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**Molecular dynamics studies of the bonding properties of amorphous silicon nitride coatings on crystalline silicon.**

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In this paper we present molecular dynamics simulations of silicon nitride, both in bulk and as an interface to crystalline silicon. We investigate, in particular, the bonding structure of the silicon nitride and analyze the simulations to search for defective geometries which have been identified as potential charge carrier traps when silicon nitride forms an interface with silicon semiconductors. The simulations reveal how the bonding patterns in silicon nitride are dependent upon the stoichiometry of the system. Furthermore we demonstrate how having an “interphase”, where the nitrogen content in silicon gradually reduces towards pure silicon across a boundary region, as opposed to an interface where there is an abrupt drop in nitrogen concentration at the boundary, can result in significantly different numbers of certain important carrier trap

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## I. INTRODUCTION

The deposition of amorphous silicon nitride ( $\text{SiN}_x$ ) on silicon semiconductors in photovoltaic cells has become a popular strategy for a number of reasons. Initially  $\text{SiN}_x$  layers were introduced for the formation of metal-insulator-semiconductor structures<sup>1</sup>. In solar cells they were first used due to their properties as an anti-reflective coating (ARC). The presence of an ARC in crystalline silicon (c-Si) solar cells is important as a clean c-Si surface reflects more than 35% of incident light. The presence of an ARC allows for effective utilization of this light. That the inclusion of a  $\text{SiN}_x$  layer increases the performance of solar cells by more than would be expected merely from its role as an ARC has been explained by its ability to reduce the rate at which excess charge carriers recombine at the semiconductor surface<sup>2</sup>, a property known as passivation. Thus  $\text{SiN}_x$  coatings address the two major loss mechanisms in solar cells, namely optical and electrical losses. Due to these dual beneficial properties for the improvement of photovoltaics as well as its importance in metal-oxide-semiconductor field effect transistors (MOSFETs) the silicon/silicon nitride interface has been the subject of much research since the early 1990's<sup>3-5</sup>.

Much of the work done on ARC development to date has focused on finding the optimal ratio of N to Si (i.e. the value of  $x$  in  $\text{SiN}_x$  in the layer) in order to maximize the percentage of incident light trapped in the c-Si layer<sup>6</sup>. There is a balance to be achieved between reducing reflection (this is minimized by increasing the refractive index ( $n$ ) of the ARC) and reducing the amount of light absorbed by the ARC. One problem is that in order to reduce the amount of light absorbed in the ARC it is necessary to reduce the number of Si-Si and Si-H bonds (i.e. higher values of  $x$ ), however to increase the refractive index lower values of  $x$  are desirable. If one were to ignore the effects of absorption of light a value of  $n=2.3-2.4$  would be optimal. The optimal value of  $x$  and hence  $n$ , is very dependent upon the type of solar cell, the optimal wavelength, the texture, the module fabrication etc, for example in air a value of  $n=1.9$  is optimal, however generally in solar cells in modules values of  $n = 2.04 - 2.08$  are preferred<sup>7,8</sup>.

Although much of the attention in the development of  $\text{SiN}_x$  layers for high performance solar cells has been focused on the optimization of the two roles as an ARC and a passivation layer<sup>9-14</sup> it is only recently that the inter-relatedness of these two functions has been elucidated. Jung *et al.*<sup>15</sup> performed a comprehensive study of band gap and defect

effects in amorphous silicon nitride (a-SiN<sub>x</sub>), indicating the important role of these effects in determining the efficiency of a solar cell.

There has been a lot of experimental work done in characterizing the stoichiometric Si(111)/Si<sub>3</sub>N<sub>4</sub> interface. Much of this effort has concentrated on identifying the types of surface reconstruction and charge transfers involved at the interface.  $7 \times 7^5$ ,  $4 \times 4^{16}$ ,  $8/3 \times 8/3^{17}$  and  $8 \times 8^3$  reconstructions have all been observed. The Si(100)/SiN<sub>x</sub> interface has been studied by electron paramagnetic resonance (EPR)<sup>18</sup>, which revealed the major electron trap in the interface layer to be due to under-coordinated Si. Further studies of this interface with transmission electron microscopy (TEM) showed that N atoms were present on the c-Si side of the interface up to the second layer<sup>19</sup>. Such penetration of N, into the c-Si layer is also known to be extremely sensitive to the conditions during the deposition of the ARC<sup>8</sup>.

