Silicon surface passivation by atomic layer deposited Al2O3

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I. INTRODUCTION

Surface passivation of crystalline silicon (c-Si) is becoming increasingly important in photovoltaics. Evidently, an excellent level of surface passivation is a requisite for high-efficiency c-Si solar cells. However, also for more conventional solar cells, which experience an increasing surface-to-volume ratio due to the cost-driven reduction of the cell thickness, surface passivation is becoming of vital importance. Recombination losses at the c-Si surface can be reduced by two fundamentally different strategies. Because the recombination rate is proportional with the interface defect density, the first strategy is based on the reduction of the number of defects states at the semiconductor surface. The surface defect density can, for example, be lowered by chemical passivation of unsaturated Si bonds (i.e., dangling bonds) by atomic hydrogen. The second strategy is based on the fact that surface recombination processes involve both electrons and holes. By the application of an internal electric field below the c-Si surface, either the electron or hole concentration can dramatically be reduced at the position of the c-Si surface. Consequently, surface recombination is significantly reduced and the surface recombination velocity roughly scales with the minority carrier concentration that is present at the c-Si surface. Both surface passivation strategies can be adopted by the application of thin films on the c-Si surface.

Currently three thin film materials are industrially employed for surface passivation of c-Si solar cells. The state of the art in surface passivation is obtained by thin films of thermally grown silicon dioxide (thermal SiO$_2$) and effective surface recombination velocities of 2 and 12 cm/s have been reported on 1.5 \( \Omega \) cm n-type and 1 \( \Omega \) cm p-type c-Si, respectively.\textsuperscript{1} The excellent level of surface passivation can mainly be attributed to the very high interface quality of thermal SiO$_2$ on c-Si. This high interface quality is warranted by the fact that the SiO$_2$ is thermally grown into the c-Si wafer at elevated temperatures (in the range of 900–1100 °C). Hence, the final interface quality remains rather unaffected by the initial c-Si surface condition. The interface quality is generally further improved by extensive postdeposition processing involving an annealing in a H$_2$ containing atmosphere (e.g., a forming gas annealing). The best interface quality is, however, obtained when first a sacrificial Al film is deposited on the SiO$_2$ film followed by an annealing at 450 °C (the so-called “alnealing”).\textsuperscript{2} Moreover also a small positive fixed charge density (\(10^{10}–10^{11} \text{ cm}^{-2}\)) is present in the thermal SiO$_2$ layer that provides field-effect passivation by shielding holes from the c-Si surface.\textsuperscript{3} Due to the high processing temperatures required to thermally grow a SiO$_2$ film, only c-Si with a low impurity content such as float zone (FZ) Si can be used without significant bulk lifetime degradation.\textsuperscript{4} As a consequence, thermal SiO$_2$ is currently only applied in the manufacturing of high-efficiency solar cells in which low impurity c-Si wafers are employed.\textsuperscript{5}

Silicon nitride (a-SiN$_x$:H) is currently the industry standard for surface passivation of c-Si solar cells. Effective surface recombination velocities as low as 6 and 15 cm/s have been reported on 1.5 \( \Omega \) cm n-type and 1.5 \( \Omega \) cm p-type c-Si, respectively.\textsuperscript{6,7} The surface passivation mechanism of a-SiN$_x$:H is mainly based on field-effect passivation induced by the high positive fixed charge density in a-SiN$_x$:H films (\(\sim10^{12} \text{ cm}^{-2}\)).\textsuperscript{8} The positive fixed charge density is particularly beneficial for the passivation of highly doped n-type c-Si because the minority carriers (i.e., the holes) are effec-

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tively shielded from the c-Si surface. On the other hand, the high positive fixed charge density can be detrimental for surface passivation of highly doped p-type c-Si surfaces because such a high positive fixed charge density increases the minority electron concentration at a highly doped p-type c-Si surface and consequently increases the recombination rate.\(^9\) Moreover, the high positive fixed charge density of a-SiNx:H can induce a parasitic shunting when applied at the rear of a p-type solar cell.\(^10\) Recently, it was, however, reported that it is possible to effectively passivate highly doped p-type c-Si surfaces by Si-rich a-SiNx:H after a 4 h post-deposition annealing. The underlying mechanism in this case is still under investigation.\(^11\) Very recently, Mihaietchi \textit{et al.} demonstrated that a nitric acid pretreatment could also significantly improve the level of surface passivation provided by a-SiNx:H on highly p-doped c-Si emitters.\(^12\)

Amorphous silicon (a-Si:H) is the third material that is used in industry for surface passivation of c-Si. Films of a-Si:H are employed in the heterojunction with intrinsic thin film (HIT) c-Si solar cell design of Sanyo.\(^13\) In HIT cells the p-n junction is not formed by conventional dopant diffusion but by the deposition of a doped a-Si:H film on a lightly doped c-Si wafer. An ultrathin intrinsic a-Si:H film between the c-Si and the doped a-Si:H film effectively passivates the c-Si surface and consequently exceptionally high open-circuit voltages in excess of 700 mV can be obtained by HIT solar cells.\(^13\) The good surface passivation by a-Si:H can mainly be attributed to the excellent electrical quality of the interface between c-Si and the a-Si:H film.\(^14,15\) Additionally field-effect passivation plays a role as a result from the amphoteric nature of the Si dangling bond defects.\(^14,15\) a-Si:H has also successfully been applied for surface passivation on the rear side of diffused emitter c-Si solar cells.\(^16\) The main drawbacks associated with the use of a-Si:H as surface passivation layer are the high absorption in the ultraviolet (UV) part of the solar spectrum and the low thermal stability of the a-Si:H films.

