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SURFACE PASSIVATION AND JUNCTION FORMATION USING LOW ENERGY HYDROGEN IMPLANTS

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I. Introduction

The use of high current, low energy ion beams in device fabrication is a relatively recent development. Outstanding examples of this new, emerging use of low energy ions are the passivation of grain boundaries in polycrystalline silicon with low energy, hydrogen ion implantation [1] and the solar cell performance improvement resulting from low energy hydrogen ion implantation into Mobil Solar ribbon [2]. Recently we have demonstrated that there are additional, innovative applications for low energy hydrogen ion implants which can be very useful for crystalline silicon. We discuss these new applications for high current, low energy hydrogen ion implants in this paper.

The paper is arranged as follows: first, the effects of low energy hydrogen ion beams on crystalline Si surfaces are considered; next, the effect of these beams on bulk defects in crystalline Si is addressed; and, finally, specific applications of H^+ implants to crystalline Si processing are discussed. In all of the situations reported on here, the hydrogen beams were produced using a high current Kaufman ion source.

II. Hydrogen Beams: A Cause and Cure for Surface Defects

When a low energy hydrogen ion beam impinges on a single crystal silicon surface, the surface is sputter etched, hydrogen is implanted, and the silicon lattice is damaged. That low energy hydrogen ion beams damage Si surfaces may seem surprising in view of the ability of such beams to passivate grain boundaries in polycrystalline Si. However, the presence of this damage is demonstrated in Fig. 1. The figure shows Rutherford backscattering (RBS) data for two silicon samples: one which was subjected to a 0.4 keV H^+ beam and another which was subjected to a 1.0 keV H^+ beam. The two traces for these cases are RBS channeling data; the third trace is the random backscattering yield from a silicon sample.

As may be seen from the figure the lattice damage from to 1.0 keV H^+ beam exists deeper down into the Si than does the lattice damage from the 0.4 keV H^+ beam. Converting the energy scale into a measure of the field of energy dissipation for the H^+ ions gives a 200 Å deep layer of lattice damage for 0.4 keV H^+ and a 400 Å deep layer for 1.0 keV H^+ ions. This measure of the lattice damage is based on the full width in energy of the RBS channeling peak at half maximum of the yield. By noting that the channeling peak yields for both 0.4 H^+ and 1.0 H^+ beam exposures coincide with the random yield, it can be deduced that this layer of lattice damage, resulting from exposure to the hydrogen beams (dosage $\sim 10^{18}$ cm^{-2} in both cases), is amorphized.

A further demonstration of the fact that low energy hydrogen ion beams themselves damage Si is provided by Table I. This table gives peak to peak heights for the electron spin resonance (ESR) signature of silicon dangling bonds [3]. The table shows how this ESR measure of lattice damage varies as the H^+ ion beam energy is varied from 0.2 keV to 1.0 keV. The data indicate that lattice damage increases up to $\sim$0.8 keV and then appears to saturate. This saturation effect may be due to sputtering which reduces the volume of damaged material or to an annealing effect. Again all the dosages were of the order of $\sim 10^{18}$ cm^{-2}.

TABLE I

Peak-to-Peak ESR Signal Heights

Ion Beam Species & Energy		ESR Signal Heights		
		Minimum Power Microwave	Medium Power Microwave	Maximum Power Microwave
Control Sample		<1	<1	--
H^+	0.2 keV	<1	~ 1	~ 2
	0.4 keV	<1	~ 2	~ 3
	0.6 keV	<1	~ 2	~ 3.5
	0.8 keV	~ 2	~ 4.5	~ 8.5
	1.0 keV	~ 1.5	~ 4	~ 7
Ar^+	0.2 keV	<1	~ 2	~ 4
	0.4 keV	<1	~ 2.5	~ 7.5
	1.0 keV	~ 5.5	~ 15	~ 40

An electrical assessment of the damage caused at crystalline Si surfaces by low energy hydrogen ion beams is presented in Fig. 2. Here current-voltage (I-V) data are given for Au dots deposited on p-Si. As expected Au contacts to chemically prepared p-Si (the control) show a low barrier height. However, Au contacts to chemically prepared p-Si, which was subsequently subjected to various H^+ beam exposures, yield I-V characteristics which indicate the presence of positive charge. That is, positive charge is created in a layer near the Si surface due to the H^+ beam exposure and this causes the increased barrier height seen in Fig. 2 [3-6].