In a-SiN<sub>x</sub> a number of geometrical defects have been identified which have important implications for the electronic structure, two of the most important, the so-called K and N defects are shown in figure 1. The K centre consists of a Si bound to three N atoms with a dangling bond ( $N_3 \equiv Si$ ) and is observable by electron spin resonance (ESR)<sup>20</sup>. A defect centre due to an under coordinated N atom ( $Si_2 = N$ ) has also been observed<sup>21</sup>. These assignments have also been supported by *ab initio* calculations<sup>22</sup>.

FIG. 1. (Color online) Important defect geometries in the SiN<sub>x</sub> layer. a) the K-defect, where Si is bonded to three N atoms with one dangling bond. b) the N defect, where N is bonded to two Si atoms with one dangling bond.

In this paper we simulate the Si(100) interface in contact with SiN<sub>x</sub>:H at an atomistic level. SiN<sub>x</sub>:H layers with  $x=0.8, 0.9, 1.0, 1.1$  and  $1.2$ , incorporating 20 at. % H are simulated. For each value of  $x$  we simulate both a clean interface, with an abrupt SiN<sub>x</sub>:H/c-Si partition and a gradual interface with a gradient of  $x$  across a distance of 20 Å. We shall refer to these interfaces as clean and gradual interfaces respectively. We perform a topological analysis of the trajectories generated in our molecular dynamics (MD) simulations. We calculate the radial distribution functions (RDFs) for bulk SiN<sub>x</sub> comparing to previous theoretical work, as a validation of our method and investigate bonding patterns in the interfacial regions of our models to identify and quantify defect centres, revealing a clear difference in the frequency of occurrence and degree of passivation of K and N defects at clean and gradual interfaces.

| x   | Si Atoms | N Atoms | H Atoms | Density<br>(g/cc) |
|-----|----------|---------|---------|-------------------|
| 0.8 | 415      | 332     | 187     | 2.86              |
| 0.9 | 409      | 368     | 194     | 2.92              |
| 1.0 | 404      | 404     | 202     | 2.98              |
| 1.1 | 400      | 440     | 210     | 3.04              |
| 1.2 | 394      | 473     | 217     | 3.10              |

TABLE I. Numbers of atoms in the simulation cells used to generate the a-SiN<sub>x</sub>.

## II. METHODS

### A. Generating amorphous SiN<sub>x</sub>

All interactions in the system were modelled using Tersoff type potentials<sup>23</sup>. We use the parameterization of these potentials developed by de Brito Mota *et al.*<sup>24</sup>, which were previously used to simulate amorphous SiN<sub>x</sub><sup>24</sup> and SiN<sub>x</sub>:H<sup>25</sup>. All MD simulations presented in this paper were run using the DLPOLY code<sup>26</sup>.

The first step in our procedure is to generate samples of a-SiN<sub>x</sub>. This is achieved by randomly placing Si and N atoms in a simulation cell, subject to certain interatomic separation criteria, in the correct ratio. The cell volumes were set by estimation of the required density, from a linear interpolation between the density of c-Si and c-Si<sub>3</sub>N<sub>4</sub>. The details of the simulation cells are presented in table I.

In order to generate the amorphous samples the initially random configurations were run at 2500 K for 0.5 ns with a timestep of 1 fs in the NVT ensemble. After this simulation, the systems were then quenched to 350 K at a rate of  $3 \times 10^{11}$  K/s. We tested various quenching rates on samples of a-Si<sub>3</sub>N<sub>4</sub> and found that this was the fastest rate at which no significant increase in the number of defects present in the sample was observed, compared to quicker rates.

We then add 20 at. % H (table I) to the simulation cells and equilibrate, again in the NVT ensemble at 2500K for 0.5 ns. In this instance a reduced timestep of 0.2 fs is employed due to the presence of H. The systems were quenched as above.

