Surface passivation by Al2O3 and a-SiNx: H films deposited on wet-chemically conditioned Si surfaces

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Surface Passivation by Al$_2$O$_3$ and a-SiN$_x$: H Films Deposited on Wet-Chemically Conditioned Si Surfaces

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The surface passivation of $p$- and $n$-type silicon by different chemically grown SiO$_2$ films (prepared by HNO$_3$, H$_2$SO$_4$/H$_2$O$_2$, and HCl/H$_2$O$_2$ treatments) was investigated after PECVD of a-SiN$_x$:H and ALD of Al$_2$O$_3$ capping films. The wet chemically grown SiO$_2$ films were compared to thermally grown SiO$_2$ and the surface passivation was benchmarked against single-layer a-SiN$_x$:H and Al$_2$O$_3$ films deposited on HF-last Si surfaces. Directly after PECVD and ALD the surface passivation was found to be similar for all chemically grown SiO$_2$ stacks and higher than for the single-layer films. After annealing at 400°C the level of surface passivation improved but remained similar for the chemically grown SiO$_2$ stacks. Corona charging experiments on the annealed chemically grown SiO$_2$/Al$_2$O$_3$ stacks revealed that the fixed charge density, i.e., the field-effect passivation, was lower compared to single-layer Al$_2$O$_3$ films whereas the chemical passivation was found to be higher. The surface passivation was also investigated as a function of the annealing time. Chemically grown SiO$_2$ stacks achieved the maximum level of surface passivation after annealing for half a minute which is different from the stacks with a-SiN$_x$:H and Al$_2$O$_3$ films deposited on thermally grown SiO$_2$ or HF-last Si surfaces which required considerably longer annealing times.

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Experimental

The surface passivation was evaluated on phosphorous doped 2-3 Ω·cm $n$-type float zone (FZ) wafers with a thickness of 200 μm and on boron doped 10-12 Ω·cm $p$-type Czochralski (Cz) Si(100) wafers with a thickness of 160 μm. The saw-damage of the Cz wafers was removed in KOH. Afterwards the $p$- and $n$-type wafers received a standard RCA cleaning procedure. The wafers passivated by the direct deposition of single-layer films on the H-terminated Si surfaces resulting from the RCA cleaning are referred to as “HF-last” for brevity. The rest of the wafers were exposed to a wet chemical or thermal oxidation process. For the wet chemical oxidation process three different solutions were used: 65% nitric acid (HNO$_3$), hydrochloric acid with hydrogen peroxide (HCl/H$_2$O$_2$) with 1:1:10 at 80°C and a sulphuric acid mixture with hydrogen peroxide (H$_2$SO$_4$/H$_2$O$_2$) with 1:5 at 140°C. The samples were treated for 10 min with the wet chemical solutions. The thermal oxidation took place at 900°C in wet ambient and resulted in a film thickness of 200 nm as measured by spectroscopic ellipsometry. All wafers were covered by PECVD a-SiN$_x$:H films or ALD Al$_2$O$_3$ films. The a-SiN$_x$:H films were deposited at 350°C by a linear microwave plasma source. The a-SiN$_x$:H film thickness was 70 nm and the refractive index was $n = 2.05$ (at 2 eV) as...
determined by spectroscopic ellipsometry. By Rutherford backscattering spectroscopy the ratio of [N]/[Si] was found to be 1.14. By elastic recoil detection the atomic hydrogen concentration was found to be 9 at.%. The Al2O3 films were deposited by an ALD process at 220°C using Al(CH3)3 and O3 as reactants. The process consisted of 330 ALD cycles that resulted in a film thickness of about 30 nm. One set of wafers was annealed at 400°C in N2 atmosphere for 10 min. In addition, an experiment was carried out in which the annealing time was varied between 30 s and 20 min.

The level of surface passivation was expressed by the injection dependent effective minority carrier lifetime \( \tau_{\text{eff}} \) that was measured with the photoconductance decay method using a Sinton instrument WCT 120 system. The surface recombination velocity \( S_{\text{eff}} \) was calculated from the effective lifetime at an injection level of \( \Delta n = 10^{14} \text{ cm}^{-2} \). The \( S_{\text{eff}} \)-values were evaluated as an upper limit, i.e. as \( S_{\text{Auger}} = W/2 \cdot (\tau_{\text{eff}} - \tau_{\text{Auger}})^{-1} \), with \( W \) representing the wafer thickness and by considering the Auger lifetime \( \tau_{\text{Auger}} \) as proposed by Kerr.