Thus the RBS data of Fig. 1, the ESR data of Table I, and the I-V data of Fig. 2 establish that low energy H^+ ion beams damage single crystal Si surfaces. All measures of this damage agree that it is worse at the higher energies (for the range used) and least at the lower energies. However, these measures do not all yield data that vary with energy in exactly the same manner. For example, the amount of positive charge present, as judged from the I-V data, is not found to be simply proportional to the strength of the ESR signal.

Other low energy ion beams also damage single crystal silicon surfaces. In fact, as we will see, they inflict more damage (holding dosage and beam energy constant) than does hydrogen. We begin this consideration of the damage

effects of other ion beams by turning to Fig. 3. This figure establishes that low energy Ar^+ beams (again a total of $\sim 10^{18}$ ions/cm^2 impinged on these samples) damage the silicon lattice at and below the surface. In this case the RBS data indicate that the field over which energy is dissipated by 0.4 keV Ar^+ ions extends ∿100 Å below the surface whereas the field over which energy is dissipated by 1.0 keV Ar^+ ions extends ∿200 Å. These RBS data point out that the Ar^+ beams, unlike the hydrogen, have not been able to amorphise the Si surface layer.

Table I also contains the ESR assessment of the damage caused by Ar^+ ion beams. As may be noted from the table, the damage increases with beam energy and the damage signal is higher for Ar^+, for the same dosage and energy, than it is for H^+. Fig. 4 shows the I-V data for a Au dot on p-Si which was chemically prepared and subsequently subjected to a 1.0 keV Ar^+ beam. The positively charged surface layer (i.e., the surface layer containing damage-induced donors) is clearly present as a result of the Ar^+ exposure as is apparent from the I-V characteristic.

Hydrogen ion beams cause surface damage and other ion beams such as would be present in plasma deposition, sputtering, or dry etching (we took Ar^+ as an example), also cause surface damage. This observation immediately causes several questions to present themselves: (1) how does the damage caused by H^+ compare with that caused by Ar^+, (2) how can H^+ implants passivate poly Si and Mobil Solar ribbon Si without causing surface damage, and (3) can H^+ implants passivate the damage caused by Ar^+ implants. We begin addressing these questions by comparing H^+ and Ar^+ ion beam damage in single crystal Si.

As may be seen from Fig. 1 and 3, the RBS channeling data indicate that the field of damage (Si atoms knocked out of their lattice positions) is more extensive, for a given dosage and energy, for hydrogen than it is for argon. Also the hydrogen amorphises the Si surface layer; the argon does not. However, the ESR data of Table I indicate that there are more dangling bonds in the damage layers caused by Ar^+ -- even though Figs. 1 and 3 show these Ar^+ - caused layers are thinner -- than there are in the corresponding (same energy and dosage) layers caused by H^+. This is our first indication that, although the H^+ beam is very effective in tearing up the Si lattice, it also passivates its own damage as seen by the reduced ESR signal. The I-V data of Figs. 2 and 4 support this contention; i.e., there is less positive charge present after a 1.0 H^+ ion beam exposure than there is after a 1.0 Ar^+ ion beam exposure (both situations had a total of $\sim 10^{18}$ ions/cm^2 impinging on the Si).

This brings us to the question of how can H^+ implants passivate poly Si and Mobil Solar ribbon Si without causing surface damage. We believe the answer must be that these H^+ implants do cause damage to the surfaces of these materials. The picture that emerges is as follows: As the H^+ is implanted, if the self-caused damage is not too severe, some H can escape the implant region and diffuse to bulk defects or grain boundaries for passivation. The implant induced surface damage is partially (or completely) passivated by remaining hydrogen. This layer can then remain, be etched off, or be annealed out. We speculate that, if it remains in a completed n^+/p solar cell structure, any residual positive charge in the layer would only perform the beneficial service of forming a front surface field which would aid short wave length spectral response. However, if dangling bonds also remain these could increase surface recombination in the emitter lead-

ing to a degradation in V_{oc} (if it is controlled by the emitter) and to a degradation in the short wavelength spectral response. Put another way, if V_{oc} degrades after H^+ implantation, it suggests that V_{oc} is being controlled by the emitter and emitter recombination is being enhanced by exposure to the H^+ beam.

