Surface Passivation of Boron-Diffused p-Type Silicon Surfaces With (1 0 0) and (1 1 1) Orientations by ALD Al_2O_3 Layers

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Abstract—Boron-diffused p+/n/p+ and undiffused silicon samples with (1 0 0) and (1 1 1) orientations passivated by aluminum oxide (Al_2O_3) that is synthesized by atomic layer deposition (ALD) have been investigated. Emitter saturation current densities of ~ 24 , 29, and 33 fA/cm^2 were obtained for (1 0 0) samples with symmetrical $85\Omega/\square$ B diffusions that were passivated by plasma-assisted, H_2O -based, and O_3 -based ALD Al_2O_3 , respectively. Compared with undiffused samples, it was found that the additional surface doping from the diffusion reduces recombination at the $\text{Al}_2\text{O}_3/\text{Si}$ interface in the case of relatively low surface boron concentrations ($< 2 \times 10^{19} \text{ cm}^{-3}$). The degree of surface passivation that is observed on (1 0 0) surfaces was generally better than on (1 1 1) surfaces, particularly for undiffused samples, but this difference effectively disappeared following the application of more negative charge by corona charging. From capacitance–voltage measurements, it was found that Al_2O_3 films on substrates with a (1 0 0) orientation display a higher negative fixed charge density Q_f than films on (1 1 1) samples. On the other hand, the interface state density D_{it} was not strongly influenced by surface orientation of the substrate. It appears that the difference in negative charge density is at least partly responsible for the differences in the observed passivation.

Index Terms—Atomic layer deposition (ALD) Al_2O_3 , boron emitter, passivation, solar cells, (1 0 0) and (1 1 1).

I. INTRODUCTION

THE imperative for continued cost reductions for crystalline silicon solar cells and systems requires continuous improvements in device efficiencies, as well as reductions in wafer thickness. However, the overall efficiency of the thinner solar cells is limited by surface recombination [1]. P-type silicon substrates are typically used in the silicon photovoltaic industry in preference over n-type substrates. One reason for this choice is the lack of high-quality passivation for p-type emitters. However, n-type solar cells show interesting benefits when compared with p-type, such as a higher tolerance to impurities present in

low-grade silicon (particularly Fe) [2] and not suffering from oxygen–boron–complex degradation [3].

Al_2O_3 has been demonstrated to be a promising material to passivate both undiffused and diffused p-type c-Si, as it features both a relatively high fixed negative charge density, as well as a low interface defect density. The high negative charge density means that the minority carrier electron concentration is effectively reduced at the surface, and consequently, the recombination rate is reduced [4].

Most of published work about Al_2O_3 passivation is based on undiffused p-type c-Si [5], but the passivation performance and mechanism on boron-doped p-type surfaces also deserves attention. Two types of ALD processes have been developed for the deposition of Al_2O_3 : plasma-assisted ALD (PA-ALD), which utilizes an oxygen plasma as the oxidant, and thermal ALD. For thermal ALD, the most commonly used oxidants are H_2O and O_3 [6]. State-of-the-art passivation performance from PA-ALD Al_2O_3 films on highly doped B emitters has been reported [7], [8], which was verified by the demonstration of high-efficiency n-type solar cells that feature a front emitter passivated by PA-ALD Al_2O_3 [9]. To our knowledge, the passivation performance of H_2O and O_3 -based ALD Al_2O_3 films on B-diffused surfaces has not been investigated, even though these two methods are more suitable for an industrial inline ALD system [10].

Direct comparisons between emitter saturation current density J_{oe} values from undiffused and diffused samples are relatively rare in the literature; nevertheless, this comparison is useful for the direct evaluation of the influence on the surface recombination that is introduced by the presence of diffusion. For example, Jin *et al.* [11] concluded that the presence of a B diffusion results in an increase in recombination at the Si/SiO₂ interface of thermally oxidized samples through lifetime measurements on metal–oxide–silicon (MOS) structures at a sufficiently large applied bias voltage. Because of the high negative charge density at the ALD $\text{Al}_2\text{O}_3/\text{Si}$ interface, it is possible in most cases to obtain accurate J_{oe} values for undiffused samples.