The density of the fixed charges \( Q_f \) incorporated in the dielectric passivation films was measured by corona charging experiments. This method combines the deposition of corona charges with Kelvin probe measurements of the surface voltage and photoconductance decay measurements to extract the corona charge density \( Q_C \) and the corresponding surface passivation, respectively. The accuracy of the measurement is related to the uniformity of the deposited corona charges which was fairly good, i.e., 10–15% within a circle with a diameter of 12 cm. The position of the peak value of \( S_{\text{eff,max}} \) in the \( S_{\text{eff,max}}(Q_C) \)-plot reflects the point where \( Q_C \) is compensated by the deposited corona charges. From this value of \( Q_C \), the number density of \( Q_f \) incorporated in the passivation films can be extracted. The \( S_{\text{eff,max}} \)-values obtained in the corona charging experiments were calculated from the effective lifetime at an injection level of \( \Delta n = 3 \cdot 10^{14} \text{ cm}^{-2} \) to assure accurate photoconductance decay measurements also at low lifetime values.

The field-effect passivation was also studied with the method of optical second harmonic generation (SHG) that is sensitive to electric fields. Electric fields are present in the space charge region at the Si surface, e.g. created by the \( Q_f \) of passivation films. The amplitude of the SHG intensity is related to the SHG photon energy of 3.4 eV is a measure for the electric field at the Si surface when probing Si surfaces covered by dielectric films such as Al2O3.

### Results and Discussion

**Surface passivation quality of Al2O3 and a-SiNx:H stacks.**—The surface recombination velocity as a measure for the surface passivation quality was determined for Al2O3 and a-SiNx:H films with chemically and thermally grown SiO2 films and for single-layer Al2O3 and a-SiNx:H films deposited on HF-last Si surfaces. In Table I the \( S_{\text{eff,max}} \)-values are listed as obtained directly after deposition of the Al2O3 and a-SiNx:H films and after annealing at 400°C. The measurement uncertainty of the \( S_{\text{eff,max}} \)-values was about 15%. The thermally grown SiO2/Al2O3 stacks resulted in low \( S_{\text{eff,max}} \)-values of 84 cm/s and the chemically grown SiO2/Al2O3 stacks in \( S_{\text{eff,max}} \)-values between 22 and 324 cm/s directly after ALD process. No clear correlation between the used wet chemical treatments and the level of surface passivation was observed. The \( S_{\text{eff,max}} \)-values directly after Al2O3 deposition on HF-last Si surfaces were significantly higher, up to 610 cm/s, than those for the chemically grown SiO2/Al2O3 stacks. For the Al2O3 coated, HF-last p-type Cz samples a larger spread in the \( S_{\text{eff,max}} \)-values was observed than for the HF-last n-type Fz samples. Some values for this p-type Si were similar to those obtained for the chemically grown SiO2/Al2O3 stacks. The can possibly be attributed to a longer time delay between the HF dip and deposition of the Al2O3 films for these samples causing a growth of native oxide films that have a thickness similar to those of the chemically grown SiO2 films of this work. Apparently a thin chemically grown or natively grown SiO2 film underneath the Al2O3 layers can result in similar \( S_{\text{eff,max}} \)-values.

Annealing of the chemically grown SiO2/Al2O3 stacks decreased the \( S_{\text{eff,max}} \)-values and resulted in quite similar values ranging between 2 and 14 cm/s. The thermally grown SiO2/Al2O3 stacks showed a similar level of \( S_{\text{eff,max}} \) of 14 cm/s. The uniformity of the surface passivation of the chemically grown SiO2/Al2O3 stacks was about 10 to 15% as obtained by photoluminescence imaging using the 6 inch p-type Cz samples. Single-layer Al2O3 films deposited on HF-last Si surfaces showed low surface recombination velocities of \( S_{\text{eff,max}} \) = 2–6 cm/s after anneal. These surface recombination values are in the same range as those obtained in earlier studies, e.g. reported by Hoex et al. with \( S_{\text{eff,max}} = 2 \text{ cm/s} \).