The pertinent question of whether or not H^+ implants can passivate surface lattice damage now needs to be considered. To do that we turn to Fig. 4. Here we see the I-V characteristic for an Au dot evaporated onto Si that has been subjected to 1.0 keV Ar^+ and then subjected to 0.4 keV H^+ beams. The Ar^+ damage seen in Fig. 4 (1.0 keV Ar^+ I-V curve) is clearly passivated by the H^+ implant. In fact, the damage (as judged by the presence of positive charge) is seen from Figs. 2 and 4 to be less after the Ar^+ plus H^+ beam exposures than it is after a simple 0.4 keV H^+ beam exposure.

Comparing Figures 2 and 4 indicates that there is a synergism involved in the 1.0 Ar^+ exposure/0.4 H^+ exposure. One can speculate that the damage caused by the 1.0 keV Ar^+ beam retards the penetration of the subsequent 0.4 keV H^+ beam trapping the hydrogen in a smaller volume. This allows it to be more effective in passivating the donor levels (positive charge) caused by the Ar^+ ion beam lattice damage. However, this retardation of the hydrogen by the Ar^+ damage has yet to be established by RBS. It also must be determined if the remaining damage (after 1.0 Ar^+/0.4 H^+) is electrically inactive but detectable by ESR measurements.

It is clear that hydrogen implants can be the cause as well as the cure for surface damage in crystalline silicon. The manner in which the hydrogen acts (i.e., whether it results in a net cure or cause of damage) depends on the previous history of the surface. It remains to be determined if implanting through thin oxide or nitride layers can modify this behavior and if hydrogen implants can improve the quality of thin oxide/Si or thin nitride/Si interfaces.

III. Hydrogen Beams: A Cure for Bulk Defects

From the results of Section II, from references [1] and [2], and from the wealth of data from amorphous Si work, it seems clear that hydrogen implants can passivate dangling bonds in silicon. Hence, if hydrogen can reach bonding defects in bulk Si, it will passivate them. Recently, it has also been suggested that hydrogen can passivate deep levels resulting from impurities in silicon [7,8]. However, this latter point remains somewhat in doubt since surface damage, incurred in the act of introducing the hydrogen, can getter fast diffusing impurities. Hence it remains to establish if hydrogen is passivating deep impurity levels or, if hydrogen-caused damage, is gettering the impurities.

IV. Applications of H^+ Implants to Crystalline Silicon

A. Surface Property Modification

As was discussed in Section II, if Si is exposed to a low energy hydrogen ion beam, there is a surface layer produced which contains lattice damage. This lattice damage gives rise to donor levels which cause the barrier seen in Fig. 2. As noted in Fig. 2, there is also an insulating Si:H or Si:H:O layer produced by the lower energy H^+ ion beam exposures [4]. Its presence is not

detected in Fig. 2 (it would appear as a large series resistance in the I-V) since these samples were all given an HF etch and DI water rinse.

The presence of this apparently wide gap, insulating Si:H or Si:H:O layer, produced by the lower energy H^+ beam exposures, is seen in Fig. 5. This figure presents capacitance-voltage (C-V) data for a p-type Si sample that was subjected to a 0.4 keV H^+ ion beam. In this case the Au front contact was deposited on the H^+ implanted surface without the HF etch to allow the insulating surface film to remain. These C-V data show that the wide gap Si:H or Si:H:O layer is capable of supporting both accumulation and inversion; i.e., the layer is a good insulator. We note that the conducting, lattice-damaged layer, containing the donor levels giving rise to the barrier seen in Fig. 2, lies below this insulating layer. This same type of insulating layer appears for the lower energy H^+ implants on n-type single crystal Si samples also.

