



Hydrogen Reintroduction by Forming Gas Annealing to LPCVD Silicon Nitride Coated Structures

Hao Jin,^z K. J. Weber, P. N. K. Deenapanray, and A. W. Blakers

Centre for Sustainable Energy Systems, Faculty of Engineering and Information Technology,
The Australian National University, Canberra ACT 0200, Australia

The effect of high-temperature anneals on low-pressure chemical-vapor deposited (LPCVD) Si₃N₄/thermally grown SiO₂ stacks on silicon wafers has been studied. Annealing in a nitrogen atmosphere at 900 or 1000°C for 30 min causes a substantial reduction in both N–H and Si–H bond concentration in the nitride film and a significant degradation of surface passivation. Subsequent anneals in forming gas (95% Ar/5% H₂) at temperatures in the range 800–900°C lead to a gradual reformation of N–H and Si–H bonds and an improvement in surface passivation. It is hypothesized that the chief mechanism responsible for the improvement in surface passivation is the passivation of defects at the Si–SiO₂ interface with atomic hydrogen, supplied by the breaking of N–H and Si–H bonds in the nitride film. Samples with SiO₂/Si₃N₄ stacks display excellent surface passivation following high-temperature forming gas anneals, indicating that such treatments can be used for photovoltaic applications, where surface passivation as well as good optical properties are important.

© 2006 The Electrochemical Society. [DOI: 10.1149/1.2207058] All rights reserved.

Manuscript submitted December 2, 2005; revised manuscript received April 5, 2006. Available electronically June 13, 2006.

Over the last few years, silicon nitride (SiN_x) deposited by plasma-enhanced chemical vapor deposition (PECVD) has become the material of choice as an antireflection coating for silicon solar cells. This can be attributed to the combination of desirable properties that this material possesses: a refractive index that can be tuned to provide an excellent antireflection coating, the ability to achieve very good surface passivation, and the fact that it is possible to considerably improve the bulk of low-quality silicon material during high-temperature thermal steps following deposition of a PECVD nitride layer. This last property results from the diffusion of hydrogen in the nitride film into the silicon bulk, where it passivates defects.^{1,2}

In comparison, low-pressure chemical vapor deposited (LPCVD) silicon nitride, while widely used in the microelectronics field, has not received much attention for photovoltaic applications. LPCVD-deposited silicon nitride displays very different properties to PECVD SiN_x. If deposition is carried out using an excess of NH₃ (the situation we consider here), the composition of the nitride is close to stoichiometric and hence may be written as Si₃N₄. When deposited directly on silicon, it provides only very poor surface passivation. Further, the hydrogen content of LPCVD Si₃N₄ is much lower (2–10 atom %) than that of PECVD SiN_x (20–25 atom %),^{3,4} and LPCVD Si₃N₄ has not been successfully used for the passivation of bulk defects. In fact, it is known that the bulk properties of wafers with LPCVD Si₃N₄ films can degrade substantially and irreversibly if subjected to high-temperature processes.^{5–7}

Despite these disadvantages, LPCVD Si₃N₄ has several properties which make it potentially very useful for silicon solar cells. It is an excellent diffusion and oxidation mask,⁸ is very hard and therefore scratch-resistant,⁹ is resistant to attack by a large range of etchants commonly used for solar cell fabrication, such as alkaline solutions and silicon etchants containing hydrofluoric and nitric acid, and has a refractive index which is near optimal as an antireflection coating behind glass.⁹ Further, LPCVD Si₃N₄ deposition is highly conformal, such that even obscured surfaces can be coated.¹⁰ This is of particular advantage in novel thin silicon cell structures such as SLIVER cells.¹¹ Finally, high-volume batch deposition processes such as LPCVD are potentially cheaper than PECVD.

