Exploratory studies of novel silicon surface passivation materials: Considerations and lessons learned

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ABSTRACT

Despite the existence of several highly effective and well-characterized passivating materials for crystalline silicon surfaces, the topic of surface passivation and the investigation of new passivating materials remain of considerable interest for silicon photovoltaics research. However, the question of whether and under what circumstances a particular material will provide effective surface passivation remains difficult to answer. In this work, we provide an overview of recent insights relating to this question, drawing from our own work on novel passivation materials including MoO\textsubscript{3}, Nb\textsubscript{2}O\textsubscript{5}, TiO\textsubscript{2}, ZrO\textsubscript{2}, and PO\textsubscript{3}, and illustrated with experimental results. Factors that strongly influence the passivation performance include the use of pre-grown interfacial oxides, the film thickness, the annealing conditions, and the presence of capping layers. The impact of these factors on the surface passivation can vary widely from material to material. Therefore, all of these factors should be taken into account when investigating potential new surface passivation materials.

1. Introduction

Effective surface passivation is critical to realizing high-efficiency crystalline silicon (c-Si) solar cells and is becoming ever more important as wafer thicknesses are reduced and bulk Si quality improves. At the same time, new demands on the functionality of passivation layers have emerged, such as the ability to form passivating contact structures that in addition to passivation, enable the selective extraction of either positive or negative charges from the Si bulk \cite{1,2}. These demands have stimulated significant research efforts aimed at evaluating the passivation properties of novel materials \cite{1–9}. As a result, in the last decade, the list of materials known to be capable of effectively passivating c-Si surfaces has expanded significantly. The well-established Si-based passivation layers (SiO\textsubscript{x}, SiN\textsubscript{y}, a-St:H \cite{12}, and more recently proven Al\textsubscript{2}O\textsubscript{3} \cite{13–16}) have been joined by a range of new materials including HfO\textsubscript{2} \cite{15,17,18}, TiO\textsubscript{2} \cite{19–23}, Ta\textsubscript{2}O\textsubscript{5} \cite{24} and Ga\textsubscript{2}O\textsubscript{3} \cite{25–27}, and new passivation materials continue to be reported, such as ZnO \cite{28}, Nb\textsubscript{2}O\textsubscript{5} \cite{29,84}, PO\textsubscript{3} \cite{30,85}, and ZrO\textsubscript{2} \cite{31}. Table 1 gives a summary of state-of-the-art passivation results for these materials and a few other notable examples. Note that some of these passivating materials can also serve as passivating contacts, as apart from preventing minority carrier recombination at the c-Si surface, they also yield a sufficiently low contact resistance to c-Si for majority carriers. In Fig. 1, transmission electron microscopy (TEM) images for a selection of these passivating materials are shown.

The mechanism of surface passivation is generally understood to involve a reduction in the number of interface states ("chemical" passivation) and/or suppression of the concentration of either electrons or holes at the semiconductor surface (historically referred to as "field-effect" passivation) \cite{19–23} (in general, effective passivation almost always involves a combination of these two mechanisms). However, the reasons for when and why a particular material will provide passivation are still not entirely clear. Some materials that were long believed not to provide effective surface passivation have now been shown to do so under the right conditions. For instance, TiO\textsubscript{2} was replaced by SiN\textsubscript{y} as the preferred anti-reflection coating (ARC) for c-Si solar cells in the late 1990s due to the more effective passivation provided by SiN\textsubscript{y}, but it has recently been shown that TiO\textsubscript{2} can in fact also provide excellent passivation \cite{19–23}. On the other hand, other materials that are potentially relevant to c-Si photovoltaics thus far do not seem to provide passivation, regardless of their preparation method and post-deposition treatment. However, the growing number of studies of different...
Typical polarity (n or p) of the silicon space-charge region is specified. n+ or p+ indicate particularly strong band bending in the silicon, while p/n indicates that band bending is tunable through doping. We also specify the passivation layer thickness and whether a pre-grown SiO2 (thermally or chemically grown) was present. Materials that are capable of forming passivating contacts for electron or holes are indicated. 