For the samples with a-SiNx:H films the highest recombination velocity was obtained for single-layer films deposited on HF-last Si surfaces with approx. 400–460 cm/s achieved directly after deposition. The chemically grown SiO2/a-SiNx:H stacks gave \( S_{\text{eff,max}} \)-values below 47 cm/s and the thermally grown SiO2/a-SiNx:H stacks yielded a slightly higher value of 67 cm/s. The annealing improved the level of surface passivation of all a-SiNx:H stacks. The best surface passivation was observed for the chemically grown SiO2/a-SiNx:H stacks with \( S_{\text{eff,max}} \)-values down to 4 cm/s.

The comparison of the different a-SiNx:H and Al2O3 stacks revealed that before annealing the a-SiNx:H-based stacks resulted in lower recombination velocities compared to the Al2O3 based stacks. Among other things, this experimental finding can probably be attributed to the different deposition temperatures for PECVD and ALD. PECVD was carried out at 350°C, i.e. at only 50°C lower than the annealing temperature, whereas ALD was carried out at 200°C. Higher deposition temperatures enable a more effective hydrogenation process of the Si surface and consequently enhance the deactivation of Si dangling bonds that otherwise act as recombination active centers.

When comparing the surface passivation results of the chemically grown SiO2/Al2O3 stacks with the a-SiNx:H based stacks after annealing a quite equal level of surface passivation can be observed. No correlation between wet chemical solution used for the growth of the SiO2 films and the surface passivation quality can be distinguished.

In the case of HF-last Si surfaces the annealed Al2O3 films showed higher passivation qualities than the a-SiNx:H films. The chemical passivation after annealing was reported to be similar for Al2O3 and a-SiNx:H films with mid-gap defect densities of about 10^11 eV^-1 cm^-2. In contrast to the chemical passivation, the field-effect passivation was found to be different as evident from the measured

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Table I. Maximum surface recombination velocity \( S_{\text{eff,max}} \) (cm/s) extracted from the effective lifetime \( \tau_{\text{eff}} \) at \( \Delta n = 10^{14} \text{ cm}^{-2} \) for wet chemically grown and thermally grown SiO2 as well as HF-last Si substrates. The \( S_{\text{eff,max}} \)-values are obtained directly after ALD of Al2O3 and PECVD of a-SiNx:H (referred to “As-dep.”) and after an additional anneal step at 400°C for 10 min in N2. The measurement uncertainty of the \( S_{\text{eff,max}} \)-values was \( \sim 15\% \).
fixed charge densities of $Q_{C} = 2 \pm 0.5 \cdot 10^{12}$ cm$^{-2}$ for $\alpha$-SiN$_{x}$H films and $Q_{C} = -6.5 \pm 0.5 \cdot 10^{12}$ cm$^{-2}$ for Al$_2$O$_3$ films. These values were obtained by corona charging experiments on the films considered in this work (see below). Consequently the difference in surface passivation of single-layer $\alpha$-SiN$_{x}$H and Al$_2$O$_3$ films can be attributed mainly to differences in field-effect passivation.

Surface passivation mechanism of Al$_2$O$_3$ stacks.— For the chemically grown and thermally grown SiO$_2$ films in stacks with Al$_2$O$_3$ the layer structure is Si/SiO$_2$/Al$_2$O$_3$. This is basically also the case for single-layer Al$_2$O$_3$ films deposited on HF-last Si surfaces because the deposition of the Al$_2$O$_3$ leads to a thin interfacial SiO$_2$ film with a typical thickness of about 1.2 nm. The thickness of the interfacial SiO$_2$ film is in the same range as those of the chemically grown SiO$_2$ films which have thickness values of 0.5-2 nm as found by spectroscopic ellipsometry. Recently it was reported that even very thin interfacial SiO$_2$ films in the range of a few nanometers have a significant effect on the field-effect passivation of Al$_2$O$_3$ films. Consequently it was decided to investigate the surface passivation mechanisms of the Al$_2$O$_3$ stacks reported in this work by corona charging experiments in order to distinguish between the role of chemical and field-effect passivation. Corona charging could not be employed successfully to the samples coated with $\alpha$-SiN$_{x}$H films due to the too low electrical resistivity of the $\alpha$-SiN$_{x}$H films which preclude the surface charging of these films by corona charges.