A prerequisite to the use of LPCVD Si₃N₄ as an antireflection coating in high-performance solar cells is the ability to achieve excellent surface passivation. In order to take full advantage of its properties, it would be highly desirable to be able to carry out further high-temperature steps following nitride deposition, without significantly degrading either the bulk or surface properties of the

substrate. It has already been demonstrated that the use of a thin (~25 nm) thermally grown oxide between the wafer and the nitride film allows excellent surface passivation to be achieved.^{5,12} The presence of the oxide will result in only a marginal decrease in the antireflection properties of the optimized oxide/nitride stack. Further, the use of a thin interfacial oxide has also been shown to significantly reduce the degradation of the silicon bulk following further high-temperature process steps.⁵

A key challenge to the use of LPCVD Si₃N₄ as both a masking material during processing and a final antireflection coating is the ability to maintain good surface passivation at the end of the process. Studies have shown that degradation of surface passivation during high-temperature processing is primarily caused by a loss of hydrogen from the Si–SiO₂ interface, where the hydrogen passivates interfacial defects.^{5–7,12} It has further been shown that subsequent high-temperature anneals in a hydrogen atmosphere can lead to substantial improvement in surface passivation, as a result of the diffusion of hydrogen to the interface and the repassivation of defects. An optimization of this process requires, among other things, an understanding of the diffusion of hydrogen through the nitride film and the influence of the various process parameters on the diffusion of hydrogen.

The diffusion of hydrogen in LPCVD silicon nitride has been studied by several authors.^{13–17} Perhaps the most comprehensive and detailed study is that of Arnoldbik et al.¹⁷ who, using elastic recoil detection (ERD) analysis, determined the diffusion coefficient of hydrogen in LPCVD nitride and oxynitride films over a range of temperatures.

In this paper, the effect of various high-temperature anneals (in the range 800–1000°C) in either a nitrogen or dilute hydrogen atmosphere, on the hydrogen concentration of LPCVD Si₃N₄ films, as well as on the surface passivation properties of samples with SiO₂/Si₃N₄ stacks, is studied in detail. The high-temperature nitrogen anneals are used to simulate high-temperature process steps such as diffusions and oxidations that may be carried out subsequent to nitride deposition in a solar cell process sequence, while the anneals in a dilute hydrogen atmosphere are used to reintroduce hydrogen into the nitride and the Si–SiO₂ interface.

Experimental

Sample preparation.— For the photoconductivity decay measurements, float zone (FZ) (100)- and (111)-oriented, high-resistivity (≥100 Ω cm), 500 μm thick wafers were initially etched in 10:1 HNO₃:HF solution until shiny. Both sides of the wafers first received a light phosphorous diffusion (with a sheet resistance R_{sh} of ~400 Ω/□ after thermal drive in), followed by the growth of an 80 nm thick thermal oxide, a 30 min in situ anneal at 1000°C in

^z E-mail: hao.jin@anu.edu.au

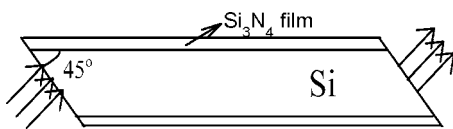


Figure 1. The structure of the samples for MIR measurements.

nitrogen, and a 30 min anneal at 400°C in forming gas (5% H₂/95% Ar). This final anneal is hereafter referred to as a low-temperature forming gas anneal (LTFGA). Selected wafers then had 50 nm thick LPCVD silicon nitride deposited on both sides. All depositions were done at 775°C, a pressure of 0.5 Torr, and ammonia to dichlorosilane (DCS) gas flow ratio of 5 to 1. The forming gas mixture used throughout (5% H₂ in Ar) was chosen because this mixture is not explosive, and hence, processing with this composition can be carried out using inexpensive processing equipment, an important consideration for potential industrial processes.

For the infrared multiple internal reflection (MIR) measurements, 500 μm thick, double-side polished wafers were used. A 200 nm thick layer of LPCVD silicon nitride was deposited on both sides of the wafers. Such a relatively thick layer was necessary in order to achieve a good signal-to-noise ratio, and ensure good accuracy of the results.

