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Impact of Black Silicon on Light- and Elevated Temperature-Induced Degradation in Industrial Passivated Emitter and Rear Cells

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ABSTRACT

Light and elevated-temperature induced degradation (LeTID) is currently a severe issue in passivated emitter and rear cells (PERC). In this work, we study the impact of surface texture, especially a black silicon (b-Si) nanostructure, on LeTID in industrial p-type mc-Si PERC. Our results show that during standard LeTID conditions the b-Si cells with atomic-layer-deposited aluminum oxide (AlOₓ) front surface passivation show no degradation despite the presence of a hydrogen-rich AlOₓ/SiNx passivation stack on the rear. Furthermore, b-Si solar cells passivated with silicon nitride (SiNx) on the front lose only 1.5 %rel of their initial power conversion efficiency, while the acidic-textured equivalents degrade by nearly 4 %rel under the same conditions. Correspondingly, clear degradation is visible in the IQE of the acidic-textured cells, especially in the ~850–1100 nm wavelength range confirming that the degradation occurs in the bulk, while the IQE remains nearly unaffected in the b-Si cells. The observations are supported by spatially-resolved photoluminescence (PL) maps, which show a clear contrast in the degradation behavior of b-Si and acidic-textured cells, especially in the case of SiNx front surface passivation. The PL maps also suggest that the magnitude of LeTID scales with surface area of the texture, rather than wafer thickness that was recently reported, although the b-Si cells are slightly thinner (140 vs. 165 µm). The results indicate that b-Si has a positive impact on LeTID, and hence, benefits provided by b-Si are not limited only to the excellent optical properties, as commonly understood.

1. INTRODUCTION

The capability to eliminate reflective losses of diamond wire-sawn multicrystalline silicon (mc-Si) wafers have made black silicon (nanostructured silicon, b-Si) of particular interest within the photovoltaics (PV) community [1, 2]. Furthermore, the large surface area of b-Si is no longer a hindrance for high-efficiency b-Si solar cells due to the advancements in novel surface passivation methods [1, 3-5]. In fact, it has been
recently shown that the large surface area can provide also an advantage, since it significantly enhances gettering of deleterious metal impurities during phosphorus diffusion [6].

The evidence for enhanced gettering implies that b-Si could also influence light- and elevated temperature-induced degradation (LeTID), which is the main culprit for efficiency deterioration in today’s multicrystalline silicon (mc-Si) solar cells [7, 8]. LeTID may result in over 10 % relative efficiency loss in cells with dielectrically-passivated rear side [7, 9, 10], which is a significant issue for the PV industry that is shifting towards passivated emitter and rear cells (PERC) [2]. Hence, several PV research groups, academic institutions and companies have recently investigated methods to mitigate the detrimental phenomenon, e.g. via dark annealing [11], reduced peak temperature or ramp rates of fast firing [12-15], application of high-intensity illumination at elevated temperature [16] or phosphorus diffusion gettering (PDG) of LeTID-causing impurities [17, 18].

In this work, we apply a b-Si nanostructure produced by deep reactive ion etching (DRIE) to PERC cells fabricated on typical LeTID-sensitive mc-Si material. Industrial b-Si solar cells are processed together with acidic-textured references following the standard PERC process at the production line of SolarWorld Industries GmbH. In order to evaluate the strength of LeTID in cells with different surface texture and front surface passivation, the recombination properties of the cells and the resulting electrical performance both at material and device level are characterized, both initially and when exposed to commonly-used degradation conditions [7].