It is also of significant interest to understand how differences in crystal orientation affect the degree of passivation achievable. Recently, Baker-Finch and McIntosh found that a (1 1 1) orientation drives a fourfold increase in surface recombination velocity (SRV) for SiO₂ passivation but a negligible change for SiN_x:H passivation compared with (1 0 0)-orientated surfaces [12]. For a-Si:H passivation, it is interesting that certain compositions of a-Si:H more successfully passivate (1 1 1) than (1 0 0) surfaces [13]. Since Al_2O_3 -passivated surfaces have been

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shown to feature a thin SiO_x layer at the interface [4], and given that the interface has been described as essentially similar to an Si-SiO₂ interface that is formed by low-temperature oxidation [14], a significant difference in passivation of (1 0 0) and (1 1 1) interfaces that are passivated by Al_2O_3 may also be expected. However, little data exist to date for Al_2O_3 films.

In this study, we investigated passivation of ALD- Al_2O_3 layers on undiffused (1 1 1) and (1 0 0) samples, and B-diffused (1 1 1)- and (1 0 0)-orientated samples, using different oxidants, in order to elucidate the impact of both the presence of a diffusion and the surface orientation on the resulting surface passivation. The films were synthesized by the use of the three commonly used oxidants in order to study the impact of the choice of oxidant on the film properties. This way, the influence of B doping, silicon crystal orientation, oxidant, and anneal time on surface passivation were studied in detail.

II. EXPERIMENTAL DETAILS

High resistivity $>100 \Omega\cdot\text{cm}$, n-type, (1 0 0), float zone (FZ) silicon wafers, and $>2000 \Omega\cdot\text{cm}$, n-type, FZ, (1 1 1)-orientated silicon wafers were used for diffusions. These samples were etched (removing at least $20 \mu\text{m}$ per side) in a HNO_3/HF solution to remove any saw damage. After receiving a standard RCA clean, selected samples were subjected to a boron diffusion at a temperature of 940°C using a liquid boron tribromide (BBr_3) source to form a symmetrical structure of p+/n/p+. The diffusion was terminated with a ramp down in oxygen ambient to grow a very thin layer of silicon oxide. After stripping the boron containing glass with a HF dip, relatively deep p-n junctions were obtained as a consequence of 1-h drive-in at 1000°C in nitrogen.

Before Al_2O_3 deposition, samples received a conventional RCA clean with a final dip in diluted HF (5%). PA-ALD, H_2O -ALD, and O_3 -ALD Al_2O_3 films were synthesized in a Beneq TFS-200 ALD system at deposition temperatures of 175°C , 200°C , and 200°C , respectively. All Al_2O_3 films had a thickness of $\sim 30 \text{ nm}$ unless stated otherwise, and the passivation performance was activated by annealing in forming gas (95% Ar_2 with 5% H_2) at 400°C . To monitor the impact of anneal time, J_{oe} measurements were conducted for cumulative anneal time up to 160 min.

Capacitance-voltage measurements (C-V) were conducted on Czochralski (Cz), n-type, 1–10 $\Omega\cdot\text{cm}$ samples, and both (1 0 0)- and (1 1 1)-orientated wafers were chosen. Prior to the C-V measurements, all insulator layers were removed from the rear of the samples, and $\sim 80\text{-nm}$ Al dots with an area of $\sim 0.004 \text{ cm}^2$ were evaporated on the front and In/Ga contacts were formed on the rear of samples.

The sheet resistance R_{sh} of the samples was determined by four-point probe measurements and the doping profiles were determined by electrochemical C-V (ECV) profiling using a CVP21 wafer profiler from the company WEP. The total recombination in the surface regions, that is, the saturation current density J_{oe} , was determined from contactless photoconductance decay measurements in both the generalized and transient modes (Sinton WCT-120) [15]. Under conditions where the wafer bulk

is in high injection and the surfaces are in low-level injection, the relationship between measured effective excess carrier lifetime τ_{eff} and J_{oe} is given by [16]

$$\frac{1}{\tau_{eff}} - \frac{1}{\tau_{Auger}} = \frac{1}{\tau_{SRH}} + 2 \cdot \frac{J_{oe} \cdot (N_d + \Delta n)}{q \cdot n_i^2 \cdot W} \quad (1)$$

where τ_{Auger} the intrinsic Auger lifetime, τ_{SRH} is the defect-related bulk lifetime, N_d is the base doping level, n_i is the intrinsic carrier concentration of c-Si, q is the elementary charge, Δn is the excess carrier density, and W is the sample thickness. The value of J_{oe} can be determined from the slope of $1/\tau_{eff}$ versus Δn once Auger recombination has been accounted for. For the latter, we use the Auger model by Kerr and Cuevas [17]. J_{oe} in this paper was extracted from the linear fit around an injection level of $8 \times 10^{15} \text{ cm}^{-3}$. For all measurements reported here, Δn is uniform throughout the wafer due to the high bulk lifetime and low surface recombination rates.