Samples for both photoconductivity decay and MIR measurements were then annealed in nitrogen at 900 or 1000°C for 30 min in order to remove hydrogen from the silicon nitride films as well as the Si–SiO₂ interface in the case of the samples for photoconductivity decay measurements. Samples were then annealed in forming gas at various temperatures (in the range 800–900°C) and for different times in order to reintroduce hydrogen into the nitride and to the Si–SiO₂ interface.

Photoconductivity decay measurements.—Minority carrier lifetime measurements were made using the quasi-steady-state photoconductivity decay (QSS-PCD) technique. The effective carrier lifetime was extracted using the equations of Sinton and Cuevas¹⁸ and Nagel et al.¹⁹ In high injection, the instantaneous decay time τ_{inst} is²⁰

$$1/\tau_{\text{inst}} = 1/\tau_{\text{hi}} + (2J_{\text{oe}}n)/(qWn_i^2) \quad [1]$$

where τ_{hi} is the high-level injection lifetime of the wafer bulk, n is the photogenerated excess carrier density in the wafer bulk, W is the wafer thickness, n_i is the intrinsic carrier concentration, J_{oe} is the emitter dark saturation current density, and q the electronic charge. J_{oe} is an indicator of the degree of surface passivation and a parameter of direct technological importance for photovoltaics, because high values of J_{oe} can limit achievable cell efficiencies by imposing limits on some or all of the key cell parameters, namely, the open-circuit voltage, fill factor, and short-circuit current. The value of J_{oe} can be determined from the slope of $1/\tau_{\text{eff}}$ vs n after Auger recombination has been accounted for.

FTIR measurements.—Fourier transform infrared spectroscopy (FTIR) was used to determine the concentration of Si–H and N–H bonds in the Si₃N₄ film. The MIR method was used to enhance the sensitivity of the measurement.^{14,21} The N–H bond and Si–H bond stretching modes at wavenumbers of 2202 and 3331 cm⁻¹, respectively, were detected.¹⁵ Because the penetration depth is 349 and 535 nm for N–H and Si–H, respectively, all hydrogen-related bonds through the LPCVD silicon nitride layer could be detected.¹⁴ The wafer was cut into 8 × 10 × 0.5 mm pieces. A 45° angle was polished at both edges of each piece. Light entering the samples is reflected about 20 times at the Si/Si₃N₄ interface before exiting, as shown in Fig. 1. Both bare silicon samples and Si/Si₃N₄ stacks were measured by MIR, and the spectra from bare Si were subtracted in order to obtain data for the Si₃N₄ layer only. The concentration of

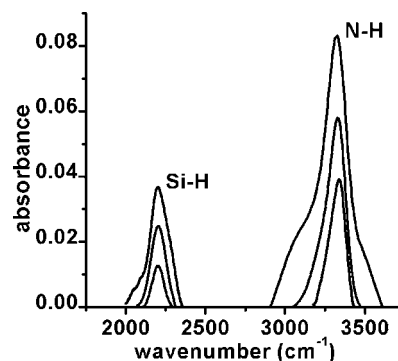


Figure 2. Absorbance spectra of the LPCVD nitride films showing N–H bonds (near 3331 cm⁻¹) and Si–H bonds (2202 cm⁻¹). The spectra are, in the order from highest to lowest peaks: as deposited; 30 min N₂ anneal at 900°C and a subsequent FGA for 1 h at 840°C; and 30 min N₂ anneal at 900°C only.

Si–H and N–H bonds was then determined by integration of the stretching mode absorbance peaks, following the method of Lanford and Rand.²²

Results and Discussion

Hydrogen concentration in the Si₃N₄ layer following annealing at 900°C.—Forming gas annealing (FGA) was carried out for 1 h at different temperatures following an initial anneal in nitrogen for 30 min at 900°C. The absorbance spectra for some of the films measured are shown in Fig. 2. After annealing in nitrogen at 900°C for 30 min, both Si–H and N–H bond concentrations decrease dramatically, while a subsequent FGA significantly increases these concentrations, indicating reintroduction of hydrogen into the nitride film.