2. EXPERIMENTAL PROCEDURES

Industrial b-Si solar cells were fabricated following the standard PERC process (outlined in Fig. 1) from material, which was known to be prone to LeTID. More than one hundred industrial-size (156 x 156 mm) 180 µm thick mc-Si wafers were taken from a single column of an ingot close to each other near the ingot center. The cells were divided into groups A and B depending on the used surface texture (acidic texture or b-Si, respectively). In Group B (B1 & B2), saw damage was first removed from slurry-cut mc-Si wafers by etching ~10 µm from each side in potassium hydroxide (KOH), followed by hydrochloric acid (HCl) and hydrofluoric acid – ozone (HF/O₃) cleanings. A b-Si nanostructure was etched on the front sides by DRIE using a process based on that reported in [4]. The resulting surface morphology is presented in a scanning electron microscope (SEM) image in Fig. 1. Due to clamping against the bottom electrode, approximately 5 mm from each wafer edge remained unetched. Additionally, the adaption of the b-Si fabrication process to a larger substrate area resulted in an increased silicon consumption, and ~15 µm was removed from the wafer front surface. In Group A (A1 & A2), instead of saw damage removal (SDR) and b-Si etching, the wafers were textured acidically in hydrofluoric-nitric acid (HF/HNO₃) solution, which etched ~5 µm from both sides of the wafer. After cleaning, all Group A and B wafers were merged into a single batch, and phosphorus emitter was formed by POCl₃ diffusion. The large surface area of b-Si resulted in a heavier emitter in the b-Si wafers (~50 Ω/□ vs. ~80 Ω/□, as determined by four-point probe), which agrees with earlier reports [19-21]. Although the wafers were diffused back-to-back, emitter formation was followed by etching of ~5 µm from the rear side with HF/HNO₃ solution to ensure the absence of a phosphorus-doped layer at the rear and to enhance internal reflection. Phosphosilicate glass (PSG) was removed in HF, which was followed by cleaning with diluted HF. The cleaning was kept dilute intentionally, since standard cleaning solutions etch heavily phosphorus-doped b-Si at a rapid rate, and hence, alter the electrical and optical properties of the front surface [22]. The back side was passivated with an AlOx/SiNx stack.
Figure 1. Process flow chart. PERC cells were fabricated mainly at an industrial production line (purple outline). Black silicon was etched using a lab-scale single-wafer equipment, and ALD coating was deposited with a large-scale batch tool. The b-Si morphology is presented in a SEM image.

Front surfaces of group A2 and B2 cells were passivated with 30 nm of ALD AlO$_x$ deposited at 275 °C, whereas group A1 and B1 cells were coated with a standard industrial plasma-enhanced (remote plasma) chemical vapor deposited (PECVD) SiN$_x$ anti-reflection film. Group A2 cells were additionally capped with a PECVD SiN, coating to reduce reflectance, while the corresponding b-Si cells (group B2) were left without additional thin films. After front surface passivation and rear contact opening, all wafers were screen printed and fired at 880 °C (set point temperature). Screen dimensions were slightly reduced from the standard size according to the dimensions of the b-Si area to have metallization only on b-Si. The cells were mainly processed at the industrial production line of SolarWorld Industries GmbH with two exceptions (highlighted with different color in Fig. 1): b-Si was etched with a lab-scale (~15 min/wafer) single-wafer equipment (Oxford Instruments Plasmalab System 100) at Aalto University, and ALD AlO$_x$ was deposited using a large-scale batch tool (Beneq P400) at Beneq Oy.
The complete cells were characterized under the standard test conditions (25 °C, 1000 W/m², AM1.5 spectrum) by current-voltage (I-V) measurements, from which typical solar cell parameters were determined, including power conversion efficiency (η), open-circuit voltage (V_{oc}) and short-circuit current density (J_{sc}). The cells were degraded under 0.5 sun illumination at 75 °C for one week (~160 hours), during which the I-V characteristics were measured at regular intervals. In addition, internal quantum efficiency spectra (IQE) and photoluminescence (PL) maps were measured from each cell, both before and after the degradation.