Table 1

<table>
<thead>
<tr>
<th>Material</th>
<th>Bandgap (eV)</th>
<th>$S_{\text{eff, max}}$ (cm/s)</th>
<th>Induced surface charge</th>
<th>Preparation Method</th>
<th>Pre-grown SiO2</th>
<th>Thickness (nm)</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>SiO2</td>
<td>9</td>
<td>2.4</td>
<td>n</td>
<td>Thermal oxidation</td>
<td>–</td>
<td>–</td>
<td>110</td>
</tr>
<tr>
<td>SiNx</td>
<td>1.8–5.3</td>
<td>3.5</td>
<td>n+</td>
<td>PECVD</td>
<td>–</td>
<td>&gt; 50</td>
<td>Common ARC/passivation for n+ Si in homojunction cells</td>
</tr>
<tr>
<td>a-Si:H</td>
<td>1.6–1.8</td>
<td>0.7</td>
<td>p/n</td>
<td>PECVD</td>
<td>–</td>
<td>280</td>
<td>Can form e- or h-contact</td>
</tr>
<tr>
<td>poly-Si</td>
<td>– 1.1</td>
<td>0.4</td>
<td>p/n</td>
<td>LPCVD</td>
<td>therm.</td>
<td>150</td>
<td>Can form e- or h-contact</td>
</tr>
<tr>
<td>Al2O3</td>
<td>6.4</td>
<td>1.3</td>
<td>p+</td>
<td>ALD</td>
<td>–</td>
<td>30</td>
<td>Common passivation for p+ Si in PERC cells</td>
</tr>
<tr>
<td>PO4</td>
<td>&gt; 5</td>
<td>2.7</td>
<td>n+</td>
<td>ALD</td>
<td>–</td>
<td>6</td>
<td>–</td>
</tr>
<tr>
<td>HfO2</td>
<td>5.3–5.7</td>
<td>3.3</td>
<td>n</td>
<td>ALD</td>
<td>–</td>
<td>15</td>
<td>–</td>
</tr>
<tr>
<td>Ta2O5</td>
<td>4.4</td>
<td>6.1</td>
<td>p</td>
<td>ALD</td>
<td>–</td>
<td>12</td>
<td>–</td>
</tr>
</tbody>
</table>
| TiO2     | 3.5         | 3.7                          | p                     | ALD               | –              | 15             | Former use as ARC.
| GeO2     | 4.2–5.2     | 6.4                          | p+                    | ALD               | –              | 4.5            | Can form h-contact with a-Si:H |
| MoO2     | 2.8         | 29                           | p+                    | ALD, chem.        | 10             | –              | This work |
| Nb2O5    | 3.6         | 6.8                          | p+                    | ALD, chem.        | 5              | Can form e-contact |
| Al      | 6.2         | 12                            | p                     | Sputtering        | chem.          | 50             | – |
| ZnO      | > 3         | 8.5                          | n                     | ALD, chem.        | 73             | –              | Common TCO, n-doping possible |
| ZrO      | 5–6         | 9.1                          | p                    | ALD               | –              | 20             | – |
|           |             |                              |                       |                   |                |                |         |

Materials means that significantly more data relating to this question are becoming available.

In our group, we have investigated a large number of materials prepared by atomic layer deposition (ALD) for surface passivation of c-Si. ALD provides distinct advantages for surface passivation studies, such as large-area uniformity, precise thickness control, and damage-free (‘soft’) deposition [71]. Materials that have recently been examined include Nb2O5, PO4, ZnO, TiO2, and MoO2. In this work, we will highlight several trends and exemplary cases taken from these and other datasets which provide insight into the factors that are important to surface passivation, with a particular emphasis on results that deviate from or go beyond conventional wisdom. While such examples do not yet permit us to outline a step-by-step procedure for reliably achieving surface passivation with new materials, they do highlight some important factors that ought to be taken into account when investigating novel materials for surface passivation properties. In particular, we will highlight the role of pre-grown interfacial oxides (prepared by e.g. a standard Radio Corporation of America (RCA) clean or a pre-grown interfacial oxide (pre-grown SiO2, poly-Si, and Ta2O5). Since there are already a number of different materials, and the potential role of capping layers in providing significant bulk lifetime. In general, we have selected the references with the lowest reported $S_{\text{eff, max}}$ in this resistivity range, giving preference to reports on more highly doped substrates in case of similar values. Representative values for the bandgap are given (additional references for these bandgap values have been included in cases where these are not mentioned in the reference for $S_{\text{eff, max}}$). The typical polarity (n or p) of the silicon space-charge region is specified. n+ or p+ indicate particularly strong band bending in the silicon, while p/n indicates that band bending is tunable through doping. We also specify the passivation layer thickness and whether a pre-grown SiO2 (thermally or chemically grown) was present. Materials that are capable of forming passivating contacts for electron or holes are indicated.