## B. Generating c-Si/SiN<sub>x</sub>:H interfaces

In order to generate clean and gradual interface structures two different procedures were followed. To generate the clean interface the amorphous sample was placed in a cell at the (100) surface of a 20 layer thick slab of c-Si. The a-SiN<sub>x</sub> was placed 2.5 Å away from this surface and dynamics were performed at 2500 K for 0.1 ns, with the c-Si held in position. The a-SiN<sub>x</sub> layer was then gradually moved closer to the surface, each time repeating the procedure until the energy of the system started to increase relative to the previous step. Having thus identified the minimum energy separation, the top two layers of the c-Si were un-locked (as illustrated in figure 2) and dynamics were performed for a further 0.5 ns at 2500 K. The system was then quenched at the same rate as previously to 350 K, where dynamics were performed for a further 0.5 ns to generate statistics.

To generate the system with the gradual interface the minimum energy separation from the previous step was again used. In this instance, however, the top 10 layers of c-Si were unlocked before performing the dynamics at 2500 K. This results in the N atoms from the SiN<sub>x</sub> penetrating into the c-Si layer, which also melts and becomes distorted. After 0.5 ns at 2500 K the system is annealed to 350 K as described above, and run for 1 ns at 350 K to generate statistics.

All bond pattern analyses were performed using the R.I.N.G.S. code<sup>27</sup>. During this analysis two atoms are said to be bonding if the interatomic separation is within the first peak of the partial radial distribution function.

FIG. 2. (Color online) Simulation cell setups for creating clean and gradual interfaces. The lines to the right indicate the  $z$  coordinates over which atoms are frozen during interface generation. The dark line is c-Si frozen during clean interface generation, the grey line is c-Si frozen during gradual interface generation. The line at the top represents a-SiN<sub>x</sub> frozen during all simulations.

## III. RESULTS AND DISCUSSION

### A. Bulk amorphous SiN<sub>x</sub>

All total and partial RDFs for the bulk samples of SiN<sub>x</sub> at each value of  $x$  are shown in figure 3. As the value of  $x$  increases the first peak of the total RDF, which occurs at 1.80 Å,

increases systematically, having values of 2.09, 2.20, 2.22, 2.25 and 2.26 at  $x=0.8, 0.9, 1.0, 1.1$  and  $1.2$  respectively. It is clear from comparison with the partial RDFs that this peak is due to the first peak of the Si-N RDF. As  $x$  increases the first peak of the Si-Si RDF, at around  $2.5 \text{ \AA}$ , is also seen to decrease dramatically due to the fact that N atoms replace Si atoms as nearest neighbours in the amorphous network. In the N-N RDF there are no N-N nearest neighbour pairs. This is because the N-N interaction within the potential parameterization used is set to zero and is also representative of the physical situation, where N-N bonds would result in molecules of N which would not be incorporated into the a-SiN network.

FIG. 3. (Color online) Total and partial radial distribution functions of a-SiN $_x$  samples at values of  $x$  from 0.8 - 1.2. Note that individual RDFs have been offset by 0.5 on the  $y$  axis for clarity.

The RDFs obtained from our simulations are much closer in character to those obtained by Alvarez and Valladares using first principles simulations<sup>28</sup> than those obtained using the same Tersoff potential as ourselves produced by de Brito Mota *et al.*<sup>24</sup>. We believe that this is due to the fact that the methodology which we have used to produce the a-SiN $_x$  is much closer to the annealing methodology of the former than the Monte-Carlo based methodology of the latter. This gives us an increased degree of confidence in the ability of this Tersoff parameterization to simulate the properties of a-SiN $_x$ .

