identical methods [24]. It is likely that poor passivation of seed/glass interface accounts for some reduction in V_{OC} . We cannot rule out impurity diffusion from the glass into the c-Si during the seed formation process or the 750 °C epitaxy step.

The solar cells grown on both polycrystalline seeds have lower V_{OC} and shorter diffusion lengths ($\approx 3 \mu m$) than the LT seed device due to the lower quality seed layers. On the MPS seed, we expect high quality epitaxy because the grains are nearly (100)-oriented, the same face on which epitaxy has been well-optimized [19]. L_D is similar to the grain size and it is likely that GB recombination plays a dominant role. TEM in Fig. 7 reveals dislocation densities smaller than those in heteroepitaxial absorbers grown on sapphire which had L_D $\approx 4.5 \mu m$ and produced 6.8%, 574 mV solar cells [22]. It is likely that the intragrain material quality is good and that recombination at the GBs between the small grains dominates losses in the MPS solar cell.

Several authors have calculated the effects of GB recombination and reported ranges of bulk and GB recombination where performance is limited by the GBs [20], [25], [26]. Taretto et al. relate the effective L_D in a polycrystalline material to the surface recombination velocity at grain boundaries, S_{GB} , the grain size, g, and the effective 'intra-grain' diffusion length ' L_{mono} ' [26]. Note that L_{mono} accounts for recombination within the grain and at the sample surface (but not at GBs) and depends on the absorber thickness. Notes that the approximate transition from bulk to GB limited performance occurs when:

$$L_{mono} > \sqrt{\frac{gD}{2S_{GB}}} = L_{GB} \quad , [Eq. 1]$$

where D is the minority carrier diffusion constant. With this definition of L_{GB} , the Taretto et al. expression for the effective polycrystalline L_D can be written [26]:

$$\frac{1}{L_D^2} = \frac{1}{L_{mono}^2} + \frac{1}{L_{GB}^2} \qquad [Eq. 2]$$

To calculate an upper bound on S_{GB} in the MPS device, we presume excellent intragrain quality, $L_{mono} \gg L_D$ and insert our measured value of $L_D = 2.7 \ \mu m$ calculated from the inverse internal QE into Eq. 2. Using g = 3 μm and D = 8 cm²/s in Eq. (1), we obtain $S_{GB} \approx 1.6 \times 10^4 \text{ cm/s}$. The 2.9%, 476 mV MPS solar cell reported here is inferior to the 5.4%, 494 mV MPS solar cell grown on a glass ceramic reported by van Gestel et al., who incorporated surface texturing and achieved >16 mA/cm² [7]. It is possible that their high (> 1000 °C) epitaxy temperature promotes better material quality, although TEM images presented by van Gestel are comparable to what we observe in Fig. 7. The <500 mV V_{OC} but low-intragrain-defect TEM of both devices suggests that GB recombination dominates. In contrast, electronic measurements of solar cells fabricated from 1 μm to 3 μm -wide-grain poly-Si formed by solid phase crystallization (SPC) suggest that recombination occurs primarily at dislocations. The dislocation density in SPC Si can be as high as 10¹⁰ cm⁻², orders of magnitude higher than observed in Fig. 7 [27]-[29]. It may be that while SPC silicon has poorer intragrain quality (more dislocations) than MPS Si, SPC silicon has better GB passivation than MPS silicon.

It is not clear what limits the V_{OC} of the EBC solar cell. Because the EBC solar cell has much larger grains (g > 200 µm), we expect GBs to play a smaller role. As discussed above, the QE of the EBC solar cell (Fig. 5) indicates that the device collects nearly all electron-hole pairs generated by absorbed light. Obviously, the 551 mV V_{OC} is much higher than the 476 mV achieved on the MPS seed and closer to the 586 mV achieved on the LT seed. It is likely that a thicker absorber would result in even higher EBC cell efficiency. On this sample, EBIC (Fig. 6) suggests that carrier collection efficiency varies substantially across different grains. It is likely that epitaxy quality would vary with the orientation of the seed grain. For example, we previously found that HWCVD epitaxial quality was better on Si (100) than Si (111) surfaces at temperatures <650 °C [18]. However, TEM cross sections reveal similar defect densities in different grains, suggesting that crystallographic defects are not responsible for reduced collection. Structural defects should reduce collection locally (within a few L_D of a dislocation); such defects could not explain the background of poor EBIC collection in the dark grains (red curves, Fig. 6f), seen far from local EBIC-active defects. One possible explanation is that some grains have higher impurity concentrations, but we have no direct evidence of grain-to-grain impurity variation. The present data does not clearly identify the strongest recombination source in the EBC seed solar cell or the reason for the difference in EBIC collection between adjacent grains.