3. RESULTS AND DISCUSSION

3.1 Quantum efficiency and bulk properties

Figure 2a presents IQE spectra of the acidic-textured cells with either SiNₓ or ALD AlOₓ/SiNₓ front surface passivation in the initial and degraded form. LeTID is clearly visible in both cells as a reduction in IQE under illumination at 75 °C. No clear difference is observed in the degradation behavior of cells with different front surface passivation layers. The reduction in IQE is the strongest in the ~850–1100 nm wavelength range, which indicates that the degradation occurs in the bulk or at the rear surface. Consistently, earlier studies have suggested that LeTID is mainly a bulk effect [12, 23, 24]. Interestingly, both surface passivation schemes produce nearly identical IQE spectra also in the short wavelength range (both before and after degradation), although it is generally understood that AlOₓ is unfavorable for phosphorus emitter passivation due to the high density of negative fixed charges in the thin film [19, 25]. The charge polarity was indeed confirmed to be negative, and hence, surface passivation is attributed to excellent chemical passivation provided by ALD AlOₓ [27].

The b-Si cells do not show such degradation. Figures 2b and 2c compare IQE of acidic-textured and b-Si cells passivated with either ALD AlOₓ or CVD SiNₓ, respectively, in the wavelength range with most significant LeTID present (~850–1100 nm). While IQE of the acidic-textured cells declines remarkably independent of the surface passivation layer, the b-Si cells nearly perfectly retain their initial performance. In particular, the AlOₓ-passivated b-Si cells show no degradation (Fig. 2b), although also they have a hydrogen-rich AlOₓ/SiNₓ dielectric stack on the rear. Since ALD AlOₓ contains only little hydrogen compared to CVD SiNₓ [28], the excellent stability of the AlOₓ-passivated b-Si cells could be at least partly explained by a lower total amount of hydrogen diffusing into the bulk to activate the LeTID-responsible defects [29, 30]. Nevertheless, b-Si cells with SiNₓ front surface passivation also show only slight degradation compared to the acidic-textured equivalents (Fig. 2c), and hence, the superiority of the b-Si cells over the acidic-textured equivalents after degradation cannot be attributed to different surface passivation films alone.

The complete IQE spectra of the b-Si cells, which are shown as insets in Figs. 2b and 2c, reveal that the b-Si cells suffer from weak blue response, which can be mainly attributed to excessively heavy emitter doping. In addition to the large number of electrically active dopants in the b-Si emitters, as indicated by the sheet resistance (~50 Ω/□ vs. ~80 Ω/□ for b-Si and acidic-textured cells, respectively), the b-Si spikes likely contain a significant amount of inactive phosphorus [19], which causes excessive SRH recombination [31]. Moreover, contrary to the acidic-textured cells, negative fixed charges in ALD AlOₓ impair the blue response of the AlOₓ-passivated b-Si cells, since surface passivation of b-Si relies more heavily on field-effect [32]. Nevertheless, the IQE of the AlOₓ-passivated b-Si cells slightly improves during the degradation treatment especially in the short wavelength range, which is most likely due to enhanced field-effect passivation induced by increased charge density in the AlOₓ thin film [28].

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Figure 2. (a) Internal quantum efficiency (IQE) of acidic-textured solar cells with front surface passivated with either CVD SiN$_x$ or ALD AlO$_x$. (b)–(c) Inset of the wavelength range with the largest IQE degradation (highlighted in (a)) for acidic-textured and b-Si solar cells with either (b) ALD AlO$_x$ or (c) CVD SiN$_x$ front surface passivation. The complete IQE spectra of the b-Si cells are presented as insets in (b) and (c). Note that the AlO$_x$-passivated acidic-textured cells have a SiN$_x$ capping. All spectra are given both directly after firing (dashed lines) and after degradation (solid lines).
The observations from IQE spectra are supported by spatially-resolved PL maps. Figure 3a shows a photograph of a final b-Si PERC solar cell, and Figs. 3b and 3c present corresponding relative PL maps for b-Si cells passivated with either ALD AlOₓ or CVD SiNₓ, respectively. Correspondingly, Figs. 3d, 3e, and 3f show the photograph and relative PL maps for the acidic-textured cells, with the exception that the cell with ALD AlOₓ front surface passivation has a CVD SiNₓ coating. The maps represent the PL intensity decrease after a one-week degradation normalized to the initial PL intensity, i.e., blue color represents no change in PL intensity, indicating no degradation in the wafer bulk, whereas red color represents the highest decrease in the PL intensity during degradation. Consistent with the IQE spectra in Fig. 2b, the AlOₓ-passivated b-Si cell completely retains its performance under illumination at elevated temperature (Fig. 3b). Simultaneously, the edges, which were not subjected to b-Si etching, show a slight degradation. Hence, hydrogen in-diffusion from the SiNₓ film on the rear is enough to activate some amount of LeTID-causing defects in the bulk on areas without b-Si. However, the degradation is more severe with SiNₓ front passivation (Fig. 3c), as expected [29, 30, 33]. Nevertheless, the b-Si area degrades only slightly, while the non-textured edges show significant deterioration. This agrees with the IQE spectra (Fig. 2c) and confirms that the reduced degradation in the b-Si cells cannot be explained only by a different amount of hydrogen in the surface passivation layers.