Figure 3 summarizes the results of 1 h isochronal FGAs carried out in the temperature range 800–900°C following a 30 min nitrogen anneal at 900°C. For reference, the total bonded hydrogen concentration immediately after the N₂ anneal is 1.1%. It can be seen that for isochronal anneals, there exists an optimum temperature which leads to the greatest density of hydrogen bonds. In the case of the 1 h anneals performed here, this optimum temperature is close to 840°C. This implies that the equilibrium hydrogen concentration in the nitride film in a forming gas ambient decreases at high temperatures. This finding is consistent with the results of Arnoldbik and Maree,¹⁷ who inferred from their measurements a decrease in the

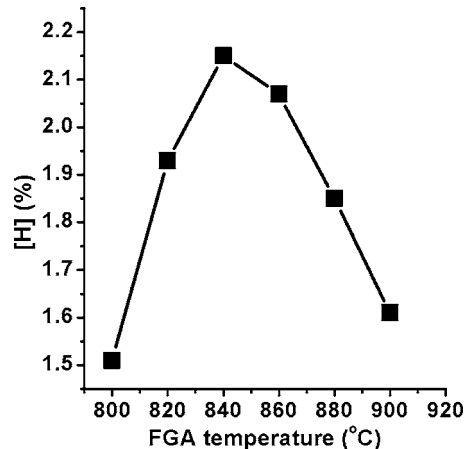


Figure 3. Total bonded hydrogen concentrations in the LPCVD nitride films following a 30 min N₂ anneal at 900°C and a 1 h isochronal FGA at temperatures in the range 800–900°C.

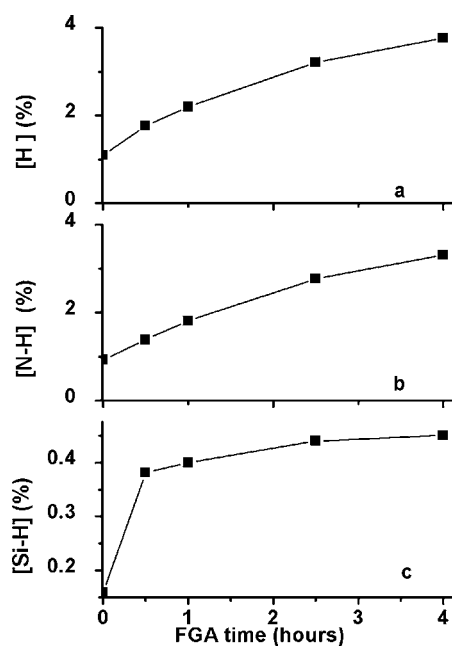


Figure 4. Bonded hydrogen concentrations in LPCVD nitride films following a 30 min N_2 anneal at $900^\circ C$ and an isothermal FGA at $840^\circ C$: (a) total hydrogen concentration; (b) N–H bond concentration; and (c) Si–H bond concentration.

surface concentration of bonded hydrogen in their nitride films following a $1000^\circ C$ anneal in 4% H_2 , compared to an anneal at 900 or $800^\circ C$.

Figure 4 shows the results of isothermal FGAs at $840^\circ C$ on samples that had previously been annealed in nitrogen at $900^\circ C$ for 30 min. The Si–H concentration increases rapidly during the first 30 min, after which the increase in concentration is only marginal. In contrast, there is a relatively steady increase in the N–H bond concentration as the anneal time is increased from 0 to 4 h. The Si–H bond concentration is always much lower than the N–H concentration. As a result, a steady increase in the total bonded hydrogen concentration is observed.