3. Results

3.1. Surface preparation and the role of pre-grown interfacial oxides

The conventional silicon surface preparation procedure prior to surface passivation involves one or several wet-chemical cleaning steps (e.g. a Radio Corporation of America (RCA) standard clean [49]) which typically results in the growth of a thin (1–2 nm) surface oxide. Immediately before the deposition of a passivating film, this oxide is removed by etching in dilute hydrofluoric acid (HF) to leave a hydrophobic, hydrogen-terminated surface. The latter is generally considered the ideal starting surface for passivation, as such low-temperature wet chemical oxides are generally regarded as being of low-quality — usually they provide negligible passivation on their own — and their removal enables the formation of a new high-quality interface. This is well-established for example for thermal SiO2 [50], PECVD SiNx [51] and PECVD Si:H [34]. Note that, at least in the case of oxide materials, the formation of this new interface generally involves the growth of a new interfacial SiOx layer either during deposition or upon subsequent annealing, (as visible in the TEM images shown in Fig. 1 of Al2O3, TiO2, and Nb2O5 deposited on HF-treated Si), where the thickness and other properties of this layer will depend on the specific processing conditions and interfacial chemistry.

Contrary to this picture, we have found that in some cases retaining the oxide grown during surface cleaning, or intentionally growing a thin chemical oxide, can be beneficial for surface passivation by metal

2. Experimental

As substrates for the ALD films discussed in this study we used double-side-polished (100) 1–5 μcm n-type floatzone (FZ) Si wafers. Films were deposited by ALD in Oxford Instruments FlexAL (in case of Nb2O5 [29], TiO2 [43], and Al2O3 [44]) or OpAL (in case of MoO2 [42,45] and H, Al or B doped ZnO [46-48]) reactors. Annealing was performed using a Jipelec rapid thermal annealing system. The passivation quality was assessed by transient and quasi-steady-state photoconductance (QSSPC) measurements using a Sinton WCT-120TS lifetime tester. Film thickness was evaluated using a J.A. Woollam M2000U variable angle spectroscopic ellipsometer.
oxides. In fact, using a starting surface with an existing thin oxide layer can sometimes even result in passivation in cases where the same procedure applied to a HF-treated surface would not. For instance, MoO$_x$, which is being explored as hole-selective contact [52, 56, 57], is a material which typically yields very poor surface passivation when directly applied to the Si surface [54-57]. For this reason, it is either used as partial rear contact [56] or combined with a dedicated amorphous silicon passivation layer [52,53] as shown in Fig. 1(c). However, we recently found that MoO$_x$ prepared by ALD [42,45] can yield surface passivation when it is deposited on an RCA oxide and subsequently subjected to a forming gas anneal (FGA) treatment. This is shown in Fig. 2. These results suggest that the interfacial oxide between MoO$_x$ and HF-treated silicon — that naturally forms during processing — is of poor quality and that this interfacial oxide cannot be made to passivate by an FGA treatment. Conversely, the well-defined RCA oxide can be used to achieve passivation by MoO$_x$, with the best results being obtained after a one hour FGA treatment at 300 °C.

The beneficial effects of an RCA oxide are also observed for ALD ZnO:B/Al$_2$O$_3$ stacks. Such stacks yield only moderate levels of passivation on HF-treated Si surfaces, whereas excellent passivation is achieved on an RCA oxide (see Fig. 1(f)). Therefore, also in the case of ALD ZnO it seems that the oxide that forms during deposition on HF-treated Si is not of high quality. Conversely, for ALD Nb$_2$O$_5$ good passivation is achieved on both HF-treated Si and Si with an RCA SiO$_x$ layer. This shows that for this process a SiO$_x$ layer is formed during deposition and post-deposition annealing that enables good surface passivation [29]. This interfacial SiO$_x$ layer is also visible in Fig. 1(e).