From the properties and performance of thermal SiO\(_2\), a-SiNx:H, and a-Si:H, it is clear that there is still interest in alternative surface passivation layers for c-Si solar cells. A material that recently gained interest for c-Si surface passivation is the amorphous aluminum oxide (Al\(_2\)O\(_3\)). Already in the late 1980s, Al\(_2\)O\(_3\) was applied for c-Si surface passivation in metal-insulator-semiconductor (MIS) solar cells by Hezel and Jaeger.\(^8\) They demonstrated that Al\(_2\)O\(_3\) could provide a reasonable level of surface passivation with an effective surface recombination velocity of \(~200\) cm/s on 2 \(\Omega\) cm p-type c-Si. The c-Si surface passivation was explained by a relatively low midgap interface defect density of \(~1\times10^{11}\) eV\(^{-1}\) cm\(^{-2}\) and a negative fixed charge density of \(~3\times10^{12}\) cm\(^{-2}\). The interface defect density of the c-Si/Al\(_2\)O\(_3\) system was also found to be unaffected by UV radiation,\(^8\) in contrast to the strong increase in the interface defect density observed under the UV exposure of both a-SiNx:H and thermal SiO\(_2\) on c-Si.\(^8,17-19\) As a matter of fact, the negative fixed charge density in Al\(_2\)O\(_3\) was even observed to increase during UV exposure due to the photon-induced electron injection. Hence, the UV exposure improved the field-effect passivation provided by the Al\(_2\)O\(_3\) film.\(^8\) Recently Al\(_2\)O\(_3\) was reinvestigated as surface passivation layer, now with the Al\(_2\)O\(_3\) synthesized by atomic layer deposition (ALD). Agostinelli \textit{et al.} and Hoex \textit{et al.} demonstrated that ALD Al\(_2\)O\(_3\) can provide an excellent level of surface passivation on p-type c-Si with effective surface recombination velocities close to 10 cm/s on 2 \(\Omega\) cm p-type c-Si.\(^20,21\)

The aim of this paper is to give an overview of the properties of ALD Al\(_2\)O\(_3\) films to evaluate their potential for the application as surface passivation film in future c-Si solar cells. The recently reported data set is therefore extended by the results obtained for the n- and p-type c-Si of various doping concentrations and data are presented for Al\(_2\)O\(_3\) films deposited on two different ALD reactors as well as for both plasma-assisted and thermal ALD Al\(_2\)O\(_3\). The level of surface passivation has been determined from techniques based on photocurrent, photoluminescence, and infrared emission and the specific merits of the techniques as well as the (good) agreement between their results are discussed. A key result of this work is that the field-effect passivation by the high negative fixed charge density in Al\(_2\)O\(_3\) has been demonstrated experimentally and the consequences for the passivation of highly doped n- and p-type c-Si will be addressed.

This paper is organized as follows: In Sec. II, the method of ALD for the deposition of Al\(_2\)O\(_3\) is discussed as the technique is not yet common in the field of c-Si solar cell fabrication. In Sec. III, the experimental details are given and especially the three techniques employed to measure the level of surface passivation on c-Si are described in more detail (Sec. III C). Subsequently, the experimental results are presented in Sec. IV. The level of surface passivation on n-type (Sec. IV A) and p-type c-Si (Sec. IV B) passivated by Al\(_2\)O\(_3\) will be presented for various doping concentrations. Finally, a brief comparison between plasma-assisted and thermal ALD Al\(_2\)O\(_3\) films is presented for low resistivity p-type c-Si (Sec. IV C). The conclusions of the work are described in Sec. V. In the Appendix, the difference observed between the effective lifetimes determined from photocurrent and photoluminescence for nearly intrinsic n-type c-Si will be addressed.

II. ATOMIC LAYER DEPOSITION OF Al\(_2\)O\(_3\)

ALD is a chemical vapor deposition-like method in which the deposition is controlled at the atomic level by self-limiting surface reactions. The substrate surface is alternately exposed to gas phase precursors which react with the surface species until the surface is saturated. After such a self-limiting half-reaction, the reactor is purged and/or pumped down to remove the reactants and the reaction products from the reactor such that the next half-reaction can be started. By carrying out these half reactions in a cyclic manner, ultrathin films can be synthesized with precise thickness control, good uniformity over large substrates, and with an excellent conformality on three-dimensional surface topologies. The most common ALD process is the deposition of Al\(_2\)O\(_3\) by alternating exposures of Al(CH\(_3\))\(_3\) and H\(_2\)O.\(^22\) In addition to H\(_2\)O, also other oxidation sources can be employed, such as O\(_3\) and an O\(_2\) plasma. When an O\(_2\) plasma is
used, the method is referred to as plasma-enhanced or plasma-assisted ALD, compared to thermal ALD when H2O or O3 are employed.

The fact that the ALD process allows the deposition of thin Al2O3 films with precise thickness control is illustrated in Fig. 1. In this figure, it is shown that the Al2O3 film thickness scales linear with the number of ALD cycles for both the thermal and the plasma-assisted ALD process. Both films were synthesized in the same ALD reactor and at the same substrate temperature of 200 °C. The slope of the film thickness as a function of the number of ALD cycles defines the so-called growth per cycle (GPC) of the ALD process. The GPC of 0.13 nm/cycle for the Al2O3 synthesized by plasma-assisted ALD is significantly higher than the GPC of 0.10 nm/cycle for the Al2O3 synthesized by thermal ALD. It is well known that the GPC of Al2O3 deposition can be different when different oxidation sources (e.g., H2O, O3, or O2 plasma) are used, even though the processes are all truly ALD with saturated half-reactions. The material properties of the Al2O3 films in Fig. 1 as synthesized by plasma-assisted and thermal ALD were studied in detail and reported in the literature. Briefly, the structural material properties of the Al2O3 films synthesized by thermal and plasma-assisted ALD are very similar for substrate temperatures ranging from 200 to 300 °C. The films are stoichiometric of composition ([O]/[Al]=1.5) and the hydrogen content [H] is typically ~2 and ~1 at. % for films synthesized at 200 and 300 °C, respectively. Figure 2 shows the thickness distribution of an Al2O3 film synthesized by plasma-assisted ALD on a 200 nm wafer. With spectroscopic ellipsometry, the thickness was sampled at 300 points. The thickness decreases from 79.8 nm in the middle of the wafer to 78.6 nm at 1 cm from the edge of the wafer indicating a good thickness uniformity. In another study it was demonstrated that the plasma-assisted ALD process of Al2O3 yields an excellent conformality on high aspect ratio structures with an aspect ratio of ~8.