Hydrogen concentration in the Si_3N_4 layer following annealing at $1000^\circ C$.— Samples were also annealed at $1000^\circ C$ in nitrogen for 30 min, followed by 1 h of isochronal FGAs at various temperatures in the range 800 – $900^\circ C$. Measurements of the bonded hydrogen concentrations again indicated an optimum anneal temperature of around $840^\circ C$. In subsequent experiments, therefore, samples were given isothermal FGAs at $840^\circ C$ after the $1000^\circ C$ N_2 anneal. Figure 5 summarizes these results. The higher anneal temperature has resulted in lower initial hydrogen bond concentrations, as well as a somewhat lower rate of reforming of Si–H and N–H bonds. Compared with Fig. 4, the hydrogen content for the post- $1000^\circ C$ nitrogen anneal structure following a 4 h FGA is only around half that of the post- $900^\circ C$ nitrogen anneal structure. It is clear that the increase in anneal temperature has resulted in a significant change in the properties of the silicon nitride layer.

The diffusion of hydrogen in LPCVD S_3N_4 films has been studied in detail by Arnolbik et al.^{16,17} using similar but not identical nitride film deposition conditions to those used here. The loss of hydrogen from our samples is substantially more rapid than that observed by Arnolbik et al.,¹⁷ who found a decrease in the total amount of bonded hydrogen to about 30% of its initial value after a 30 min, $1000^\circ C$ anneal, for nitride films of similar thickness to those used here. In contrast, we find a reduction to $\sim 10\%$ of the initial concentration for the same anneal conditions. The difference may be a result of the different deposition conditions, such as the

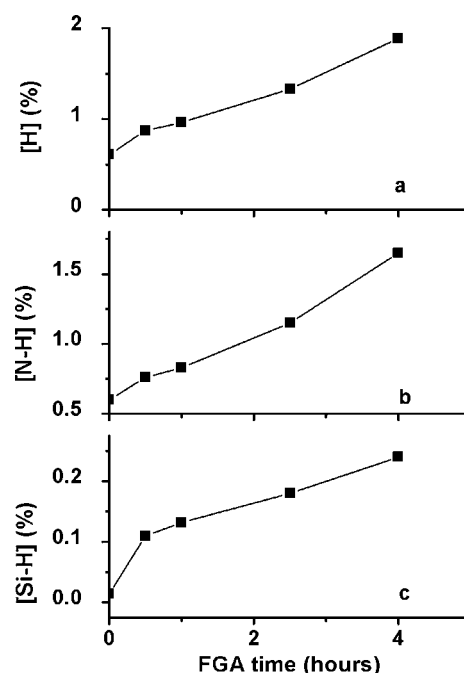


Figure 5. Bonded hydrogen concentrations in LPCVD nitride films following a 30 min N_2 anneal at $1000^\circ C$ and an isothermal FGA at $840^\circ C$: (a) total hydrogen concentration; (b) N–H bond concentration; and (c) Si–H bond concentration.

slightly lower deposition temperature used here ($775^\circ C$ compared to $800^\circ C$). Arnolbik et al. explained the process of hydrogen reintroduction into Si_3N_4 films annealed in molecular hydrogen at high temperatures as a two-step reaction. In the first step, molecular hydrogen diffuses into a surface layer of Si_3N_4 to form N–H and Si–H bonds. The subsequent step is the diffusion of the bonded hydrogen from the surface deeper into the film, with this step requiring first the breaking of hydrogen bonds and then the diffusion of atomic hydrogen into the film. It is likely that the same mechanism is responsible for the observations reported here.

Effect of anneals on the dark saturation current density J_{oe} .—

Table I shows the change in dark saturation current density as silicon samples are taken through various processing steps. For these experiments the thicknesses of the oxide and nitride films were 80 and 50 nm, respectively. Si(100) and (111) surfaces were compared. Three samples were used for the average values after each step shown in Table I. Just after nitride deposition, the surfaces are well-passivated with a low value of J_{oe} . The nitrogen anneal results in a

Table I. Emitter saturation current (J_{oe}) values ($fA/cm^2/side$) of Si(100) and (111) wafers measured after various process steps: initial oxidation and LTFGA; subsequent nitride deposition and N_2 anneal at $900^\circ C$ for 30 min; subsequent FGA at $840^\circ C$ for 1 h; and a subsequent further FGA at $840^\circ C$ for 3 h. Every value is the average of measurements on three samples. Values measured on individual samples were within 10% of the average value.