Corona charging experiments were carried out on the $n$-type Fz samples preclude directly after the ALD process as well as after annealing. However directly after ALD the corona charging experiments turned out not to be feasible on single-layer Al$_2$O$_3$ films and stacks with SiO$_2$ synthesized in HNO$_3$ because the $\tau_{\text{eff}}$-values were too low for an accurate measurement of the photoconductive decay method that is part of the corona charging experiments. The corona charging results of the chemically grown SiO$_2$/Al$_2$O$_3$ stacks are shown in Fig. 1 as a plot of $S_{\text{eff, max}}$ as a function of $Q_{C}$. The change of $S_{\text{eff, max}}$ with $Q_{C}$ reflects the transition of the surface carrier concentration from inversion conditions via flatband conditions to accumulation conditions. This transition is caused by the $Q_{C}$ that compensates the $Q_{C}$ of the passivation layers. In the case of flatband conditions the surface recombination is at maximum leading to a peak value of $S_{\text{eff, max}}$ in the $S_{\text{eff, max}}(Q_{C})$-plot. At this point the peak value of $S_{\text{eff, max}}$ can be regarded as a measure for the chemical passivation due to absence of the field-effect passivation. Furthermore the nullification of the field-effect passivation with $Q_{C} = -Q_{C}$ allows the extraction of the number density of the fixed charges in the passivation material. The H$_2$SO$_4$/H$_2$O$_2$-based SiO$_2$/Al$_2$O$_3$ stacks resulted in a peak value of $S_{\text{eff, max}}$ of about 1670 cm/s and a $Q_{C}$ of $-2.6 \pm 0.2 \cdot 10^{12}$ cm$^{-2}$ as shown in Fig. 1. The HCl/H$_2$O$_2$-based SiO$_2$/Al$_2$O$_3$ stacks showed a peak value of $S_{\text{eff, max}}$ of about 1200 cm/s and a $Q_{C}$-value of $-2.4 \pm 0.2 \cdot 10^{12}$ cm$^{-2}$. Therefore the level of field-effect passivation and chemical passivation were found to be quite equal for both as-deposited chemically grown SiO$_2$/Al$_2$O$_3$ stacks.

Annealing of the three chemically grown SiO$_2$ stacks resulted also in quite similar values of $Q_{C}$ and peak values of $S_{\text{eff, max}}$. The $Q_{C}$-values were in the range of $-4.4 \cdot 10^{12}$ to $-4.6 \cdot 10^{12}$ cm$^{-2}$ and the peak values of $S_{\text{eff, max}}$ between 230 and 260 cm/s, as shown in Fig. 1. Both the chemical passivation and field-effect passivation of the chemically grown SiO$_2$/Al$_2$O$_3$ stacks improved with annealing. To verify the increase in field-effect passivation SHG measurements were carried out. The SHG intensity at a photon energy $E_{\text{ph}} = 3.4$ eV is dominated by field-induced second harmonic generation and is therefore a reasonable measure of the field-effect passivation. As shown by Terlindен et al. differences in the field-effect passivation of Al$_2$O$_3$ films can be analyzed with SHG measurements. In Fig. 2 the SHG intensity is depicted as a function of the SHG photon energy for as-deposited and annealed chemically grown SiO$_2$/Al$_2$O$_3$ stacks.

The SHG intensity at $E_{\text{ph}} = 3.4$ eV was found be similar for the three different chemically grown SiO$_2$/Al$_2$O$_3$ stacks measured directly after the ALD process as well as measured after annealing. Furthermore an increase in SHG intensity, and therefore in field-effect passivation, was observed after annealing. Consequently both corona charging and SHG experiments revealed that the field-effect passivation obtained by the wet chemically grown SiO$_2$ films was similar directly after the ALD process and after annealing and that the field-effect passivation improved by the annealing process.