III. EXPERIMENTAL DETAILS

A. ALD reactors

The experiments reported in this paper were conducted in a laboratory-scale open-load reactor (ALD-I, home built) and in a commercially available reactor (FlexAL™, Oxford Instruments Plasma Technology) with a loadlock system. Both systems are equipped with a remote inductively coupled plasma source located above the wafer for plasma-assisted ALD. The ALD-I and FlexAL reactor are capable of handling wafers up to 100 and 200 mm, respectively. The FlexAL reactor is equipped with multiple precursor pods and it can be used both for plasma-assisted and thermal ALD processes. Both reactors have optical viewports that enable in situ monitoring of the film growth by, e.g., spectroscopic ellipsometry. Other in situ diagnostic techniques have been fitted as well.

In the current study, the Al2O3 films were deposited at 200 °C unless indicated otherwise. The precursor doses and exposure times were chosen such that all films were deposited under saturated self-limiting conditions. A typical plasma-assisted ALD cycle at the FlexAL reactor consisted of a 20 ms injection of Al(CH3)3 vapor followed by a 1.5 s purging with O2. Subsequently a 400 W O2 plasma was ignited for 2 s followed by a 0.5 s purging with O2. Consequently, the ALD cycle time for plasma-assisted ALD on the FlexAL reactor was typically 4 s. A thermal ALD cycle at the FlexAL reactor consisted of a 20 ms injection of Al(CH3)3 vapor followed by a 3 s Ar purge. The oxidation step consisted of a 240 ms injection of H2O vapor followed by a 12 s purge with Ar resulting in an ALD cycle time of 16 s. The ALD process in the ALD-I reactor was not optimized with respect to cycle time and it consisted of five consecutive Al(CH3)3 dosings, 2 s Ar purging, and 2 s 100 W O2 plasma exposure. This resulted in a total ALD cycle time of 22 s.

For the surface passivation experiments, H-terminated c-Si wafers (see Sec. III C) were loaded in the reactor and preheated at the deposition temperature for 5 min before the start of the ALD process. Typically 50–255 ALD cycles were carried out resulting in Al2O3 films with a thickness of 7–30 nm as was monitored by in situ spectroscopic ellipsometry. For the lifetime measurements the c-Si wafers were coated at both sides. To do so, the wafer was turned over after the ALD of the Al2O3 film on the front side. Subsequently the sample was reloaded in the reactor without additional cleaning steps and an identical Al2O3 film was deposited at the rear side of the wafer. This resulted in symmetrically coated
For the depositions in the FlexAL reactor, it was found that a c-Si carrier wafer was necessary to protect the uncoated rear side of the c-Si wafer during the deposition of the Al₂O₃ film on the front side (see also Sec. III B). For the ALD-I reactor this was not necessary. The difference can most likely be attributed to differences in the design and materials used for the substrate stages in the two reactors.

**B. In situ spectroscopic ellipsometry**

The Al₂O₃ ALD processes were monitored by in situ spectroscopic ellipsometry measurements in the 1.24–6.50 eV photon energy range (J.A. Woollam M2000-D ellipsometer). Measurements were taken both prior to and during the Al₂O₃ deposition process. Spectroscopic ellipsometry measures the change in polarization of the reflected light expressed in the two ellipsometric angles Ψ and Δ. From the ellipsometric angles a pseudodielectric function \( \langle \varepsilon \rangle = \varepsilon_1 + i\varepsilon_2 \) can be calculated for the sample by assuming that \( \Psi \) and Δ arise from the reflection on a planar interface between air and a semi-infinite material with a pseudodielectric function \( \varepsilon_0 \). The imaginary part \( \varepsilon_2 \) of a c-Si substrate is extremely sensitive to the presence of native oxide or surface roughness at the c-Si surface, especially at the c-Si bulk critical point energy of 4.2 eV. Therefore, the \( \varepsilon_2 \) of c-Si was used to evaluate the condition of the c-Si surfaces (both for the front and rear sides of the wafer) prior to the deposition of the Al₂O₃ film.

Figure 3 shows \( \varepsilon_2 \) for a c-Si wafer as measured at both surfaces prior to the deposition of Al₂O₃ film. For the front side of the wafer, the \( \varepsilon_2 \) value of 47 at the \( E_2 \) bulk critical point energy indicates a clean H-terminated c-Si surface. For the rear side of the wafer, the \( \varepsilon_2 \) value at the \( E_2 \) bulk critical point energy is only 33. This significantly lower \( \varepsilon_2 \) value can be attributed to the presence of a native oxide on the rear side of the wafer, and from model calculations the presence of a 2.2 nm native oxide film on the c-Si could be derived. It was experimentally found that the native oxide formed at the rear side of the c-Si wafer during ALD on the front side of the wafer had a detrimental effect on the level of surface passivation obtained. The formation of such a native oxide on the rear surface during ALD on the front side of the wafer was only observed for the FlexAL reactor and could be prevented by the application of a carrier wafer. Therefore, a carrier wafer was used for all surface passivation experiments carried out in the FlexAL reactor, as mentioned in Sec. III A. With spectroscopic ellipsometry it was verified that all the results reported in this paper were obtained on “clean” H-terminated front and rear surfaces. The preparation of these surfaces is described in Sec. III C.

The ellipsometry spectra of the wafers as obtained during the Al₂O₃ deposition process were analyzed by a two layer optical model. This model consisted of a substrate with its optical properties defined by the measurement prior to deposition, and an Al₂O₃ film with the optical properties described by a Cauchy model. The typical refractive index for the Al₂O₃ films was in the range of 1.60–1.65 at a photon energy of 2 eV and the Al₂O₃ films did not show any absorption in the photon energy range investigated.

**C. c-Si surface passivation**

The surface passivation induced by the Al₂O₃ films was tested on double-side polished FZ p-type and n-type c-Si substrates of various doping concentrations. The c-Si substrates were cleaned using a conventional RCA1 and RCA2 clean and received a final treatment in diluted HF (1%) prior to deposition. This treatment results in a H-terminated c-Si surface which remains stable in open air for typically 15 min. After being symmetrically coated with Al₂O₃, the c-Si wafers were annealed at 425 °C for 30 min in a N₂ atmosphere. As reported in previous publications, this post-deposition annealing is found to be essential for obtaining the high levels of surface passivation reported.