This wide gap insulating surface layer created by the lower energy H^+ implants appears attractive for surface passivation of solar cells. The H^+ implants which produce this layer seem especially attractive for n^+ emitter structures since the wide-gap layer could passivate the emitter surface while the donor layer, produced by the H^+ implant, would create a front surface field. Such H^+ implants, when done through thin oxides or nitrides, may produce optimum emitter structures. However, this all remains speculation at this point since the effect of this processing on emitter recombination has not been determined nor has there been any attempt at optimization, or at studying, the effects of the presence of thin oxide or nitride layers.

B. Low Temperature Junction Formation

From the preceeding sections and from the literature there is extensive evidence pointing towards the virtues of hydrogen in the passivation of electrically active bonding defects in silicon. Hence, it seems only natural to determine if H^+ implants can improve the quality of implanted, shallow n^+/p and p^+/n junctions. The objective is to obtain high quality, low leakage implanted shallow junctions without recourse to the high temperature processing currently employed.

In our study, n^+/p diode structures were fabricated by implanting 75-keV As^+ ions into (100) p-type Si which had channel stops present. The dosage used was 5×10^{15} As atoms/cm^2 and the implant area was 3.7×10^{-2} cm^2. Using a Kaufman-type ion source, we subsequently implanted low-energy hydrogen ions into the damaged silicon. The hydrogen ion beam had an accelerating voltage of 0.4 keV, an extractor voltage of 0.3 keV, and an accelerator current of 200 mA. These silicon samples were then annealed at 500 or 600°C for 1 hr. in nitrogen or argon ambients. Spreading resistance measurements were subsequently made to examine the dopant profile and the degree of dopant activation. Current-voltage-temperature (I-V-T) measurements were made to examine the diode characteristics and transport mechanisms. Experimental details may be obtained from ref. [9].

We found that the low-temperature anneal, following the H^+ implant, is a very necessary part of this processing. The characteristics of our diodes, immediately after the H^+ implantation but before the low-temperature anneal, are extremely poor. The high series resistance and very large leakage currents (of the order of hundreds of μA/cm^2) indicate that the damage has not been fully passivated and that the dopant is not activated by the hydrogen implant alone. The inactivity of the dopant is borne out by the spreading resistance measure-

ments. (See Fig. 6). The lack of dopant activation after the H^+ implant is not surprising as the temperature of the silicon sample during the H^+ implant rises to temperatures less than 150°C. The results of our spreading resistance measurements (Fig. 6) indicate that junction depth after a 600°C 1-hr. anneal shows very little dopant redistribution. The dopant concentration as measured from spreading resistance shows that most of the implanted As has been activated by this 600°C anneal following the H^+ implantation.

Turning specifically to the diode I-V behavior resulting from this HILT processing (from hydrogen ion-assisted, low temperature anneal), it is seen in Fig. 7 that excellent diode characteristics result even though processing temperatures have never exceeded 600°C. Leakage currents of the order of 5 nA/cm^2 at 1.5 volt reverse bias are attained for ~5-10Ω-cm p-Si base material. The forward bias n-factors in this case are n = 1.03 over ~5 decades. We have obtained similare results for p^+/n implanted junctions.

In summary, we assert that this novel, genuinely low-temperature process offers an alternative to the numerous other processes being investigated to anneal out implantation damage and to activate dopants in implanted Si. Our preliminary studies, with very little optimization, show that this low-energy hydrogen-ion implantation/low-temperature anneal processing for damage passivation and dopant activation gives results comparable to the other annealing processes (we have looked at p^+/n junctions also). In addition it allows for passivation of any residual defects and thereby provides an advantage not offered by other techniques. Although at present we use furnace anneals to activate the dopants, we are by no means limited to them and can use low-energy H^+ implants in conjunction with lower-temperature RTA approaches. This will reduce the processing time involved. With the development of broad-beam ion sources, this process can very easily be adapted to the fabrication of the extremely shallow junctions needed for solar cells.

At this time the mechanism by which hydrogen implants improve device characteristics is not clear. Indeed several groups (including us) have found that H begins to evolve from silicon at a temperature of ~350°C and is completely lost by 600°C. Thus the simple explanation of hydrogen bonding itself to the silicon "dangling bonds" does not seem to be adequate. Further work to identify the mechanisms of annealing and dopant activation is under way.