We now compare the the bonding patterns from our simulations with the experimentally determined values<sup>29</sup>. We note that in this case we are comparing the results of our simulations which include hydrogen to the experimental results as they are also measured in the presence of hydrogen. Figure 4 shows the total coordination numbers of Si and N in the various SiN $_x$ :H samples as  $x$  increases, we concentrate on the range  $x = 0.8 - 1.2$ , while the experimental results of Guraya *et al.*<sup>29</sup> (figure 4) cover a much wider range. In the graph of the partial coordinations the pairs A-B represent the number of atoms B bonded to atom A. In this case the atoms are said to be bonded if the interatomic separation is within the first peak of the partial RDF. These results are presented with the experimentally determined values of Guraya *et al.*, who obtained the values from x-ray photoemission spectroscopy.

Our bonding patterns match very well with the experimental data. We observe in our simulations that N and H bond almost entirely to Si, with very low quantities of N-H and N-N bonding, as observed in the experimental data. Also we observe that the number of N atoms bonding to a Si centre increases in a linear manner with the value of  $x$ . Conversely

number of Si atoms bonded to an Si centre decreases linearly with increasing  $x$ . This is an indication that the Si nearest neighbours in the network are replaced by N atoms as  $x$  increases. As in previous theoretical and experimental studies the coordination of the N atoms is largely unaffected by the varying  $x$ <sup>24,28,29</sup>.

FIG. 4. (Color inline).Upper: Total and partial coordination numbers for Si and N atoms from simulation. A-B means the number of atoms of B bonded to centre A. Lower: Partial coordination numbers from the experimental study of Guraya *et al.*<sup>29</sup>(reproduced with permission of APS).  $n_B(A)$  refers to the number of atoms of B bonded to centre A.

The above results demonstrate that using our melting and quenching regime we can generate reliable a-SiN<sub>*x*</sub>:H structures. The trends observed in bonding patterns and the RDFs are in excellent agreement with previous *ab initio* and experimental studies<sup>28,29</sup>. We now proceed to use the samples thus generated and the methodologies established in order to model the interface between a-SiN<sub>*x*</sub>:H and the c-Si(100) surface.

## B. Clean c-Si/SiN<sub>*x*</sub>:H interfaces

We first examine interfaces in which there is an abrupt change in the N concentration at the interface with c-Si, which we refer to as clean interfaces. For this analysis we consider the SiN<sub>*x*</sub>:H layer within 1 nm of the c-Si layer. In figure 5 we can see that at  $x=0.8$  the total coordination number of Si is roughly the same as that in the bulk sample, however as  $x$  increases, although the coordination of Si does increase, this increase is not as pronounced as in the bulk sample. The coordination number of N remains constant as  $x$  increases, as was observed in the bulk samples. However, the coordination number of N is slightly decreased with respect to bulk samples by  $\sim 0.1$ . Whilst these changes may seem minor, closer inspection of the bonding patterns, from the partial coordination numbers, reveals a significant difference in bonding. The number of nearest neighbour N atoms for each Si centre is decreased with respect to bulk, varying from 1.6 to 2.5 across the range of  $x$  values. N coordination is again largely accounted for by Si atoms, with a slight increase in the number of N-H bonds. The values for Si-H are very similar to those in the bulk SiN<sub>*x*</sub>:H.

We now consider the presence of bonding defects which are important in solar-cells in terms of electrical performance. The two states in particular which we consider are the K



FIG. 5. (Color online) Total and partial bond densities in  $\text{SiN}_x$  close to the clean interface with c-Si as a function of nitrogen content ( $x$ ).

and N defects which were described in the introduction. In addition passivation of these defects by H is very important, therefore we consider how these defects are passivated in our simulations. The results are presented in table II. Considering the total density of K-defects, the most striking feature is that there is a sharp increase in this number after  $x=1.1$ . It is interesting to note that this change occurs at the percolation limit of  $\text{SiN}_x$ <sup>30</sup>, at which point the properties of  $\text{SiN}_x\text{:H}$  are known to alter in a non-linear fashion with varying  $x$ <sup>31</sup>. These results match well with the observation that above the percolation limit the density of Si dangling bonds can increase<sup>30</sup>. The presence of H leads to a passivation of these dangling bonds by bonding of the H to the under-coordinated Si centre. The results in table II indicate that, in the presence of 20 atomic % H, between 30 and 40 % of the K defects at the interface are passivated in our simulations.