Fig. 1. Cross-sectional bright-field transmission electron microscopy images of various surface passivation materials on n-type c-Si (100). (a) ALD Al$_2$O$_3$ prepared at 200 °C on HF-treated Si after a 450 °C forming gas anneal (FGA) [13]. (b) p-type poly-Si on SiO$_x$ prepared by low-temperature oxidation [39]. (c) ALD MoO$_x$ on a-Si:H prepared by chemical vapor deposition [40] and capped with ALD In$_2$O$_3$H [41], after annealing at 200 °C in N$_2$ [42]. (d) ALD TiO$_x$ prepared at 200 °C on HF-treated Si (e) ALD Nb$_2$O$_5$ prepared at 150 °C on HF-treated Si and subjected to a 300 °C FGA [29] (f) ALD boron-doped ZnO prepared at 200 °C on c-Si with an RCA oxide and subjected to a 450 °C FGA.

Fig. 2. Effective excess carrier lifetime for 10 nm ALD MoO$_x$ films, 5 nm ALD Nb$_2$O$_5$ films, or 75/30 nm boron-doped ZnO/Al$_2$O$_3$ stacks prepared at 200 °C on n-type Si (3.3 Ω cm), either HF-treated or with an RCA oxide. The Nb$_2$O$_5$ sample was prepared at 150 °C, whereas the MoO$_x$ and the boron-doped ZnO/Al$_2$O$_3$ samples were prepared at 200 °C. The MoO$_x$ and Nb$_2$O$_5$ films were subjected to 1 h optimized post-deposition FGA treatments of 1 and 2 h at 300 °C, respectively. The ZnO:B/Al$_2$O$_3$ stacks received a 30 min FGA at 450 °C.
The use of a chemical oxide can also strongly alter the passivation behavior of materials that do provide passivation to HF-treated Si. This is clearly illustrated by the case of ALD Nb2O5 [29]. On HF-treated Si, the surface passivation quality improves when using thicker Nb2O5 films and lower deposition temperatures, as shown in Fig. 3. However, the use of an RCA oxide not only yields a better passivation quality, but in addition it renders the passivation level much less dependent on the experimental conditions. Strikingly, this even enables excellent passivation of the surface using only 1 nm of ALD Nb2O5. The sensitivity of the passivation quality on HF-treated Si surfaces to the experimental conditions is thought to originate from differences in the formation of the interfacial oxide during deposition [29]. It is known that SiOx interfaces formed by different chemical processes [58] and by the same process at different temperatures (e.g. thermal oxidation [59]) can exhibit significant differences in interface state and fixed charge densities, which determine passivation quality. Conversely, the fact that the interfacial oxide is formed prior to Nb2O5 deposition in the case of RCA-treated surfaces renders such surfaces much less sensitive to the exact process conditions.

The nature of the pre-grown oxide is also important. Ultrathin oxides may be prepared by a wide variety of methods, including growth in air (native), wet-chemical (e.g. nitric acid oxidation (NAOS), RCA, boiling water), UV/ozone, O2 plasma, and thermal oxidation, and these can yield significantly different levels of surface passivation [60–67]. An example of this can be seen from the comparison of lifetimes for TiOx-passivated samples with RCA and NAOS oxides in Fig. 4, where the RCA oxide results in significantly better passivation than the NAOS. Note that, based on our own data and that of others, it does not so far appear possible to establish a simple generally applicable correlation between the readily observable macroscopic properties (thickness, stoichiometry/oxidation state) of a given SiOx layer (whether pre-grown or formed during processing) and its suitability for passivation, so that this needs to be tested directly in each case.

Note that beneficial effects from pre-grown interfacial oxides have been reported before. For example, Mihalec et al. reported that the use of a 1.5 nm oxide grown by NAOS enabled effective passivation of p+ Si surfaces by PECVD SiNx [68]. Ultrathin oxides are also routinely employed as interlayers in polycrystalline silicon (poly-Si) passivating contacts [69–72].