It is useful to compare the 4.1 %, 551 mV EBC solar cell to previous polycrystalline solar cells that also had grains significantly larger than the absorber thickness. Dore et al. prepared a 11.7 %, 585 mV solar cell with laser crystallization [9] and Qiu et al. prepared a 8.5 %, 523 mV V_{OC} solar cell by 1100 °C epitaxy on an aluminum-induced-crystallization (AIC) seed on alumina substrates [30]. The laser crystallization process described by Dore et al. produces grains very similar to our EBC seeds. The 11.7% laser crystallized cell had a SiOX/SiNX/SiOX buffer between the glass and silicon and used the $\approx 10 \,\mu m$ crystallized layer as the absorber (so epitaxy was not required). Using a SiOX/SiCX buffer (similar to the one used on our EBC cell), solar cells were only 3.3 %, 436 mV.[31] The mechanism responsible for the differences in these laser crystallized solar cells with different buffer layers is still under investigation. The biggest difference between the optimized laser crystallized devices (using a SiOX / SiNX / SiOX buffer) is in J_{SC}. The J_{SC} of our EBC solar cell is limited by the thin absorber, extremely thick seed layer and lack of light trapping texture. Unfortunately, the present results are insufficient to reveal whether epitaxy was better or worse than the EBC seed. In contrast to EBC or laser crystallization, AIC produces a significant distribution of grain sizes, with most grains >10 µm in width. Qiu et al. were able to optimize hydrogenation and surface texturing to achieve $J_{SC} > 21$ mA/cm², so a thicker HWCVD EBC seed solar cell might be expected to reach the same efficiency.

Finally, we note that the S_{GB} that we measure for the MPS cell could explain the poor performance of solar cells that we previously fabricated from 0.5-µm-wide, (111)-oriented c-Si grains grown epitaxially on CaF₂ seeds.[10] With $S_{GB} = 1.6 \times 10^4$ cm/s and g = 0.5 µm, Eq. 1 indicates $L_{GB} \approx 1.1$ µm, too small for the 1.6 µm absorber used in those solar cells.

5. Summary

In summary, we report 3 crystal silicon solar cells fabricated on seeds on display glass or silica. Using a 2 μ m thick absorber on a monocrystalline seed, we achieve L_D \approx 7 μ m, V_{OC} = 586 mV and 6.5% efficiency, demonstrating the quality of display-glass-compatible HWCVD epitaxy. Using polycrystalline silicon seeds, lower V_{OC}'s and efficiencies indicate that higher quality seed layers are needed for seed and epitaxy to achieve commercially relevant efficiencies. With MPS seeds, grain boundaries reduce solar cell performance and we estimate the surface recombination velocity to be S_{GB} \approx 1.6x10⁴ cm/s. With EBC seeds, we achieve 4.1% efficiency and V_{OC} of 551 mV. However, in the EBC solar cell, current collection is limited by the thin absorber and the lack of light trapping.

6. Acknowledgements

We thank Ta-Ko Chang of Corning Inc. for supplying the layer transfer seeds on display glass (LT seed). Work at NREL was funded by the U.S. Department of Energy Office of Energy Efficiency and Renewable Energy under Contract DE-AC36-08GO28308. We gratefully acknowledge Bill Nemeth for the amorphous silicon growth used to create the MPS seed layers. H. P. Yoon acknowledges support under the Cooperative Research Agreement between the University of Maryland and the National Institute of Standards and Technology Center for Nanoscale Science and Technology, Award 70NANB10H193, through the University of Maryland.

Figures



Fig. 1. Schematics of the three solar cells

Fig. 2. Optical microscope images of finished solar cells on MPS (a and b) and EBC (c and d) seeds. In the low magnification images, most of the solar cell mesa (see Fig. 1) is visible. The white partial circle is the front metal grid. The long grains of the solar cell on the EBC seed are visible in the low-mag image c. In the higher magnification image in b), the \approx 3 µm grains of the MPS sample are visible. In d), intermediate magnification reveals some \approx 100 µm grains in the EBC sample.



Fig. 3. Crystallographic analysis by X-ray diffraction. a) and b) show x-ray detector frames for the EBC solar cell. c) and d) show x-ray detector frames for the MPS solar cell. In e), we show the result of integrating the detector frames in a-d and of similar measurements on the LT seed over χ . In f), we indicate the 2 θ positions and peak intensities for a Si powder. The arrows in b) indicate diffraction peaks outside the integration used to generate the EBC curve in e).



Figure 4. J-V measurement of the 3 solar cells with a back reflector placed below the substrate. The table indicates the solar cell parameters. L_D was measured from plots of inverse QE vs. absorption depth (see text).



Figure 5. External QE measurements and calculations. The QE for the EBC and LT seed devices were measured without a back reflector (BR). In panel a, the calculation assumes a 2 μ m absorber, 0.5 μ m n⁺ contact layer, and 95 % back reflection. In panel b, the calculation assumes a 1 μ m absorber and no BR.



Figure 6. EBIC characterization of the 4.1% EBC solar cell. In a), low magnification EBIC reveals the contrast from grain to grain. The image has been rotated so that the solar cell is in the same orientation as in Fig. 2c. In b) and c) we show higher magnification SEM and EBIC images of the indicated regions of a 'darker' grain with relatively poor collection efficiency. In d) and e), we show SEM and EBIC images of a 'brighter grain.' In f), we plot the collected EBIC current for line scans along the four arrows indicated in frames c) and e).



20 keV, 1.3 nA

Fig. 7. Cross sectional TEM of the MPS seed solar cell



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