The density of N defects shows a similar large jump to the K defect density, however this occurs at  $x = 1.0$ , slightly before the percolation limit. We note, however, that below the percolation limit defects on N centres are unlikely to be noticeable as the band structure is dominated by Si bands and the N defect state is not in the band gap<sup>30</sup>. Interestingly the N defects are passivated to a greater degree, between 40 and 50 %, than the K centres. This is despite the fact that, from figure 5, the degree of Si-H bonding is greater than the degree of N-H bonding, indicating that H bonds more selectively to under-coordinated N than under-coordinated Si. This supports the hypothesis that the reduction in density of paramagnetic Si dangling bonds observed upon increasing annealing temperature is a result of charge transfer between paramagnetic defects, rather than the passivation of dangling Si bonds by hydrogen<sup>32</sup>.

The results show that producing a  $\text{SiN}_x\text{:H}$  layer with a stoichiometry optimized for optical performance (i.e. larger values of  $x$ ) does not necessarily lead to a layer with optimal electrical properties. This, coupled to the fact that lower values of  $x$  are desirable for the role of  $\text{SiN}_x\text{:H}$  as an ARC, means that a balance must be found to optimize optical and electrical properties. This conclusion is in line with that of Jung *et al.*<sup>15</sup>, who suggest that stoichiometric values result in more effective devices than either high or low extremes of  $x$ .

| $x$   | 0.8  | 0.9  | 1.0  | 1.1  | 1.2  |
|-------|------|------|------|------|------|
| $K^t$ | 9.3  | 18.0 | 17.9 | 29.0 | 33.6 |
| $K^p$ | 3.6  | 5.5  | 6.1  | 12.3 | 12.1 |
| $N^t$ | 17.8 | 19.3 | 27.7 | 27.8 | 32.6 |
| $N^p$ | 7.6  | 8.5  | 13.2 | 11.2 | 12.8 |

TABLE II. Defect concentrations, in  $10^{21}/\text{cm}^3$ , at the clean interfaces across the  $x$  range in  $\text{SiN}_x\text{:H}$ .  $p$  refers to passivated meaning the defect has a H atom bonded,  $t$  refers to total and is passivated plus un-passivated centres.

### C. Gradual c-Si/ $\text{SiN}_x\text{:H}$ interfaces

Thus far we have considered only clean interfaces where there is an abrupt drop in the N concentration at the c-Si surface. In reality there is often a layer of gradual depletion of the N concentration. The width of this layer varies significantly depending on the deposition method, and can also vary within a single sample. However, this layer is typically between 1-3 nm wide. In order to simulate the effect such a gradual depletion may have on the structure of the  $\text{SiN}_x\text{:H}$  and the interface we consider a number of systems which were set up as described in the methods section. We should point out that in the case of the simulations the width of the layer is restricted to around 2 nm (figure 6), which we choose as being representative of the experimental range. We are aware, however, that the slope and width of the gradient may play an important role in determining the structure in this region, the effect of the width shall be considered in future studies. For now we shall restrict ourselves to establishing the difference in structure between the clean and gradual interfaces.

| $x$   | 0.8 | 0.9  | 1.0 | 1.1  | 1.2  |
|-------|-----|------|-----|------|------|
| $K^t$ | 9.6 | 10.8 | 7.1 | 14.7 | 13.6 |
| $K^p$ | 1.7 | 1.4  | 2.0 | 2.3  | 2.8  |
| $N^t$ | 3.2 | 2.7  | 3.2 | 5.3  | 8.8  |
| $N^p$ | 1.9 | 1.2  | 1.7 | 2.2  | 3.3  |

TABLE III. Defect concentrations, in  $10^{21}/\text{cm}^3$ , at the gradual interfaces across the  $x$  range in  $\text{SiN}_x\text{:H}$ .  $p$  refers to passivated meaning the defect has a H atom bonded,  $t$  refers to total and is passivated plus un-passivated centres.

FIG. 6. (Color online) Value of  $x$  in  $\text{SiN}_x$  across the various simulated interfaces.  $r$  is the coordinate orthogonal to the interface and the c-Si layer is located at  $r=10\text{\AA}$ .