The injection level dependent effective lifetime of the charge carriers in the symmetrically passivated c-Si wafer was determined by two different techniques. In the first technique, the injection level dependence of the effective lifetime was determined by photoconductance (Sinton Consulting WCT-100 lifetime tester). The effective lifetime of a passivated c-Si wafer was determined over a broad injection level range by measuring the excess conductance (photoconductance) by a contactless inductively coupled coil during a short \( \sim 10 \mu s \), photoconductance decay (PCD) or long \( \sim 2 \) ms, quasi-steady-state photoconductance (QSSPC) light flash. As all charge carriers contribute equally to the excess conductance of the c-Si wafer, the photoconductance technique is sensitive to carrier trapping or depletion region modulation. Both carrier trapping and depletion region modulations, if not correctly accounted for, lead to a strong increase in the apparent effective lifetime for low injection levels (typically \( 10^{13–15} \) cm⁻³). As a consequence, the level of surface passivation cannot be determined from photoconductance at low injection levels for wafers with a high trap density and/or a depletion region at the c-Si surface.

To assess the level of surface passivation at lower injection levels, a second technique, quasi-steady-state photoluminescence, was employed to determine the injection level...
dependent effective lifetimes. With this technique the injection level dependent effective lifetime is determined from the steady-state radiative recombination of charge carriers under steady-state illumination. The effective lifetime is calculated from the photoluminescence intensity using the self-consistent calibration method proposed by Trupke et al. As the photoluminescence signal is proportional to the electron-hole product, the measurements are basically unaffected by carrier trapping and depletion region modulation. Consequently, the effective lifetime can be determined at lower injection levels than for the photoconductance technique. It has, for example, been shown by Trupke et al. that effective lifetimes could accurately be determined down to the injection levels of $10^9$ cm$^{-3}$. This study also demonstrated that photoluminescence and photoconductance show a good agreement in the injection level range studied by the photoconductance measurements.

The uniformity of surface passivation was determined by the infrared lifetime mapping method. The effective lifetime of the charge carriers was determined from the infrared emission by the free carriers during steady-state illumination. The infrared emission is proportional to the total carrier concentration and, hence, is sensitive to carrier trapping and depletion region modulation. As a matter of fact, the sensitivity for carrier trapping is exploited when the infrared lifetime mapping technique is used for the mapping of the trap density of (multi)crystalline Si wafers.

All three techniques measure the effective lifetime $\tau_{\text{eff}}$ for symmetrically passivated c-Si wafers from which the level of surface passivation can be obtained. As will be demonstrated in Sec. IV, the values of $\tau_{\text{eff}}$ derived from the three techniques show good agreement in all cases reported apart from the results obtained for nearly intrinsic c-Si wafers. As discussed in the Appendix, under these conditions the results obtained by the photoconductance technique suffer from an artifact as recently also reported by McIntosh et al.

The effective surface recombination velocity $S_{\text{eff}}$ can be calculated from $\tau_{\text{eff}}$ by the expression

$$\frac{1}{\tau_{\text{eff}}} = \frac{1}{\tau_{\text{bulk}}} + \frac{2 S_{\text{eff}}}{d},$$

with $\tau_{\text{bulk}}$ being the c-Si bulk lifetime and $d$ the wafer thickness. As high purity FZ c-Si with a moderate doping level was used in this study, no empirically determined upper level for the intrinsic c-Si bulk lifetime (e.g., due to the Auger recombination) was assumed throughout this paper. Hence the bulk lifetime was set to infinity and the values of the effective surface recombination velocity $S_{\text{eff,max}}$ reported in this work should be considered as upper limits.

### D. Capacitance voltage and corona charging experiments

Capacitance-voltage (C-V) measurements on metal-oxide-semiconductor (MOS) structures were carried out on selected samples. The Al$_2$O$_3$ films, deposited on HF-dipped 20–30 $\Omega$ cm p-type c-Si wafers, were annealed for 30 min at 425 °C in $N_2$. Subsequently Al dots with an area of $\sim 1$ mm$^2$ were sputtered on the Al$_2$O$_3$ film. The C-V curves were measured at 10 kHz with an HP4275A multifrequency LCR meter.

Corona charging experiments were conducted to study the impact of the negative fixed charge density in the Al$_2$O$_3$ films on the level of c-Si surface passivation. By applying a high electric field at a tungsten needle, air particles and molecules were ionized in a corona chamber and positive ions were created. The positive charges, mainly consisting of (H$_2$O)$_n$H$^+$ ions, were deposited in small amounts on one side of a symmetrically passivated c-Si wafer. Subsequently the surface potential was determined by a vibrating Kelvin probe and the positive surface charge density was calculated using the dielectric constant derived from the C-V measurements and the film thickness obtained by spectroscopic ellipsometry. Simultaneously the effective lifetime of the passivated c-Si wafer was determined by the infrared lifetime mapping. This procedure was repeated multiple times for different positive surface charge densities up to a maximum positive surface charge density of $2 \times 10^{13}$ cm$^{-2}$.