![Figure 3](image-url)

**Figure 3.** (a) Photograph of a 156 x 156 mm multicrystalline b-Si PERC solar cell with AlOₓ surface passivation. Approximately 5 mm from wafer edges remained without b-Si due to clamping during the b-Si etching process. (b)–(c) PL intensity change during degradation normalized to the initial intensity of an (b) ALD AlOₓ or (c) CVD SiNₓ-passivated b-Si cell. (d) Photograph of a standard acidic-textured PERC with SiNₓ anti-reflection coating processed in the same batch. (e)–(f) Normalized PL intensity change during degradation of an (e) ALD AlOₓ / CVD SiNₓ or (f) CVD SiNₓ-passivated acidic-textured cell. The same color scale applies to all maps: red color represents a large decrease in the PL intensity indicating strong degradation, while blue color represents stable PL intensity and no degradation. *Note that the acidic-textured cell in PL map (e) has an additional SiNₓ coating compared to the b-Si cell in map (b).
The PL images show also that both acidic-textured cells degrade more than either of the b-Si cells (Figs. 3e and 3f). However, contrary to the IQE spectra in Fig. 2a, the acidic-textured cell with CVD SiNₓ front surface passivation (Fig. 3f) shows on average more severe degradation than the equivalent with an ALD AlOₓ interlayer (Fig. 3e). This observation suggests that 30 nm of ALD AlOₓ is thick and dense enough to impede hydrogen diffusion from SiNₓ to the silicon bulk [34], contrary to earlier reports for CVD AlOₓ [33] or thinner ALD films [29]. Hence, hydrogen can mainly diffuse in the wafer only from the rear side through the CVD AlOₓ film, resulting in weaker LeTID. Alternatively, the charge polarity in the thin film may have an effect, since both plasma-enhanced CVD SiNₓ and ALD AlOₓ are known to getter metal impurities [35, 36]. Nevertheless, the largest degradation is observed in the non-textured edges of the SiNₓ-passivated b-Si cells, which is an interesting finding and will be discussed later in more detail. Finally, Fig. 3c reveals that the degradation is rather uniform laterally in good grain areas, which is characteristic of LeTID [12, 14]. However, defect clusters show slightly stronger LeTID, as reported recently [37, 38], which is even more pronounced in the acidic-textured cells (Figs. 3e and 3f).