Treatment	Wafer orientation	
	(100)	(111)
As oxidized	7	20
Nitride + N_2 anneal	70	1700
1 hour FGA	13	65
3 hour FGA	8	25

substantial increase in J_{oc} , chiefly as a result of a loss of hydrogen from the Si–SiO₂ interface where the hydrogen passivates interface defects. The subsequent high-temperature, 840°C FGAs result in a gradual repassivation of the interface, indicating that hydrogen has diffused through the nitride and oxide films and has bonded with interfacial defects, despite the relatively high anneal temperature. Diffusion of hydrogen through the oxide layer at such a high temperature will be extremely rapid, so that the presence of the oxide does not present a significant additional barrier to the diffusion of hydrogen to the silicon surface.

Both (100) and (111) samples follow a very similar trend. However, (111) surfaces exhibit higher J_{oc} values at each stage and a much faster depassivation rate²³ during the N₂ anneal, which is probably due to a higher concentration of (electrically active) defects at (111) compared to (100) Si–SiO₂ interfaces prepared under similar conditions. The results of Stesmans²⁴ indicate postoxidation vacuum anneals above 640°C result in about 2 times more paramagnetic defects at (111) than (100) Si–SiO₂ interfaces.

A plausible explanation of the results of the FTIR and photoconductivity decay measurements is that the hydrogen for repassivation of the interfacial defects is supplied chiefly by atomic hydrogen created by the breaking of Si–H and N–H bonds. A qualitative description for the process is as follows. The rate of generation of atomic hydrogen would be expected to increase with increasing concentration of bonded hydrogen. Once atomic hydrogen is liberated, it diffuses with a certain mean path length prior to rebonding in the nitride film. Atomic hydrogen in the vicinity of the SiO₂ layer may also diffuse across the SiO₂ layer and passivate defects at the Si–SiO₂ interface. Thus, the degree of interface passivation would be expected to improve with increasing concentration of bonded hydrogen in the nitride film.

The conclusion that the bonded hydrogen in the nitride film can play a key role in Si–SiO₂ interface passivation is consistent with the results of McCann et al.,⁷ who showed that the degradation in J_{oc} of Si/SiO₂/LPCVD Si₃N₄ stacks during high-temperature nitrogen anneals decreased substantially as the nitride layer thickness was increased, due to an increased supply of atomic hydrogen from hydrogen bonds in the nitride layer.

The fact that Si–SiO₂ interfaces can effectively be repassivated at a temperature as high as 840°C is remarkable. Stesmans,²⁵ studying the depassivation of the dominant P_b center defect on Si(111), showed that this defect is virtually entirely depassivated after a 1 h vacuum anneal at a temperature of only 650°C. Calculations using the model developed by Stesmans also indicate that a 1 h anneal in pure H₂ at 840°C may still result in the passivation of around 99.9% of all P_b centers for a good quality Si–SiO₂ interface²⁶ and in the presence of only an oxide layer, which allows rapid diffusion of hydrogen to the interface. However, in the case studied here the supply of hydrogen to the interface is likely to be orders of magnitude lower so that the same efficiency of passivation would not be expected. It is possible that the passivation process is more accurately described as a two-step process, where the first step is the rehydrogenation of the nitride film at a high temperature, and the second step, namely, the effective passivation of interface defects, may occur at intermediate temperatures as the sample cools to room temperature. If the latter description is more appropriate, one may expect the rate of cooling of the samples to have a significant effect on the measured value of J_{oc} . However, experiments with different cooling rates (one, a rapid cool in which the samples were removed and cooled to <100°C within 10 s and the other, a slow cool in which the samples were ramped down to <100°C in 10 min) have not revealed significant differences in J_{oc} .