### IV. RESULTS AND DISCUSSION

#### A. n-type c-Si passivated by plasma-assisted ALD Al$_2$O$_3$

The level of surface passivation was tested on n-type c-Si wafers (1.9 $\Omega$ cm (100)) with a thickness of 275 $\mu$m. In Fig. 4, the effective lifetimes are shown for the c-Si wafer.
fers passivated by Al$_2$O$_3$ synthesized on the ALD-I and FlexAL reactor. The results obtained on the ALD-I reactor were reported previously. From Fig. 4, it is clear that lifetimes in excess of 3 ms were obtained on both reactors. The 7 nm and 13 nm thick Al$_2$O$_3$ films synthesized on the ALD-I reactor had a better level of surface passivation than the 10 and 20 nm thick Al$_2$O$_3$ films synthesized on the FlexAL reactor. For the Al$_2$O$_3$ films synthesized on the FlexAL reactor, the level of surface passivation clearly increases for increasing film thickness up to 30 nm which is not uncommon for thin films used for surface passivation on c-Si. When a-Si$_x$H, a-Si$_x$N$_y$H, and a-SiC$_z$H are used to passivate c-Si, it is generally found that the level of surface passivation increases with film thickness and typically only saturates for films with a thickness >20 nm.\textsuperscript{44–46} The small difference between the results obtained on the ALD-I and FlexAL reactor can most probably be related to a small difference in the fixed charge density in the Al$_2$O$_3$ films provides field-effect compensation for the negative fixed charge density in thermal SiO$_2$ and c-Si surface. This effect was experimentally demonstrated by depositing positive corona charges on the Al$_2$O$_3$ surface to compensate the negative fixed charge density in the Al$_2$O$_3$ film. A similar approach using negative corona charges was used by Glunz et al. and Dauwe et al. to study the positive fixed charge density in thermal SiO$_2$ and a-Si$_x$N$_y$H films.\textsuperscript{41,42} The effective lifetime as a function of the positive-charge density $Q_{\text{corona}}$ deposited at the surface of the Al$_2$O$_3$ film in a corona chamber is shown in Fig. 5(a). Apart from a steady decline, which is most likely caused by too extensive handling of the c-Si wafer, the effective lifetime of the c-Si wafer is initially hardly affected by the positive-charge density deposited at the surface. The effective lifetime remains $>3$ ms up to a positive $Q_{\text{corona}}$ of $1.0 \times 10^{13}$ cm$^{-2}$. However, for a $Q_{\text{corona}}$ between (1.2–1.4) $\times 10^{13}$ cm$^{-2}$ a strong decrease in the effective lifetime is observed. A minimal effective lifetime of $\sim 100$ $\mu$s is obtained for a positive $Q_{\text{corona}}$ of $1.3 \times 10^{13}$ cm$^{-2}$. Consequently, the recombination rate at the c-Si surface is maximized for a positive $Q_{\text{corona}}$ of 1.3 $\times 10^{13}$ cm$^{-2}$. When the $Q_{\text{corona}}$ at the surface is further increased to $2 \times 10^{13}$ cm$^{-2}$, the effective lifetime increases again to values close to 2 ms.

In Fig. 5(b), the recombination rate at the surface of a 1.9 $\Omega$ cm n-type c-Si is simulated as a function of the total charge density present at the c-Si surface without taking additional charge carrier creation by external illumination into account. These simulations were performed in the software package PCDL that solves the coupled nonlinear electron and hole transport equations in crystalline semiconductors numerically.\textsuperscript{47} For simplicity we have assumed identical capture cross sections for electrons and holes at the c-Si surface. Consequently, for a $Q_{\text{corona}} < 1.3 \times 10^{13}$ cm$^{-2}$ the sum of $Q_f$ and $Q_{\text{corona}}$ is negative and for $Q_{\text{corona}} > 1.3 \times 10^{13}$ cm$^{-2}$ the sum is positive. The simulations were conducted with the software package PC1D (Ref. 47).
rate at the c-Si surface occurs when the sum of the $Q_f$ and $Q_{\text{corona}}$ is approximately zero. Hence, the minimum effective lifetime in Fig. 5(a) indicates the point at which the positive $Q_{\text{corona}}$ is balancing the negative $Q_f$ in the Al$_2$O$_3$ film. Consequently, we can estimate the negative fixed charge density present in this 26 nm Al$_2$O$_3$ film at $1.3 \times 10^{13}$ cm$^{-2}$.

From the simulation results in Fig. 5(b), we can also obtain some more insights into the field-effect passivation of c-Si. If the fixed charge density is sufficiently large (e.g., $Q_f > 5 \times 10^{11}$ cm$^{-2}$), the recombination rate approximately scales with $1/Q_f^2$ on lightly doped c-Si surfaces, irrespective of the polarity of the fixed charge density. Hence, the fixed charge density $>10^{13}$ cm$^{-2}$ measured for the 26 nm Al$_2$O$_3$ film implies that the field-effect passivation by this Al$_2$O$_3$ film is two orders of magnitude stronger than for a typical a-SiN$_x$:H film ($Q_f \approx 10^{12}$ cm$^{-2}$) and up to four orders of magnitude stronger than a typical thermal SiO$_2$ film ($Q_f < 10^{11}$ cm$^{-2}$). Moreover, Fig. 5(b) also clearly demonstrates that a high level of field-effect passivation can be obtained by the shielding of either electrons or holes from the c-Si surface. For the 1.9 $\Omega$-type c-Si wafer in Fig. 5, the level of field-effect passivation is independent of the polarity of the fixed charge density if $|Q_f| > 5 \times 10^{11}$ cm$^{-2}$. It is expected that a similar effect holds for p-type c-Si and hence, a good level of field-effect passivation can be obtained on lightly doped n- or p-type c-Si by either dielectrics with a positive or negative $Q_f$ if the fixed charge density is sufficiently high. This explains why dielectrics with a high fixed charge density such as a-SiN$_x$:H and Al$_2$O$_3$ can demonstrate a high level of surface passivation on both lightly doped n- and p-type c-Si.

The polarity of the fixed charge density is, however, of influence when considering the injection level dependence of the effective lifetime curves. The negative polarity of the fixed charge density in Al$_2$O$_3$ can provide an explanation of the injection level dependence of the effective lifetime for the passivated n-type c-Si wafers, as shown in Fig. 4. The effective lifetime peaks at an injection level within the range of $(1–3) \times 10^{15}$ cm$^{-3}$. At high injection levels $(>5 \times 10^{15}$ cm$^{-3}$), the effective lifetime decreases because the Auger recombination in the bulk of the c-Si wafer dominates. At lower injection levels the effective lifetime also decreases and this decrease is even more pronounced for the samples deposited in the FlexAL [Fig. 4(b)] reactor than for the samples deposited in the ALD-I reactor [Fig. 4(a)]. A similar injection level dependence of the effective lifetime was reported for p-type c-Si passivated by dielectrics with a positive fixed charge density such as thermal SiO$_2$ and a-SiN$_x$:H. In this case, the injection level dependence was primarily attributed to bulk recombination in the depletion region that is formed by the shielding of holes from the c-Si surface. Consequently, the decrease of the effective lifetime at low injection levels for n-type c-Si wafers passivated by Al$_2$O$_3$ can most probably be attributed to bulk recombination losses in the depletion region near the c-Si surface induced by the negative fixed charge density in the Al$_2$O$_3$ film.