Equation (1) is valid for both diffused and undiffused samples provided the surface remains in low-level injection during the measurement [16]. For the diffused samples, the condition is satisfied for all passivation layers, as long as the fixed charge in the passivation layer is negative. For the undiffused samples that are used here, calculations indicate that a value of $-1 \times 10^{12} \text{ cm}^{-2}$ of Q_f results in a surface hole density $p_s = 3.8 \times 10^{18} \text{ cm}^{-3}$, which means that for films with a negative charge density of at least $-1 \times 10^{12} \text{ cm}^{-2}$ the extracted J_{oe} values should be valid. For anneal times of 15 min or longer, charge densities that were extracted from C-V measurements were at least $-1.2 \times 10^{12} \text{ cm}^{-2}$ for all the films that were investigated. For shorter anneal times, the validity of the extracted J_{oe} values cannot be guaranteed since samples with 5 min-anneal were not measured to extract charge density.

In this paper, we focus on J_{oe} values as the primary indicator of surface passivation, since this is the most appropriate and easily interpreted quantity for the surfaces that are investigated here. For PECVD SiN_x films, it is known that the effective lifetime of undiffused samples can decrease at low-injection levels, which has been attributed to the presence of a surface-damaged region (SDR) that is typically in the order of $1 \mu\text{m}$ in depth or less [18]. When such an SDR is present, the determination of J_{oe} alone does not completely characterize the sample, since the SDR can have a significant detrimental impact on cell characteristics near maximum power point, even for low emitter saturation current densities. However, recent measurements [19] have indicated that this effect is small for ALD Al_2O_3 -passivated samples and is unlikely to significantly impact on cell performance. In this paper, we, therefore, focus on measurement of J_{oe} .

Corona charging was carried out with a conventional setup by applying a voltage of $\pm 6 \text{ kV}$ to a steel needle about 10 cm above the samples.

III. RESULTS AND DISCUSSION

A. Impact of Anneal Time

In Fig. 1, the selected curves of Auger-corrected inverse effective lifetime as a function of excess carrier density for samples that are passivated by PA-ALD Al_2O_3 are shown. In general,

Fig. 1. Measured Auger-corrected inverse effective lifetime as a function of excess carrier density for both B-diffused and -undiffused c-Si samples with (1 0 0) or (1 1 1) orientation. The samples were passivated on both sides by ~ 30 -nm PA-ALD Al_2O_3 layers. Solid symbols represent data of samples after 5-min FGA; open symbols are for samples after 100-min FGA.

these curves show a reasonably linear relationship with injection level for both diffused and undiffused samples, even for the undiffused samples with the shortest anneal times (5 min), where, as noted earlier, the conditions to extract meaningful J_{oe} values may not have been strictly met. The curves can be used for estimation of J_{oe} , although the validity of the extracted values is not guaranteed and hence these particular values must be interpreted with great caution.

The extracted J_{oe} values of undiffused and B-diffused samples ($R_{sh} \sim 85 \Omega/\square$) are shown as a function of increasing forming gas anneal (FGA) time in Fig. 2(a)–(c) for PA-ALD-, H_2O -ALD, and O_3 -ALD Al_2O_3 -passivated surfaces, respectively. The level of surface passivation for all samples is dramatically improved with cumulative FGA time, which is indicated by $J_{oe} \sim 24, 29$, and 33 fA/cm^2 for PA-ALD, H_2O -ALD, and O_3 -ALD Al_2O_3 -passivated emitters after a 100-min FGA.

Postdeposition annealing has previously been shown to affect both the chemical and field-effect contribution of the ALD Al_2O_3 film to surface passivation. For PA-ALD and H_2O -ALD Al_2O_3 films, it has been reported that the fixed negative charge density Q_f increases substantially during annealing and induces effective field-effect passivation [6], [20]. Meanwhile, the interface state density D_{it} decreases during annealing, which has been ascribed to the effective transport of hydrogen from the Al_2O_3 bulk to the interface [21].