3.2 Current-voltage characteristics

The lower magnitude of LeTID in b-Si cells is also reflected in cell efficiency and other essential solar cell parameters, such as \( V_{oc} \) and \( I_{sc} \). Figure 4 compares the evolution of power conversion efficiency of b-Si and acidic-textured cells normalized to the initial efficiency measured throughout a one-week degradation. As expected from the IQE spectra (Fig. 2a) and PL maps (Figs. 3e and 3f), the efficiencies of both acidic-textured cells decline significantly (nearly 4 %rel) under 0.5 sun illumination at 75 °C. Additionally, the AlOₓ interlayer has only modest effect on the degradation trend, and the AlOₓ/SiNₓ-passivated cell slightly outperforms the SiNₓ-passivated equivalent, which is in agreement with the PL maps. The magnitude and time scale of the degradation agree with earlier reports on LeTID [7, 10]. The cells show no regeneration until the end of the one-week degradation cycle, which is natural, since complete saturation is not yet reached. On the other hand, the AlOₓ-passivated b-Si cells not only retain their performance, but their efficiency even improves by \( \sim 1 \% \text{rel} \), which is most likely due to enhanced field-effect passivation as concluded earlier from the IQE behavior. Likewise, the SiNₓ-passivated b-Si cells are substantially more stable than either of the acidic-textured cells with only 1.5 %rel degradation, consistent with both IQE spectra and PL maps. Similar trends are visible also in \( V_{oc} \) and \( I_{sc} \) of the cells.

![Figure 4. Power conversion efficiency of ALD AlOₓ or CVD SiNₓ-passivated b-Si and acidic-textured cells as a function of degradation time normalized to the initial efficiency. The dotted lines act as a guide for the eye.](image-url)
Finally, Table I presents the cell parameters of the AlO₅₃ or SiNx-passivated b-Si and acidic-textured cells measured directly after firing and their absolute change during a one-week degradation. The best performance is obtained with b-Si texture and SiNx passivation, despite a slight degradation. Furthermore, both b-Si cells maintain their level of performance substantially better than either of the acidic-textured cells in all relevant solar cell parameters. However, the AlO₅₃-passivated cells suffer from increased contact resistance due to challenges in screen-printing through a 30 nm thick ALD AlO₂ layer, which is expected, as even 10 nm films have been reported to increase contact resistance losses independent of firing temperature [39]. This is seen as reduced fill factor, which is pronounced in the acidic-textured cells due to the additional SiNx film. Moreover, the excessively heavy emitter doping likely reduces the Voc of the b-Si cells, which was observed also in IQE (Figs. 2b and 2c). The effect is even more pronounced in the AlO₅₃-passivated b-Si cells due to the non-optimal charge polarity for phosphorus emitter passivation. Interestingly, both SiNₓ-passivated cells reach approximately the same Voc after degradation. Nevertheless, while also bulk quality affects Voc, the low Voc of the b-Si cells can be mainly attributed to poor electrical performance at the front surface, since the long-wavelength IQE (Fig. 2b-c) and PL maps (Figs. 3b-c and 3e-f) indicate higher bulk quality in the b-Si cells after degradation.

Table I. Cell parameters of b-Si and acidic-textured solar cells with different front surface passivation directly after firing (initial) and amount of degradation in absolute values. The reduced active area of the b-Si cells is taken into account in the reported efficiency.

<table>
<thead>
<tr>
<th></th>
<th>η (%)</th>
<th>Voc (mV)</th>
<th>Jsc (mA/cm²)</th>
<th>FF</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Initial</td>
<td>Change</td>
<td>Initial</td>
<td>Change</td>
</tr>
<tr>
<td>Black silicon</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>AlO₅₃</td>
<td>17.2</td>
<td>+0.2</td>
<td>613</td>
<td>+1.4</td>
</tr>
<tr>
<td>SiNx</td>
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<td>-0.3</td>
<td>635</td>
<td>-3.0</td>
</tr>
<tr>
<td>Acidic texture</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>AlO₅₃/SiNx</td>
<td>14.5</td>
<td>-0.6</td>
<td>636</td>
<td>-8.8</td>
</tr>
<tr>
<td>SiNx</td>
<td>18.7</td>
<td>-0.7</td>
<td>643</td>
<td>-10.7</td>
</tr>
</tbody>
</table>

The efficiency of the AlO₅₃/SiNx-passivated acidic-textured cells is further lowered by optically unfavorable total thickness of the AlO₅₃/SiNx dielectric stack, which heavily reduces Jsc. Since fair comparison between different cell types was prioritized instead of specifically optimized b-Si cell process, none of the process steps were particularly adjusted for b-Si surfaces. Hence, the performance of the b-Si cells could be further improved by optimization of e.g. emitter diffusion and front side metallization.