In addition to the lightly doped n-type c-Si wafers (1.9 $\Omega$ cm), nearly intrinsic n-type c-Si (2 k$\Omega$ cm, (111)) and 20 k$\Omega$ cm, (100)) wafers were passivated by 30 nm thick Al$_2$O$_3$ films synthesized by plasma-assisted ALD in the FlexAL reactor. The effective lifetime of these passivated high-resistivity wafers was measured by both the photoconductance and photoluminescence technique. Significant differences were observed for the effective lifetimes extracted from the photoconductance technique versus the photoluminescence technique. As discussed in the Appendix, these differences can most likely be attributed to the nonlinear calibration of the Sinton lifetimeter for high-resistivity c-Si wafers, as shown by McIntosch et al. Consequently, only effective lifetime curves extracted from photoluminescence are considered for passivated high-resistivity wafers in this section.

The injection level dependent lifetime determined from quasi-steady-state photoluminescence technique is shown in Fig. 6. This figure reveals that lifetimes in excess of 18 ms can be obtained for c-Si wafers passivated by Al$_2$O$_3$. These lifetimes are among the highest measured effective lifetimes for passivated c-Si wafers reported in the literature. Different from the results obtained on lightly-doped n-type c-Si, the injection level dependence of the effective lifetime for these high-resistivity c-Si wafers is flat at low injection level, which is most probably related to the fact that these c-Si wafers are only lightly doped with a P density in the range of $10^{11}–10^{12}$ cm$^{-3}$. Hence, an inversion layer at the c-Si surface can only form if the excess carrier density is lower than the bulk doping level of $10^{11}–10^{12}$ cm$^{-3}$, which is below the detection limit of the photoluminescence technique for these high-resistivity c-Si wafers. The effective lifetimes obtained for these nearly intrinsic c-Si wafers unambiguously demonstrate that $\text{eff}_{\text{max}}$ values below 1 cm/s can be obtained on c-Si passivated by ALD-grown Al$_2$O$_3$ films.

**B. p-type c-Si passivated by plasma-assisted ALD-grown Al$_2$O$_3$**

In Fig. 7, the results obtained on p-type c-Si (2.0 $\Omega$ cm(111)) wafers with a thickness of 300 $\mu$m are shown. Effective lifetimes in excess of 1 ms were obtained for films deposited on the ALD-I and on the FlexAL system. The results obtained on the ALD-I reactor, which were already reported previously, reveal that the maximum
effective lifetime of the $c$-Si wafers passivated by 15 and 30 nm $Al_2O_3$ films is almost equal, i.e., 1.2 ms. However, the injection level dependence of the effective lifetime is completely different for these $c$-Si wafers. This difference will be addressed in more detail below. For $Al_2O_3$ films deposited at the FlexAL, effective lifetimes of 1.2 and 2.5 ms were obtained for $c$-Si wafers passivated by 10 and 30 nm thick $Al_2O_3$ films, respectively. The latter value exceeds the results obtained on the ALD-I reactor and corresponds to a $S_{eff,max}$ of 6 cm/s. Once again it is observed that the level of surface passivation provided by a 30 nm $Al_2O_3$ film is roughly a factor of 2 higher than the surface passivation provided by the 10 nm $Al_2O_3$.

The uniformity of the surface passivation was determined by means of infrared lifetime mapping on 100 mm $c$-Si wafers. The results are shown in Fig. 8 for samples deposited at both the ALD-I and the FlexAL reactor. As expected the measured effective lifetimes are in good agreement with the values obtained from photoconductance. The uniformity of the surface passivation of the sample deposited on the ALD-I reactor was rather poor whereas the uniformity for the sample deposited on the FlexAL reactor was good. The effective lifetime of the 100 mm $c$-Si wafer deposited in the FlexAL reactor is $>2$ ms for the largest part of the wafer and it only decreases slightly at the edges.

In addition to the photoconductance measurements, the injection level dependence of the effective lifetime was also determined from quasi-steady-state photoluminescence measurements for the $c$-Si wafers passivated by $Al_2O_3$ films deposited on the FlexAL reactor. From Fig. 7(b), it can be observed that there is an excellent agreement between the effective lifetime determined by both techniques down to an injection level of $1 \times 10^{13}$ cm$^{-3}$. This corroborates that the lifetime measurements by photoconductance on these $p$-type wafers were not affected by artifacts such as carrier trapping or depletion region modulation. The injection level dependence can therefore be related to the surface passivation mechanism.\(^{35,50}\) The effective lifetime of the passivated $p$-type wafers in Fig. 7 is dominated by Auger recombination in the bulk of the $c$-Si wafer for injection levels $>5 \times 10^{13}$ cm$^{-3}$.\(^{49}\) For lower injection levels the effective lifetime is roughly flat for the $p$-type $c$-Si wafers passivated by 10 and 30 nm thick $Al_2O_3$ deposited on the FlexAL reactor and a 10 nm thick $Al_2O_3$ deposited on the ALD-I reactor. The $p$-type $c$-Si wafer passivated by the 30 nm thick film deposited on the ALD-I reactor, however, shows a decrease of the effective lifetime for lower injection levels. A similar injection level dependence of the effective lifetime (i.e., a decrease at lower injection levels) was reported by Agostinelli et al. for 2 $\Omega$ cm $p$-type $c$-Si passivated by $Al_2O_3$ synthesized by thermal ALD.\(^{28}\) As shown in Fig. 8(a), the uniformity of the effective lifetime of the $c$-Si wafer passivated by a 30 nm thick $Al_2O_3$ film deposited on the ALD-I reactor was rather poor. We argue therefore that this nonuniformity of the effective lifetime over the $c$-Si wafer likely results in an apparent injection level dependence of the effective lifetime as determined by the photoconductance technique due to its rather large detection area of $\sim 6$ cm$^2$.\(^{33}\) It should be noted that all of the $p$-type $c$-Si wafers passivated by $Al_2O_3$ films synthesized by plasma-assisted ALD on the FlexAL reactor demonstrated a good uniformity for the level of sur-
face passivation as well as a fairly constant effective lifetime at low injection levels.