We note that different anneal times are needed for optimal surface passivation, with PA-ALD films requiring the longest and H_2O -ALD films the shortest amount of time. This is likely to be the result of a much higher defect density at PA-ALD $\text{Al}_2\text{O}_3/\text{Si}$ and O_3 -ALD $\text{Al}_2\text{O}_3/\text{Si}$ interfaces than at H_2O -ALD $\text{Al}_2\text{O}_3/\text{Si}$ interfaces in the as-deposited state [6]. Furthermore, we suppose that defects generated during PA-ALD and O_3 -ALD are different, with the former requiring longer anneal times for passivation. Profijt *et al.* have demonstrated experimentally that vacuum ultraviolet (VUV) radiation in the O_2 plasma during

PA-ALD deposition can introduce interfacial defects [22], but no process has been found to affect the interface properties of as-deposited films when synthesized by O_3 -ALD.

Fig. 2 illustrates that the passivation performance of Al_2O_3 films that are synthesized by the three deposition methods is relatively thermally stable at 400°C in forming gas ambient, especially for PA-ALD Al_2O_3 . This is in contrast with the results of Schmidt *et al.* [23], who observed the degradation of the PA-ALD-deposited Al_2O_3 layers after 15 min FGA at 425°C . The difference could be because of the slightly lower annealing temperature that is used in this study. We also note that Schmidt *et al.* used a slightly different anneal ambient (95% N_2 with 5% H_2) compared with 95% Ar_2 mixed with 5% H_2 used in this study. For B-diffused samples in Fig. 2, J_{oe} is dominated by Auger recombination in the diffused region. This is discussed in more detail in Section III-B.

It is noticeable that, for H_2O -ALD Al_2O_3 -passivated, undiffused samples, passivation saturation was observed after 5 min FGA, while around 20 min is required for diffused samples. It is possible that on the as-deposited H_2O -ALD $\text{Al}_2\text{O}_3/\text{Si}$ interface, additional defects are introduced by the presence of a B diffusion, and that at least some of these defects are annihilated by the subsequent anneal or are effectively passivated by hydrogen. For the other two passivation methods, apart from the surface B diffusion, a high density of defects is also introduced during deposition [6] and govern the annealing behavior of both diffused and undiffused samples. Therefore, a similar anneal behavior is observed for diffused and undiffused samples.

In Fig. 2, comparing the passivation performance of the films that are deposited using the three different oxidants on B-diffused and undiffused n-type samples, it can be seen that PA-ALD films display the highest level of passivation for boron emitters after a sufficiently long FGA at 400°C , while the highest J_{oe} values are from O_3 -ALD Al_2O_3 films, which is consistent with the results by Dingemans *et al.* on low resistivity p-type wafers that are passivated by ALD Al_2O_3 [6]. For the undiffused samples, films that are deposited using PA-ALD and H_2O -ALD display similar J_{oe} values, while those of the O_3 -ALD films are significantly higher. Despite these differences, in all cases very good surface passivation is obtained on B-diffused and -undiffused n-type samples.

B. Emitter Passivation

In Fig. 3, the impact of additional negative charge on the passivation of the layers of Fig. 2 is summarized. An increasing density of negative charge was deposited on both sides of each sample until saturation of J_{oe} was observed. J_{oe} before corona charging was measured from the samples in Fig. 2 after one month storage in air; therefore, a slight discrepancy of J_{oe} values between Figs. 2 and 3 can be observed for some samples. It is evident that additional negative charge improves the passivation performance for all the films that are investigated, which illustrate that the inherent negative charge in the films is not enough to optimally passivate the samples.

Fig. 3(a) shows that, following corona charging, the J_{oe} values of samples with B diffusion are all remarkably similar to

Fig. 2. Emitter saturation current density J_{oe} for different ALD- Al_2O_3 (PA-ALD, H_2O -ALD, and O_3 -ALD) passivated (1 0 0) and (1 1 1) silicon surfaces on B-diffused and -undiffused samples as a function of cumulative anneal time. (a) PA-ALD Al_2O_3 passivation, (b) H_2O -ALD Al_2O_3 passivation, and (c) O_3 -ALD Al_2O_3 passivation. The lines are guides to the eye.

Fig. 3. Improved passivation performance of PA-ALD, H_2O -ALD, and O_3 -ALD Al_2O_3 layers obtained by negative corona charging on both sample surfaces until no further improvement in J_{oe} was observed. The samples used for these measurements were those used for the measurements in Fig. 2. (a) B-diffused. (b) Undiffused.

each other. In contrast, there are significant differences in the J_{oe} values of the undiffused samples, which indicate that significant differences exist at the Al_2O_3 -Si interfaces of the films that are synthesized using different oxidants.

Compared with undiffused samples, the presence of a boron diffusion introduces an additional component to J_{oe} due to (Auger dominated) recombination in the diffused region. In addition, the increased surface doping will alter the J_{oe} contribution from the surface itself.