As was the case for the clean interface the total Si coordination number (figure 7) is similar to the bulk at  $x=0.8$ , and remains more or less constant across the stoichiometry range. At the gradual interface the coordination number of N is very similar to that in the bulk samples, and again is due to three Si atoms coordinated to the N centre (figure 7). The number of N atoms bonded to each Si centre is decreased with respect to the bulk and clean interface values, obviously due to a lower N content near the interface with c-Si in the gradual interface system. The hydrogen coordination numbers to both Si and N are reduced with respect to both the bulk and the clean interface, suggesting that the degree of passivation of dangling bonds may be reduced in a gradual interface. We now consider this specifically relating to the K and N defects.

FIG. 7. (Color online) Total and partial bond densities in  $\text{SiN}_x$  close to the interface with c-Si as a function of nitrogen content ( $x$ ) in systems with a N gradient.

Table III presents the density of K and N defects found in the gradual interface systems. The total density of K-defects at  $x=0.8$  is slightly greater in the gradual interface than the clean interface. However, as we go to larger values of  $x$  the total density does not increase as rapidly as in the clean interface systems. Thus, for all values of  $x > 0.8$  the total density of K-defects is reduced in the gradual interface. As suggested by the coordination numbers, the degree of passivation of K-centres is significantly reduced in the gradual interface ranging

between 15 to 30 %. However, despite the reduction in the degree of passivation, the density of un-passivated K-defects is still lower in the gradual interfaces compared to the clean ones for all values of  $x > 0.8$ . The total density of N defects on the other hand is greatly reduced across the composition range. This is because the density of N defects within 1 nm of the c-Si is reduced in the gradual interface and the coordination numbers suggested that N centres were more fully bonded in the gradual than in the clean interface. This indicates that, in addition to improved optical performance, the gradual interface can be expected to have improved electrical performance to the clean interface.

Moreover, the low values of  $x$  close to the c-Si layer would result in improved ARC properties. Thus our simulations provide clear evidence that a N gradient across the boundary from  $\text{SiN}_x$  to c-Si improves the optical properties of the layer by both reducing absorption and increasing light trapping, while also improving electrical properties by reducing the number of carrier traps at the interface.

#### IV. CONCLUSIONS

In this paper we have demonstrated that using the Tersoff potential the structure of a- $\text{SiN}_x$  can be simulated in excellent agreement with *ab initio* and experimental structures. We demonstrate that using a methodology based on melting the sample and quenching it a better representation of the true amorphous structure can be obtained compared to previous studies which used Monte-Carlo methods to generate a- $\text{SiN}_x$  using the Tersoff potential<sup>24</sup>.

We have then used this methodology to simulate  $\text{SiN}_x$  interfaces with the c-Si(100) surface. We have simulated this as both a clean interface and with a gradient of N concentration across the boundary. To our knowledge this is the first systematic theoretical examination of the effect of  $\text{SiN}_x$  stoichiometry on the structure of the  $\text{SiN}_x$ /c-Si interface, as well as the first time a system with a N gradient across the interface has been modelled atomistically. The results of the simulations have implications for future optimization of the  $\text{SiN}_x$  layer in high performance c-Si solar cells.

The simulations reveal that in the case of a clean interface higher values of  $x$  in  $\text{SiN}_x$  would be expected to improve optical properties of the ARC, but may adversely affect the electrical properties due to increased densities of both K- and N defects at the interface, although these centres are to some extent passivated by the presence of H. The increased

density of K-defects is particularly pronounced at  $\text{SiN}_x$  compositions with  $x \geq 1.1$ , which is also the percolation limit of  $\text{SiN}_x$ . These findings agree previous work which showed that Si defects can increase in density above the percolation limit<sup>30</sup>. This finding also agrees with recent experimental work which found that stoichiometric values of  $x = 1$  offer the best performance, compared to extreme values<sup>15</sup>. Finally we have shown that a gradient interface results in fewer electron traps, namely the N and K centre defects in the interface region.

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