A further interesting observation relates to hydrogen-induced defect creation at the Si–SiO₂ interface. Stesmans showed that extended high-temperature (>700°C) hydrogen anneals lead to a marked generation of P_b centers on (111) surfaces,²⁷ which is detected after vacuum annealing by EPR. At a temperature of 800°C, hydrogen annealing led to a sixfold increase in P_b center density.

Table II. Emitter saturation current (J_{oc}) values (fA/cm²/side) of Si(100) samples measured after various process steps: initial oxidation and LTFGA (labeled “As oxidized”); 1 h, 840°C HTFGA; a subsequent LTFGA; and reoxidation and LTFGA (“Re-oxidized”). Every value is the average of measurements on three samples. Values measured on individual samples were within 10% of the average value.

As oxidized	HTFGA	LTFGA	Reoxidized
7.2	32	21	7.8

This increase in defect density is also hinted at by an increase in the value of J_{oc} . Table II compares the values of J_{oc} for Si/SiO₂ stacks at various stages of processing. These samples were first oxidized and passivated with an LTFGA. This was followed by a 30 min 900°C nitrogen anneal and a 1 h 840°C, high-temperature FGA. The stacks then received a second LTFGA. Finally, the oxide was stripped and regrown, and the stacks were again repassivated using another LTFGA. Comparing the results with those of Table I, the value of J_{oc} can be seen to be higher for the Si/SiO₂ stacks just after the high-temperature FGA. The fact that the subsequent low-temperature FGA did not reduce J_{oc} to near its initial value can be attributed to the generation of additional defects during the high-temperature FGA, resulting in a greater density of residual (unpassivated) defects following the LTFGA. Only reoxidation reduces the defect density, and the value of J_{oc} , back to near its initial value, in agreement with the results of EPR measurements.

Thus, it is clear that great care needs to be exercised when employing high-temperature hydrogen anneals. There are two possible explanations for the better passivation observed following high-temperature FGAs when a nitride layer is present. First, the amount of hydrogen reaching the interface in the presence of a nitride layer is much less than in the case where only an oxide is present, so that the rate of generation of additional defects may be much lower. Second, it is possible that the atomic hydrogen supplied by the nitride layer provides much more efficient interface passivation than the molecular hydrogen for the case where only an oxide is present. It is well known that much lower values of J_{oc} can be achieved following the FGA of oxides if a thin layer of aluminum is deposited on the oxide prior to annealing.²⁸ Such “anneals” are believed to be more efficient at interface passivation as a result of the generation of atomic hydrogen at the aluminium–oxide interface.

Conclusion

In conclusion, FGA at temperatures in the range 800–900°C has been demonstrated to cause a gradual reformation of hydrogen bonds in Si₃N₄ films which had previously been subjected to high-temperature treatments, causing a loss of hydrogen. In parallel with an increase in the hydrogen content in the nitride film, an improvement in surface passivation of samples with oxide/nitride stacks, as indicated by a decrease in the value of the emitter saturation current J_{oc} , is observed. This is ascribed to a repassivation of interfacial defects with (atomic) hydrogen. Remarkably, such high-temperature FGAs can lead to very effective interface passivation on both (111)- and (100)-oriented surfaces, and to lower values of J_{oc} than are achieved (for identical treatments) on surfaces with a thermally grown oxide layer only.

Acknowledgments

The authors thank Dr. Lan Fu and Dr. Weitang Li at the Research School of Physics for their help with MIR measurements. Financial support for this project by the Australian Research Council (DP0557398) is gratefully acknowledged.

Australian National University assisted in meeting the publication costs of this article.