3.3 Physical background for impact of b-Si on LeTID

The results demonstrate that the amount of LeTID is reduced by the application of b-Si, indicating that the defect responsible for the phenomenon must be in an inactive state or removed in the b-Si cells. Next, we discuss possible explanations for the capability of b-Si to reduce LeTID.

Recent studies suggest that P or Al-gettering can suppress LeTID to some extent [17, 18], especially in material with low lifetime [37]. Zuschlag et al. showed that P-diffused SiNx-passivated mc-Si lifetime samples experienced only slight degradation, whereas a stronger and faster LeTID was observed in ungettered sister samples [17]. Chakraborty et al. presented similar results and reported that PDG reduced the concentration of light-induced defects, which were suspected to be Cu, Ni and Ti, by nearly one order of magnitude [18]. Interestingly, our previous study demonstrated that heavy phosphorus-doping in the b-Si spikes after an industry-typical POCl₃ diffusion process enhances iron segregation to the emitter, which substantially improves minority carrier lifetime in iron-contaminated substrates [6]. Consistent with earlier reports [19-21], sheet resistance is indeed substantially lower in the b-Si than in the acidic-textured cells.
(~50 Ω/□ vs. ~80 Ω/□). Thus, the reduced LeTID could be at least partly explained by a heavier phosphorus emitter, and hence enhanced gettering, in the b-Si cells [6], although the defect responsible for the degradation is likely another impurity than Fe [40, 41]. This explanation supports the conclusion that LeTID is a bulk effect, as suggested previously [12, 23, 24], instead of e.g. deterioration of rear surface passivation.

Additionally, gettering may be intensified also by accumulation of impurities at the AlOx–Si interface or into the SiNx bulk during post-deposition heat treatments [35, 36]. This may have a stronger effect in b-Si, since the large surface area of the nanostructure provides more gettering sites at the dielectric–Si interface. Furthermore, the negative fixed charges in ALD AlOx are known to generate an enhanced electric field in b-Si due to conformal coating on the high aspect ratio nanostructures [32], which may affect impurity gettering.

In addition to beneficial gettering of defects within the silicon wafer, application of b-Si may remove the deleterious impurities altogether. In typical acidic texturing, saw damage resulting from slurry-based wire sawing is used as an initiator for the texture etching process [42]. However, saw damage may contain detrimental contaminants, such as copper [43, 44], which lead to the following degradation [45]. Instead, fabrication of b-Si by DRIE is independent of the initial surface condition [46, 47], which allows a complete removal of the contaminated surface layers. In this work, saw damage was removed from the b-Si cell precursors as a separate wet-chemical etching step prior to texturization in order to remove the damage also from the wafer edges, which were not etched during b-Si fabrication. Nevertheless, since a clear contrast between the b-Si area and the unetched edges is visible in the PL maps, especially in that of the SiNx-passivated b-Si cell (Fig. 3c), the separate SDR alone cannot explain the difference between cells with different textures, but the reduced LeTID is likely a b-Si related phenomenon.