While the use of a pre-grown oxide can enable passivation in some cases, in other cases an HF-treated starting surface results in better passivation, as seen for example for the TiOx films deposited at 50 °C in Fig. 4. This is also generally the case for ALD Al2O3 which provides a better surface passivation quality on HF-treated silicon in comparison to chemically pre-oxidized silicon [73]. Generally, it appears that some materials and some deposition processes are capable of forming their own high-quality interfaces, while others benefit from the presence of a pre-formed oxide interface. Therefore, the influence of the starting surface needs to be investigated for different materials and processes on an individual basis.

### 3.2. Influence of film thickness

The thickness of a passivation layer is another known parameter affecting passivation quality. It is conventionally expected that very thin films will provide worse passivation than thicker films. More specifically, the passivation quality is usually observed to improve with film thickness up to a certain point (typically ~10 nm) and then to saturate. This is well-established for example for thermal SiO2 [50], PECVD SiNx [51], and ALD Al2O3 [74–76]. At least in case of Al2O3, which owes its excellent passivation partly to a strong field-effect passivation, it is the chemical passivation which deteriorates when moving to very thin films [77]; this may also apply to other passivating materials. Note that for a-Si:H, saturation in fact appears not to occur even for thicknesses up to several hundred nm, but the passivation quality is still observed to improve monotonically with thickness [34].

In contrast to this simple picture, we have found that for some materials passivation is not a monotonic function of film thickness. In certain cases there is a clear optimum at a relatively low thickness value, as is illustrated in Fig. 5 for TiOx after an FGA treatment. A similar optimum thickness behavior was also observed for TiOx by Yang et al. who also used an FGA treatment [22,23]. Consequently, if only relatively thick films had been investigated in these cases, as is often the case, passivation would have appeared considerably worse.

Our results and those of Yang et al. [22,23] are in contrast to those of Cui et al., who used an N2 annealing treatment and observed saturation of passivation for TiOx thicknesses between ~8 and 63 nm, which resembles the classical case where the passivation quality saturates above a certain layer thickness [21]. Therefore, the thickness dependence of passivation appears not to be an intrinsic property of TiOx, but depends on the processing history of the films. Since each of these studies employed different ALD precursors, deposition temperatures, and annealing treatments, these may account for the difference in passivation behavior.

Alternatively, for some materials saturation may occur, but much more rapidly than for SiO2 or Al2O3. Indeed saturation of passivation...
can occur after only a few monolayers for some materials. An example of this was shown in Fig. 3, where passivation by Nb2O5 apparently saturates already after 1 nm on RCA oxide surfaces. Film thickness can therefore influence passivation in unexpected ways and the influence of film thickness on passivation should be carefully considered especially when investigating new materials.

3.3. Post-deposition annealing: Temperature optima and the role of capping layers

It is well-known that a post-deposition annealing treatment, either in N2 or in a hydrogen-containing gas mixture (FGA), can improve the surface passivation quality. This is for instance the case for materials such as SiO2, a-Si:H, SiN, and Al2O3 [16]. The general aim of such an annealing treatment is to effectively hydrogenate and passivate the dangling bonds present at the silicon surface, i.e. to improve the chemical passivation. Generally, a sufficiently high temperature has to be applied to enable the diffusion of hydrogen from either the passivating film itself or from a hydrogen-containing ambient (e.g. forming gas) towards the silicon interface, while at the same time avoiding excessive hydrogen loss from the film via effusion. In the case of both SiO2 and Al2O3, the optimal annealing temperature is known to be close to 400 °C [16], while much lower temperatures around 180 °C appear optimal for a-Si:H [78]. This already indicates that the optimal annealing temperature may vary from material to material.

This point is made even more obvious when looking at the wide range of novel passivating materials, as is shown in Fig. 6. It is striking to see that the optimal annealing temperatures for MoO3, Nb2O5, Al2O3, ZnO:Al, and PO4 are all different, indicating different annealing kinetics for each of these materials. It should be noted that since the annealing step time was not equal for all of the materials, a one-to-one comparison cannot be made for all the materials in this graph. Nonetheless, the observed trends should still be dominated by the temperature rather than by the anneal step time, since the rate of thermally-activated processes depends much more strongly on temperature. This is for example seen in the case of SiO2/Al2O3 stacks, where the apparent saturated level of passivation differs for different annealing temperatures [79].