In Fig. 3 the results of the corona charging experiments of the chemically grown SiO$_2$/Al$_2$O$_3$ stacks are compared to the results...
obtained for the thermally grown SiO$_2$/Al$_2$O$_3$ stacks and Al$_2$O$_3$ films deposited on HF-last Si surfaces. All samples were annealed at 400°C. The $S_{\text{eff,max}}(Q_C)$-plot obtained for Al$_2$O$_3$ films deposited on HF-last Si surfaces resulted in $Q_C$-values of $-6.5 \pm 0.5 \cdot 10^{12}$ cm$^{-2}$ which was about 1.5 times higher than the $Q_C$-value of the chemically grown SiO$_2$/Al$_2$O$_3$ stacks. However the maximum value of $S_{\text{eff,max}}$ of 290 cm/s was about a factor of 1.6 higher than for the chemically grown SiO$_2$ stacks. Consequently the chemically grown SiO$_2$/Al$_2$O$_3$ stacks showed a lower level of field-effect passivation but a higher level of chemical passivation than the Al$_2$O$_3$ films deposited on HF-last Si surfaces. Nevertheless both passivation materials resulted in a similar overall level of passivation as shown in the previous section. Stacks with thermally grown SiO$_2$ films yield a significantly better chemical passivation after annealing with peak values of $S_{\text{eff,max}} = 76$ cm/s. The field-effect passivation of the thermally grown SiO$_2$/Al$_2$O$_3$ stacks is of minor importance as this stack has the lowest fixed charge density, i.e. $Q_C = 4.6 \cdot 10^{11}$ cm$^{-2}$. Moreover the polarity of the thermally grown SiO$_2$/Al$_2$O$_3$ stacks was found to be positive in contrast to the results with the chemically grown SiO$_2$ stacks. A similar $Q_C$ and a similar high level of chemical passivation has also been reported for single-layer thermally grown SiO$_2$ films, e.g. in Ref. 3. The positive polarity of $Q_C$ for the stacks with thick (200 nm) thermally grown SiO$_2$ is in agreement with recently published results of ALD SiO$_2$/Al$_2$O$_3$ stacks. In that work the authors showed that with increasing SiO$_2$ thicknesses the polarity of $Q_C$ changed from negative to positive. The reason is that for SiO$_2$ thicknesses larger than approximately 10 nm the positive fixed charge density of the SiO$_2$ bulk starts dominating the field-effect passivation. The data presented here are perfectly in line with this observation because the SiO$_2$/Al$_2$O$_3$ stack with thin (<5 nm) SiO$_2$ films also resulted in a negative fixed charge density whereas for thick (200 nm) SiO$_2$ films a positive fixed charge density was obtained. Moreover the results clearly demonstrate that in all cases a similar overall level of surface passivation quality was obtained for the different films and stacks but that nevertheless the surface passivation mechanism in terms of chemical and field-effect passivation were significantly different.

**Improvement of surface passivation with annealing time.**— Commonly lifetime samples including Al$_2$O$_3$ films are exposed to ~400°C at annealing times between 10 and 30 min to activate the surface passivation. In this work successive anneal steps were applied at $T = 400°C$ to investigate the improvement of surface passivation during annealing at shorter times, in particular for annealing times less than 10 min for both the Al$_2$O$_3$ and a-Si$_3$N$_4$:H stacks. For comparison purposes, the $S_{\text{eff,max}}$-values at different annealing times $t_{\text{ann}}$ were normalized to the $S_{\text{eff,max}}$-values obtained after depositing the Al$_2$O$_3$ or a-Si$_3$N$_4$:H films. The normalized $S_{\text{eff,max}}(t_{\text{ann}})$-plots are shown in Fig. 4 for the Al$_2$O$_3$ stacks and in Fig. 5 for the a-Si$_3$N$_4$:H stacks.

**Annealing of Al$_2$O$_3$ stacks.**—The chemically grown SiO$_2$/Al$_2$O$_3$ stacks showed the fastest improvement of the surface passivation with $t_{\text{ann}}$ and the lowest $S_{\text{eff,max}}$-values were already obtained after annealing for 30 s. Annealing of single-layer Al$_2$O$_3$ films deposited on HF-last Si surfaces resulted in a relatively slow improvement of the passivation level. After annealing for 2 min the lowest $S_{\text{eff,max}}$-values were reached. Moreover, in contrast to the chemically grown SiO$_2$ stacks first an increase in the normalized $S_{\text{eff,max}}$-values was observed after 30 s annealing time. Thermally grown SiO$_2$ stacks showed an even slower reduction of the normalized $S_{\text{eff,max}}$-values compared to the single-layer Al$_2$O$_3$ films and chemically grown SiO$_2$ stacks. This slower reduction of the normalized $S_{\text{eff,max}}$-values could be attributed to a small improvement of the chemical passivation as revealed by corona charging experiments. The improvement of the chemical passivation with $t_{\text{ann}}$ is illustrated by the reduction of the peak values of the $S_{\text{eff,max}}$ in Fig. 6.