Fig. 4. C-V measurements of MOS structures containing ~ 30 -nm-thick Al_2O_3 films. The Al_2O_3 films were synthesized by PA-ALD, H_2O -ALD, and O_3 -ALD on single side polished p-type c-Si wafers and subsequently annealed for 30 min at 400 in forming gas. The lines are guides to the eye. (a) C-V curves. (b) Dit distribution.

Contributions to J_{oe} from the diffusions and the surface regions are summarized in Table I, using the data from Fig. 3. J_{oe} values of perfectly passivated emitters were determined from the experimental B doping profiles and the assumption of zero surface recombination ($\text{SRV} = 0$) using the emitter simulation spread sheet EDNA [24] that is based on Fermi-Dirac statistics [25].

In Table I, the J_{oe} values of the samples with B emitter following corona charging ($J_{oe \text{ diff cor}}$) are equal or nearly equal to the calculated J_{oe} contribution from the diffused region ($J_{oe \text{ ideal}}$), which indicates that the J_{oe} contributions from the surfaces themselves are very small in these samples. In other words, following corona charging, the surfaces of these samples—regardless of crystal orientation—are nearly perfectly passivated. Thus, negative charge is highly effective in improving interface passivation, even for diffused samples.

For the diffused samples, the J_{oe} contribution from the surface itself following corona charging ($J_{oe \text{ diff cor}} - J_{oe \text{ ideal}}$) is in the range 0–5 fA/cm^2 . In contrast, the J_{oe} for undiffused surfaces ($J_{oe \text{ undiff cor}}$) is in the range 4–17 fA/cm^2 following corona charging. Similarly, the surface J_{oe} contributions for B-diffused samples prior to the application of corona charge ($J_{oe \text{ diff}} - J_{oe \text{ ideal}}$) are generally, but not always, lower than the corresponding values for undiffused samples ($J_{oe \text{ undiff}}$).

TABLE I

EMITTER SATURATION CURRENT DENSITY J_{oe} FROM SAMPLES WITH AND WITHOUT BORON EMITTER AND SIMULATED IDEAL $J_{oe\ ideal}$ FOR SAMPLES WITH MEASURED BORON DOPING PROFILE AND ASSUMED SRV = 0

Related Saturation currents (fA/cm ²)	PA-ALD Al ₂ O ₃		H ₂ O-ALD Al ₂ O ₃		O ₃ -ALD Al ₂ O ₃	
	(111)	(100)	(111)	(100)	(111)	(100)
Simulated $J_{oe\ ideal}$ with ideal passivation SRV=0	22	22	22	22	22	22
Measured $J_{oe\ undiff}$ without emitter	8	6	12	9	28	16
Measured $J_{oe\ diff}$ with emitter	26	27	36	33	39	33
Contribution from surface $J_{oe\ diff} - J_{oe\ ideal}$	4	5	14	11	17	11
Measured $J_{oe\ undiff\ cor}$ without emitter after corona	5	5	9	8	18	13
Measured $J_{oe\ diff\ cor}$ with emitter after corona	22	25	26	27	22	26
Contribution from surface $J_{oe\ diff\ cor} - J_{oe\ ideal}$	0	3	4	5	0	4

TABLE II

CALCULATED FIXED NEGATIVE CHARGE DENSITY Q_f , AND MIDGAP INTERFACE STATE DENSITY D_{it} FROM CV MEASUREMENTS FOR BOTH (1 1 1) AND (1 0 0) SUBSTRATES PASSIVATED BY PA-ALD, H₂O-ALD, AND O₃-ALD Al₂O₃ LAYERS, RESPECTIVELY

	PA-ALD Al ₂ O ₃		H ₂ O-ALD Al ₂ O ₃		O ₃ -ALD Al ₂ O ₃	
	(111)	(100)	(111)	(100)	(111)	(100)
Fixed negative charge density Q_f (10 ¹² cm ⁻²)	1.3±0.2	2.7±0.3	1.8±0.2	2.8±0.3	3.5±0.5	4.0±0.5
Midgap interface state density D_{it} (10 ¹⁰ eV ⁻¹ cm ⁻²)	7.8±1.4	5.4±1.0	12.3±3	14.4±3	31.1±6	26.2±5

Therefore, the additional surface B doping from the diffusion can reduce recombination at the Al₂O₃/Si interface in the case of relatively low surface B doping concentration ($\sim 2 \times 10^{19}$ cm⁻³ for the samples that are used in this study). This is likely to be largely the result of lower minority carrier densities at the surface of diffused samples, regardless of charge density in the film. The result also suggests that the presence of a light or moderate B diffusion does not introduce significant additional defects at the interface compared with an undiffused interface, in contrast with thermal oxide-passivated Si surfaces [11], although the introduction of some additional defects cannot be ruled out on the basis of these results.