When comparing the three samples on the RCA oxide, i.e. MoO3, Nb2O5, and ZnO:Al/Al2O3, it can be seen that the ZnO:Al/Al2O3 sample has a much higher onset and optimum temperature. This is striking if one assumes that the annealing kinetics depend on the hydrogenation reaction at the interface, as it is the same RCA SiO2 interface that needs to be passivated during the anneal treatment. These differences are therefore likely linked to differences in the kinetics of hydrogen diffusion through the material and toward the interface and effusion of hydrogen from the material. Conceivably, the transport of hydrogen takes place more readily and at lower temperatures in the amorphous MoO3 and Nb2O5 films as compared to the polycrystalline ZnO/Al2O3 stack. Since diffusion plays a role, film thickness is also likely to play a role in the exact annealing trend. Also material stability and possible crystallization during annealing can play an important role in degradation at higher temperatures and thus the location of the optimum. For the case of MoO3, macroscopically visible formation of etch pits as well as a reduction of the material from near-stoichiometric MoO3 to MoO2 was observed by XPS when performing FGA treatments exceeding 300 °C, explaining the strong reduction in lifetime observed.

Capping of passivation layers prior to annealing by other materials, most notably Al2O3 and SiN, has proven to be a highly successful approach to improve or enable passivation. As illustrated in Fig. 7, it has for example been demonstrated that applying an Al2O3 capping layer can greatly enhance the passivation provided by ALD SiO2, SiO2/poly-Si, and SiO2/ALD ZnO. For SiO2 and poly-Si, it has been shown that the Al2O3 layer serves both as a hydrogen source and a hydrogen effusion barrier [78,80,81]. Note that hydrogen effusion barriers can also be formed by other materials, such as SiN [82]. Conversely, in the case of ALD ZnO, ample hydrogen for passivation is already present at the SiO2 interface and in the ZnO layer after the ALD process. In this case, the Al2O3 capping layer merely serves as a hydrogen effusion barrier that prevents a significant effusion of hydrogen at 400 °C [28].

However, the success of such Al2O3 capping layers does not necessarily translate to other passivation materials, as Fig. 7 shows for the cases of Nb2O5, MoO3 and TiO2. For these three materials, the addition of the Al2O3 capping layer leads to a strong reduction in passivation. In the case of MoO3, capped with Al2O3, the deterioration is likely related to the visible formation of blisters when annealing at 300 °C. Such blistering was not observed for Nb2O5 and TiO2 capped by Al2O3, and the precise reasons for the detrimental effect of capping by Al2O3 are so far not known. Conceivably, the photocatalytic nature of both TiO2 and Nb2O5 plays a role. Both TiO2 and Nb2O5 yield improved passivation upon light soaking, which is thought to originate from an enhanced
The main conclusion that can be drawn from this exploration is the fact that all these considerations strongly influence the passivation performance of most passivation materials. In addition, the optimal choice of one experimental parameter is also often found to depend on the choice of another. Therefore, the interdependence of all these factors needs to be carefully considered when investigating new materials for surface passivating properties. These lessons can prove useful in the design and optimization of future passivation schemes and passivating contacts for c-Si solar cells.

Acknowledgements

The authors gratefully acknowledge the Top consortia for Knowledge and Innovation Solar Energy programs “AAA” (grant 1409104), “RADAR” (grant TEUE116905), and “COMPASS” (grant TEUD215022) of the Ministry of Economic Affairs of the Netherlands for financial support. The work of J. Melskens was supported by the Netherlands Organisation for Scientific Research under the Dutch TTW-VENI grant 15896. Marcel A. Verheijen is acknowledged for performing the TEM studies. Solliance and the Dutch province of Noord-Brabant are acknowledged for funding the TEM facility. Additionally, we would like to acknowledge J.H. Deijkers, V.H.M. Evers, and J. Palmins from Eindhoven University of Technology (Netherlands) for useful discussions and support during the experiments.

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