An initial increase in surface recombination was only observed for single-layer Al$_2$O$_3$ films. The origin of that effect could not be determined by corona charging experiments. Therefore $C-V$ measurements and infrared spectroscopy were used to investigate the field-effect passivation and the chemical passivation, respectively. With $C-V$ measurements a flatband voltage shift was observed that indicated that the fixed charge density, i.e. the field-effect passivation, steadily improved with increasing annealing time. The infrared spectroscopy data revealed that the density of Si-H stretching bonds initially decreased after annealing for 30 s and afterwards increased with annealing times leading to the highest absorption values after 10 min. The density of Si-H bonds can be only present at the Si interface and is consequently taken as a measure for the chemical passivation. The change in infrared absorption is seen as an indirect evidence for

![Figure 4](image-url)  
**Figure 4.** The change of the surface passivation performance, expressed in terms of normalized maximum surface recombination velocity $S_{\text{eff,max}}$, with annealing time $t_{\text{ann}}$ for Al$_2$O$_3$ stacks. The $S_{\text{eff,max}}$-values were normalized to the values obtained directly after the ALD process ($t_{\text{ann}} = 0$ min). Annealing took place at 400°C in N$_2$.

![Figure 6](image-url)  
**Figure 6.** The surface recombination velocity $S_{\text{eff,max}}$ for thermally grown SiO$_2$/Al$_2$O$_3$ stacks as a function of deposited corona charge density $Q_C$ after annealing at 400°C for various annealing times.
an initial decrease in chemical passivation that improved again after longer annealing times. As the field-effect passivation was found to be constant it is suggested that the change in chemical passivation is the reason for the change in the overall surface passivation with annealing time.

**Annealing of a-SiNx:H stacks.**—The improvement of the surface passivation with annealing time of the a-SiNx:H stacks was found to be similar to the one achieved for the Al2O3 stacks. Again the stacks with chemically grown SiO2 films showed the fastest improvement and the lowest values of $S_{t, max}$ were already achieved after 30 s. In the case of single-layer a-SiN$\textsubscript{x}$ and Al2O$\textsubscript{3}$ films, suggests that (de)passivation of the Si surface happens with two timescales: a fast depassivation of Si interface bonds during annealing at short times and a longer-term passivation of the Si dangling bonds by hydrogen provided from the bulk of the a-SiNx:H and Al2O3 films. The latter largely compensates the initial depassivation such that the overall surface recombination decreases by annealing. More data is however necessary to further support this hypothesis.

**Conclusions**

The effect of three wet chemically grown SiO2 films on the surface passivation of Al2O3 and a-SiN$\textsubscript{x}$H films was studied after the deposition processes of the films as well as after annealing. Directly after ALD of Al2O3 and PECVD of a-SiN$\textsubscript{x}$H the stacks with chemically grown SiO2 films showed the lowest recombination values compared to single-layer Al2O3 and a-SiN$\textsubscript{x}$H films deposited on HF-last Si surfaces. Moreover, for the stacks with wet chemically grown SiO2 films the surface passivation was found to be rather independent of the used wet chemical solutions, both directly after the deposition processes and after annealing. For the annealed films the lowest recombination was found for single-layer Al2O3 films but the slightly higher level of surface recombination velocity obtained for the chemically grown SiO2 films is still appropriate for the application of these stacks in solar cell devices. Moreover, although a similar overall level of surface passivation quality was achieved for chemically grown SiO2/Al2O3 stacks and single-layer Al2O3 films, the level of chemical and field-effect passivation differed significantly as concluded from corona charging and second harmonic generation measurements. A better chemical passivation but lower field-effect passivation was found for the chemically grown SiO2/Al2O3 stacks compared to single-layer Al2O3 films. It was also shown that the commonly used relatively long annealing times can be replaced by annealing times in the range of one or a few minutes when stacks with chemically grown SiO2 films are used.

The high level of surface passivation obtained for chemically grown SiO2-based stacks with Al2O3 or a-SiN$\textsubscript{x}$:H films offers the possibility to straightforwardly establish stable surface conditions, i.e. the formation of ultrathin SiO2 films, prior to the deposition of the passivation layers without significantly comprising the passivation quality. Therefore this kind of surface conditioning might be preferred under industrial circumstances when working with (unstable) H-terminated surface conditions is not preferred. Moreover, such stacks allow for the use of shorter annealing times to reach the maximum level of surface passivation. Also this is beneficial from an industrial point of few.

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