The flat injection level dependence of the effective lifetime at low injection levels is consistent with the extended Shockley–Read–Hall model proposed by Girish et al. because for p-type silicon the bulk minority charge carriers, the electrons, are effectively shielded from the c-Si surface by the negative fixed charge density present in the Al2O3 film. For dielectrics containing a positive fixed charge density (e.g., thermal SiO2 and a-SiNx:H), the injection level dependence of the effective lifetime is flat for low injection levels on n-type c-Si when the bulk minority charge carrier, the holes, are shielded from the c-Si surface. On the other hand, p-type c-Si wafers passivated by dielectrics with a positive fixed charge density demonstrate a strong reduction of the effective lifetime for decreasing injection levels. This pronounced injection level dependence can primarily be attributed to the fact that in this case the bulk majority charge carriers, the holes, are shielded from the c-Si surface which results in an inversion layer at the c-Si surface. Consequently, from the results presented in the previous section and in this section, it can be concluded that the injection level dependence of the effective lifetime of c-Si passivated by the negative-charge dielectric of Al2O3 basically “mirrors” the injection level dependence for c-Si wafers passivated by the positive-charge dielectrics of thermal SiO2 and a-SiNx:H. The effective lifetime can thus be made fairly constant at low injection levels for both n-type and p-type c-Si wafers by choosing a surface passivation dielectric with fixed charges whose polarity lead to shielding of bulk minority carriers from the c-Si surface. It should be realized that the injection level dependence of the effective lifetime is not only of academic interest. A low level of surface passivation at low injection levels is detrimental for low-light performance of c-Si solar cells and will therefore have a negative effect on their yearly kW h output per watt peak, as shown by Aberle et al.

Another aspect of high relevance is that when the positive-charge dielectric a-SiNx:H is applied at the rear of p-type c-Si solar cells, the positive fixed charge density causes a parasitic shunting effect which significantly lowers the c-Si solar cell efficiency. This parasitic shunting effect is not observed when the negative-charge dielectric Al2O3 is applied at the rear of p-type c-Si solar cells. The excellent level of surface passivation of lightly doped p-type c-Si by Al2O3 was recently confirmed by passivated emitter and rear cell solar cells in which Al2O3 yielded an equal performance as annealed thermal SiO2. An independently confirmed solar cell efficiency of 20.6% was obtained for these solar cells.

In Fig. 9, the level of surface passivation obtained by Al2O3 films synthesized by plasma-assisted ALD on p-type c-Si is compared with the best results published to date for thermally grown SiO2, as-deposited a-SiNx:H, and a-Si:H [the latter two films were synthesized by plasma-enhanced chemical vapor deposition (PECVD)]. The surface recombination velocity for lightly B-doped c-Si surfaces was calculated using the highest effective lifetimes reported in the literature assuming an infinite bulk lifetime (see Sec. III A). The surface recombination velocities at the heavily B-doped c-Si surfaces were extracted from simulations by the device simulation package SENSURAUS, as described by Altermatt et al. and Hoex et al. From Fig. 9, it is clear that Al2O3 provides a state-of-the-art level of surface passivation on p-type c-Si with an arbitrary doping level. The high negative fixed charge density effectively shields the bulk minority carriers (i.e., the electrons) from the p-type c-Si surfaces. Evidently, the level of surface passivation obtained by Al2O3 on low resistivity p-type c-Si is comparable to the best results reported for thermal SiO2, a-SiNx:H, and a-Si:H whereas heavily B-doped c-Si surfaces are even more effectively passivated by Al2O3. The excellent level of surface passivation of highly doped p-type c-Si by Al2O3 was recently confirmed by passivated emitter with rear locally diffused (PERL) solar cells in collaboration with the Fraunhofer ISE Institute. These n-type PERL solar cells in which Al2O3 was applied to passivate the p-type emitter yielded an independently confirmed solar cell efficiency of 23.2%.

C. Lightly-doped p-type c-Si passivated by thermal ALD Al2O3

The results presented in this work so far were obtained for Al2O3 synthesized by plasma-assisted ALD. The study of Agostinelli et al. demonstrated that also Al2O3 films synthesized by thermal ALD with H2O as oxidizing agent can provide a very high level of surface passivation on p-type c-Si. Therefore, a brief experiment was carried out to directly investigate a possible influence of the oxidizing agent selected, i.e., the level of surface passivation obtained by thermal and plasma-assisted ALDs was compared by synthesizing the Al2O3 films by both processes in the same reactor.

The injection level dependent effective lifetime was determined for a p-type c-Si wafer (2 Ω cm, (111)) passivated by a 30 nm Al2O3 synthesized by thermal and plasma-assisted ALD in the FlexAL reactor. As can be observed in

![FIG. 9. (Color online) Upper level of the effective surface recombination velocity as a function of the B-density at the surface of c-Si wafers passivated by thermal SiO2, a-SiNx:H, a-Si:H, and Al2O3. The values for highly B-doped surfaces for thermal SiO2, a-Si:H, and Al2O3 were taken from a previously published study (Ref. 32). The effective surface recombination values for low B-densities were calculated from the lifetime values reported by Kerr et al. and Dauwe et al. assuming an infinite bulk lifetime (Ref. 1, 6, and 53). The a-Si:H and a-SiN:H films were grown by PECVD and the Al2O3 films were synthesized by plasma-assisted ALD. The SiO2 films were grown by thermal oxidation. The error bars are smaller than the symbol size unless indicated otherwise.](image-url)
min at 425 °C in N₂. It can clearly be seen that the flatband voltage, which is related to the fixed charge density in the \( \text{Al}_2\text{O}_3 \), is significantly different for the two films. The flatband voltage shift of the MOS structures corresponds to a negative fixed charge density of \( 1 \times 10^{12} \) and \( 7 \times 10^{12} \text{ cm}^{-2} \) for the thermal and plasma-assisted ALD \( \text{Al}_2\text{O}_3 \) films, respectively. From Fig. 5(b), it can be deduced that this difference in negative fixed charge density between the thermal and plasma-assisted ALD \( \text{Al}_2\text{O}_3 \) films results in a 50 times higher surface recombination rate for the thermal ALD \( \text{Al}_2\text{O}_3 \) film. Consequently, the difference between the level of surface passivation of the thermal and plasma-assisted ALD \( \text{Al}_2\text{O}_3 \) films can be attributed mainly to a difference in level of field-effect passivation. Cho et al. and Kim et al. also reported significant differences in the electrical properties of thermal and plasma-assisted ALD \( \text{Al}_2\text{O}_3 \). The fact that the negative fixed charge density in the \( \text{Al}_2\text{O}_3 \) film synthesized by plasma-assisted ALD on the FlexAL reactor is lower than for the 26 nm thick film grown on the ALD-I reactor also explains the difference in the level of surface passivation obtained for the two reactors for films <30 nm thick, as shown in Fig. 4.