Due to the increased silicon consumption of the applied b-Si fabrication process, the final b-Si cell thickness was decreased to ~140 µm, which appears as a dip in IQE at long wavelengths (Figs. 2b and 2c). The acidic-textured cells are slightly thicker, ~165 µm, while the non-textured edges of the b-Si cells, which experienced SDR instead of acidic-texturing, have a final thickness between these two (~155 µm). Interestingly, Bredemeier et al. reported that wafer thinning reduces LeTID-causing effective defect density [41]. In our experiments, the non-textured edges of the SiNx-passivated b-Si cells show more pronounced degradation in the PL map (Fig. 3c) than the thicker acidic-textured cells (Fig. 3f). Hence, the magnitude of LeTID in our cells seems to scale rather with surface area than wafer thickness: b-Si cells with the largest surface area (ratio to polished surface $S_t \approx 5-7$ [32, 48]) experience the weakest degradation, followed by acidic-textured cells ($S_t \approx 2$ [49, 50]), and the strongest LeTID is shown at the edges of b-Si cells, which were chemically polished by SDR. Nevertheless, we want to stress that the heavy silicon consumption is a feature of the specific b-Si etching process used in this study and could be drastically reduced by further optimization of the DRIE process. The reduced IQE in the b-Si cells may be also contributed by weaker light-trapping due to the nanometer-scale dimensions of the nanostructure, which could be enhanced by coupling b-Si with micrometer-sized pyramid texture [51].

Another important aspect affecting the magnitude of LeTID is the applied firing process [12-15]. Although the cells are fired with identical set peak temperature and profile, the actual temperature reached by the wafers may vary due to differences in optics of the front surface or wafer thickness. Indeed, the higher absorbance of b-Si compared to either of the AR-coated acidic-textured surfaces indicates that b-Si cells may reach a higher temperature during firing. Correspondingly, a thinner wafer may heat up more compared to a thicker one due to the smaller volume of the wafer, although recent results [41] report thinner samples to experience weaker LeTID. Hence, the possible effect of surface texture or wafer thickness on the actual firing temperature experienced by the wafers should be visible only as more intense LeTID in the b-Si cells.

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Thus, the difference between the b-Si and acidic-textured cells could be even larger if the actual wafer
temperatures were exactly identical. Furthermore, since the set peak firing temperature was 880 °C, it is
highly unlikely that any of the cells would have experienced a temperature lower than the threshold for
LeTID (720-740 °C [52, 53]).

4. CONCLUSIONS

This work demonstrated that a b-Si nanostructure reduces the amount of LeTID in industrial p-type mc-Si
PERC. Acidic-textured reference cells showed clear degradation in IQE independent of charge polarity in
the front surface passivation film (SiNₓ or AlOₓ/SiNₓ), especially in the ~850–1100 nm wavelength range,
which suggests that the degradation occurred in the bulk. Instead, the b-Si cells retained their performance
nearly perfectly, despite the presence of a hydrogen-rich AlOₓ/SiNₓ passivation stack on the rear. Especially,
the b-Si cells with ALD AlOₓ front surface passivation showed no degradation, which may be in part due
to a lower total amount of in-diffused hydrogen in the bulk, preventing the activation of LeTID-responsible
defects [28, 29]. Furthermore, the presence of a hydrogen-rich SiNₓ film on the front induced only slight
degradation in the b-Si cells compared to the acidic-textured equivalents. The observations were supported
by spatially-resolved PL imaging, which showed a clear contrast between the b-Si and acidic-textured cells
after degradation, especially in the case of SiNₓ front surface passivation.

The lower magnitude of LeTID eventually reflected to the power conversion efficiency of the b-Si solar
cells. The efficiency of the AlOₓ-passivated b-Si PERC increased by 1 %rel under a one-week illumination
at 75 °C, most likely due to enhanced front surface passivation. Simultaneously, the SiNₓ-passivated b-Si
cells lost only 1.5 %rel of their initial efficiency, while the acidic-textured equivalents degraded by nearly 4
%rel. Although the exact mechanism for LeTID mitigation by b-Si remained unresolved, the phenomenon
could be contributed by the strong ability of b-Si to getter LeTID-causing impurities. Indeed, the PL maps
suggested that the magnitude of LeTID scales with surface area. Furthermore, hydrogen-rich SiNₓ anti-
reflection coatings can be replaced by ALD AlOₓ due to the negligible reflectance of b-Si, which aids to
suppress LeTID efficiently. Since the majority of the cell processing was performed at an industrial
production line, the results should be of direct interest to PV industry.

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