C. Impact of Surface Orientation

Fig. 2 shows that the J_{oe} of (1 1 1) samples is consistently higher than that of (1 0 0) samples, regardless of deposition method. The difference tends to be larger for undiffused than for diffused samples. For undiffused samples, the difference is up to a factor of 2. One possible explanation is that the Al₂O₃-passivated silicon surface with a (1 1 1) orientation is associated with a comparatively higher defect density than the (1 0 0) surface, as was observed of silicon nitride-passivated surfaces [26]. However, it is interesting to note that the application of corona charge effectively removes the difference in J_{oe} values. This observation is not easy to reconcile with an explanation that is based on different interface defect densities. An alternative explanation is that the charge density in the films on (1 1 1) substrates may be lower than on (1 0 0) substrates.

Fig. 4 shows the $C-V$ and D_{it} curves that are obtained from MOS structures with Al₂O₃ films that are synthesized using PA-ALD and annealed at 400 °C in forming gas for 30 min. The D_{it} curves were determined by the application of the method of Castagne and Vapaille [27]. The shift of the $C-V$ curves to the left for the (1 1 1) sample compared with the (1 0 0) sample indicates a lower negative charge density in the (1 1 1) samples.

Extracted values of the fixed negative charge density Q_f and interface state density at midgap D_{it} are summarized in Table II. In terms of the negative fixed charge density Q_f , substrates with a (1 0 0) orientation display consistently higher values than (1 1 1) samples, with the difference being largest for PA-ALD-based films. The reasons for this interesting phenomenon require further investigation.

In contrast with the fixed charge density, the interface state density D_{it} was not strongly influenced by surface orientation of the substrate. Low D_{it} values at midgap were obtained for PA-ALD- and H₂O-ALD-based Al₂O₃/Si interfaces, with about 6.0×10^{10} , 1.3×10^{11} eV⁻¹cm⁻², respectively, while significantly higher D_{it} values were extracted for O₃-ALD Al₂O₃/Si interfaces, which can explain the worse passivation of O₃-ALD Al₂O₃ layers on undiffused samples both before and after negative corona charging. Compared with the data shown elsewhere [6], for PA-ALD Al₂O₃ layer, much lower D_{it} value is obtained in this paper, but Q_f falls by half. As far as we know, Q_f can be optimized to as high as -1×10^{13} cm⁻² [28]. This illustrates that PA-ALD can be optimized potentially to afford versatile properties for possible applications.

The Q_f and D_{it} values of the films suggest that differences in fixed charge density are partly responsible for the observed differences in passivation between (1 0 0) and (1 1 1) surfaces. However, the differences in charge density cannot fully explain the differences in passivation for O₃-based films, for which the difference in charge density is relatively small.

IV. CONCLUSION

In conclusion, PA-ALD, H₂O-ALD, and O₃-ALD Al₂O₃ films can effectively passivate both B-diffused and -undiffused Si surfaces. For (1 0 0)-orientated substrates with symmetrical B emitters (85Ω/□), J_{oe} of 24, 29, and 33 fA/cm² was obtained for PA-ALD, H₂O-ALD, and O₃-ALD Al₂O₃ passivation, respectively, and corresponding J_{oe} values from undiffused samples

are 6, 9, and 16 fA/cm². Generally, the sample surface orientation, or presence or absence of diffusion, has little effect on the annealing behavior, with the exception of H₂O-based films, where the annealing behavior of undiffused and diffused films differs slightly.

Negative corona charging was used to study the influence of extra B diffusion on ALD Al₂O₃ passivation performance. Results before corona and post corona charging illustrate that light to moderate B diffusions can reduce recombination at the ALD Al₂O₃/Si interface.

J_{oe} values after annealing show that the (1 0 0) surface is easier to passivate than (1 1 1), while the application of negative corona charge effectively removes the difference in J_{oe} values. $C-V$ measurements show that the charge density in the films on (1 1 1) substrates is significantly lower compared with (1 0 0) substrates, while differences in interface defect density are relatively small. The difference in fixed charge density, therefore, appears to be a significant contributor to the differences in passivation between (1 0 0) and (1 1 1) crystal orientations.

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Authors' photographs and biographies not available at the time of publication.