From the electrical characterization it was found that the negative fixed charge density in plasma-assisted ALD \( \text{Al}_2\text{O}_3 \) films in the as-deposited state is significantly lower (up to one order of magnitude) than after the postdeposition annealing. It is therefore not expected that the \( \text{O}_2 \) plasma itself results in this high fixed charge density. Moreover, Agostinelli et al. reported negative fixed charge densities up to \( 1 \times 10^{13} \text{ cm}^{-2} \) in their \( \text{Al}_2\text{O}_3 \) films synthesized by thermal ALD, but without specifying the post-thermal treatment applied. It is therefore plausible that the optimal postdeposition annealing required to obtain the highest level of surface passivation is different for the thermal and plasma-assisted ALD synthesized \( \text{Al}_2\text{O}_3 \).

V. CONCLUSIONS

From the results reported in this paper, it is evident that \( \text{Al}_2\text{O}_3 \) films synthesized by plasma-assisted ALD are an interesting candidate for passivation of \( c\)-\( \text{Si} \) surfaces as required for the next-generation (high-efficiency) \( c\)-\( \text{Si} \) solar cells. The self-limiting nature of the ALD process results intrinsically in the precise thickness control and good thickness uniformity of the thin films. Moreover, the level of surface passivation provided by the low-temperature deposited \( \text{Al}_2\text{O}_3 \) films is at least equal to the state-of-the-art level of surface passivation obtained by high-temperature thermal \( \text{SiO}_2 \) films on lightly doped \( n\)-\( \text{Si} \) wafers and \( p\)-\( \text{Si} \) wafers with an arbitrary doping level. The level of surface passivation by \( \text{Al}_2\text{O}_3 \) was determined by the techniques based on photocurrent, infrared emission, and photoluminescence showing an excellent agreement for \( p\)-\( \text{Si} \) unless the \( c\)-\( \text{Si} \) wafer was nearly intrinsic. The excellent surface passivation by \( \text{Al}_2\text{O}_3 \) films can mainly be attributed to a high negative fixed charge density in the \( \text{Al}_2\text{O}_3 \), as directly demonstrated in this work. A negative fixed charge density is shown to be especially beneficial for the passivation of (highly doped) \( p\)-\( \text{Si} \) as the bulk minority charge car-
The effective lifetime was measured by the pronounced negative fixed charge density in the plasma-assisted ALD Al₂O₃ films. It was demonstrated that the effective lifetime would decrease for a decreasing injection level, which implies that this kind of artifacts cannot explain the anomalous behavior observed. To date no physical mechanism has been reported that can explain the subsequent decrease in the effective lifetime for injection levels lower than 2×10^{14} cm⁻³. However, by McIntosh et al. it was shown that the photocurrent could be induced by the Sinton WCT-100 (Ref. 33) lifetime tester, which is routinely used within the photovoltaics community, shows a strong nonlinearity for a conductance lower than 1 mS (i.e., corresponding to a 300 μm c-Si wafer with a resistivity of >30 Ω cm). Although the physical origin of this nonlinearity is still under investigation, this nonlinear response can likely explain the particular values of the effective lifetime deduced from the photocurrent measurements for these high-resistivity n-type c-Si wafers. Accordingly, for the high-resistivity c-Si wafers employed in this work, it was decided to use the effective lifetimes extracted from the photoluminescence measurements with the self-consistent calibration proposed by Trupke et al. 37

FIG. 12. (Color online) Injection level dependent effective lifetime of a nearly intrinsic n-type c-Si wafer (2 kΩ cm, (111), 300 μm) symmetrically passivated with a 30 nm Al₂O₃ film. The effective lifetime was measured by both in PCD and QSSPC modes and quasi-steady-state photoluminescence.

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APPENDIX: EFFECTIVE LIFETIME FOR NEARLY INTRINSIC C-SI AS DETERMINED BY THE PHOTOCURRENT AND PHOTOLUMINESCENCE TECHNIQUES

Figure 12 shows the injection level dependence of the effective lifetime as determined from the photocurrent and photoluminescence techniques for nearly intrinsic n-type c-Si (2 kΩ cm, (111)) wafer symmetrically passivated by a 30 nm Al₂O₃ film synthesized by plasma-assisted ALD. It can be seen that a strong discrepancy exists between the two results. The photoluminescence measurement shows a constant effective lifetime of 18 ms at low injection levels and a decrease at higher injection levels due to the Auger recombination. The photocurrent data show a pronounced behavior with an extremely high effective lifetime of 72 ms at an injection level of (2–4)×10^{14} cm⁻³. A similar, but less pronounced, anomalous behavior of the effective lifetime was deduced from photocurrent measurements for the 90 Ω cm n-type c-Si passivated with forming gas annealed thermal SiO₂ by Kerr et al. 4 If the photocurrent measurement in Fig. 12 would be affected by depletion region modulation or carrier trapping, the effective lifetime would demonstrate an increase for a decreasing injection level, 34,35 which implies that this kind of artifacts cannot explain the anomalous behavior observed. To date no physical mechanism has been reported that can explain the subsequent decrease in the effective lifetime for injection levels lower than 2×10^{14} cm⁻³. However, by McIntosh et al. it was shown that the photocurrent could be induced by the Sinton WCT-100 (Ref. 33) lifetime tester, which is routinely used within the photovoltaics community, shows a strong nonlinearity for a conductance lower than 1 mS (i.e., corresponding to a 300 μm c-Si wafer with a resistivity of >30 Ω cm). Although the physical origin of this nonlinearity is still under investigation, this nonlinear response can likely explain the particular values of the effective lifetime deduced from the photocurrent measurements for these high-resistivity n-type c-Si wafers. Accordingly, for the high-resistivity c-Si wafers employed in this work, it was decided to use the effective lifetimes extracted from the photoluminescence measurements with the self-consistent calibration proposed by Trupke et al. 37