References

1. Z. Chen, A. Rohatgi, R. O. Bell, and J. P. Kalejs, *Appl. Phys. Lett.*, **65**, 2078 (1994).
2. F. Duerinckx and J. Szlufcik, *Sol. Energy Mater. Sol. Cells*, **72**, 231 (2002).
3. F. H. P. M. Habraken, R. H. G. Tijhaar, W. F. van der Weg, A. E. T. Kuiper, and M. F. C. Willemsen, *J. Appl. Phys.*, **59**, 447 (1986).
4. H. J. Stein, P. S. Peercy, and R. J. Sokel, *Thin Solid Films*, **101**, 291 (1983).
5. M. J. McCann, K. J. Weber, and A. W. Blakers, *Prog. Photovoltaics*, **13**, 195 (2005).
6. M. J. McCann, K. J. Weber, and A. W. Blakers, in *Proceedings of the 3rd World Conference on Photovoltaic Solar Energy Conversion*, p. 771 (2003).
7. M. J. McCann, M. J. Stocks, K. J. Weber, and A. W. Blakers, in *Proceedings of the 17th European Photovoltaic Solar Energy Conference*, p. 1708 (2001).
8. B. Bazin, *Solid-State Electron.*, **15**, 649 (1972).
9. K. E. Bean, P. S. Gleim, R. L. Yeakley, and W. R. Runyan, *J. Electrochem. Soc.*, **114**, 733 (1967).
10. D. Teasdale, Y. Senzaki, R. Herring, G. Hoeye, L. Page, and P. Schubert, *Electrochem. Solid-State Lett.*, **4**, 11 (2001).
11. K. J. Weber, A. W. Blakers, M. J. Stocks, J. H. Babaei, V. A. Everett, A. J. Neundorff, and P. J. Verlinden, *IEEE Electron Device Lett.*, **25**, 37 (2004).
12. H. Jin, K. J. Weber, and A. W. Blakers, in *Proceedings of the 17th European Photovoltaic Solar Energy Conference*, p. 230 (2005).
13. J. Z. Xie, S. P. Murarka, X. S. Guo, and W. A. Lanford, *J. Vac. Sci. Technol. B*, **7**, 150 (1988).
14. H. J. Stein, S. T. Picraux, and P. H. Holloway, *IEEE Trans. Electron Devices*, **ED-25**, 1008 (1978).
15. J. G. E. Gardeniers, H. A. C. Tilmans, and C. C. G. Visser, *J. Vac. Sci. Technol. A*, **14**, 2879 (1996).
16. W. M. Arnoldbik, R. N. H. Linssen, F. H. P. M. Habraken, W. F. van der Weg, and A. E. T. Kuiper, *Appl. Phys. Lett.*, **56**, 2530 (1990).
17. W. M. Arnoldbik, C. H. M. Maree, A. J. H. Maas, M. J. van den Boogard, and F. H. P. M. Habraken, and A. E. Kuiper, *Phys. Rev. B*, **48**, 5444 (1993).
18. R. A. Sinton and A. Cuevas, *Appl. Phys. Lett.*, **69**, 2510 (1996).
19. H. Nagel, C. Berge, and A. G. Aberle, *J. Appl. Phys.*, **86**, 6218 (1999).
20. D. E. Kane and R. M. Swanson, in *The Eighteenth IEEE Photovoltaic Specialists Conference*, p. 578 (1985).
21. N. J. Harrick, *Internal Reflection Spectroscopy*, Interscience, New York (1967).
22. W. A. Lanford and M. J. Rand, *J. Appl. Phys.*, **49**, 2473 (1978).
23. H. Jin, K. J. Weber, and A. W. Blakers, in *The 15th International Photovoltaic Science and Engineering Conference*, p. 30 (2005).
24. A. Stesmans and V. V. Afanas'ev, *J. Vac. Sci. Technol. B*, **16**, 3108 (1998).
25. A. Stesmans, *J. Appl. Phys.*, **92**, 1317 (2002).
26. A. Stesmans, *J. Appl. Phys.*, **88**, 489 (2000).
27. A. Stesmans and V. V. Afanas'ev, *Appl. Phys. Lett.*, **72**, 2271 (1998).
28. M. A. Green, *Silicon Solar Cells: Advanced Principles and Practice*, Center for Photovoltaic Devices and Systems, Sydney, NSW, Australia (1995).