C. Enhancement of Web Solar Cell Performance

As we first pointed out at the 1983 European Photovoltaics Conference Meeting [10], low energy hydrogen ion implants can improve the performance of solar cells fabricated on Westinghouse web material. Subsequently [11] we have shown that these low energy implants reduce the J_o for the recombination-diffusion current controlling V_{oc} and, thereby, enhance V_{oc} in these cells. Also the long wavelength spectral response is improved by the implants [11] implying the diffusion length is increased in the base. This latter fact suggests that J_o is controlled by the base; consequently, V_{oc} and the spectral response improve due to an improved base diffusion length. From this picture one is led to the deduction that the H^+ implant is passivating bulk defects in the web Si.

There is an alternative explanation: the H^+ implants may be reducing recombination in the emitter; i.e., J_o may be controlled by the emitter. The enhanced long wavelength spectral response would not be due to enhance base properties in

this picture; but, rather, it would be due to a widened band gap in the emitter which allows long wavelength photons to penetrate into the base. That is, before the H^+ implant the band gap in the emitter is narrowed due to the high doping levels used in the Westinghouse cells. This narrow band gap causes enhanced long wavelength photon absorption in the emitter. After the H^+ implant, the band gap is widened and, as noted, these photons can now penetrate into the base giving an apparent increase in the base diffusion length.

Whether base material properties are actually being improved by these H^+ implants into web material or whether all the changes (reduced recombination and widened band gap) are taking place in the emitter remains to be determined. Given the fact that sample history affects the way hydrogen interacts with Si - as we established in Section II - it follows that the interaction of H^+ implants with web Si can be a complicated phenomenon which needs additional study.

V. Conclusions

As a result of our examination to date of the effects of hydrogen ion beam exposure on crystalline silicon, the following conclusions may be drawn:

1. Hydrogen ion exposure can cause, as well as cure, surface defects. These defects include damage-caused donor states and dangling bonds. Whether a hydrogen implant results in increased damaged or passivation depends on the energy of the implant and on the past processing history of the silicon surface.

2. Very low energy hydrogen ions can produce a wide-gap insulating film at the surface of silicon. It may be possible to use this layer, perhaps by implanting through thin oxide or nitride films, to passivate surfaces.

3. Hydrogen ion exposure can passivate bulk bonding defects.

4. Hydrogen implants can be used to give high quality, shallow implanted junctions. The use of the hydrogen implant allow this to be accomplished with processing temperatures of only 600°C.

5. Low energy hydrogen implants can improve solar cell performance. It appears that there are two possible scenarios that can explain this:

   One assumes that the hydrogen implants do not affect the emitter or, in the other extreme, damage the emitter. It is assumed that the implants passivate bulk defects. In this case, cells where V_{OC} is controlled by the base would show enhancement of the V_{OC} after the hydrogen implant and cells where V_{OC} is controlled by the emitter would show no change in V_{OC} or degradation. In both situations the long wavelength spectral response should improve.

   The other scenario assumes that the hydrogen implants reduce recombination in the emitter and also may widen the band gap

in the emitter, depending on emitter doping level. In this case, cells where V_{OC} is controlled by the base would show no change in V_{OC} after an implant; however, cells where V_{OC} is controlled by the emitter would show improvement. In both situations the long wavelength spectral response may improve due to band gap widening.

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References

1. C. H. Seager, D. J. Sharp, J. K. G. Panitz, and R. V. D'Aiello, J. Vac. Sci. and Technol., 20 (3), 430, 1982.

2. J. Hanoka, C. Seager, D. Sharp, and J. Panitz, Appl. Phys. Lett. 42, 618 (1983).

3. R. Singh, S. Fonash, P. Caplan, and E. Poindexter, Appl. Phys. Lett., 43, 502 (1983).

4. R. Singh, S. Fonash, S. Ashok, P. Caplan, J. Shappirio, M. Hage-Ali, and J. Ponpon, J. Vac. Sci. Technol. A1, 334 (1983).

5. J.-S. Wang, S. Fonash, and S. Ashok, IEEE Electron Device Lett., EDL-4 433 (1983).

6. A. Climent and S. J. Fonash, J. Appl. Phys., July 1984.

7. S. Pearton and A. Tavendale, Phys. Rev. B26, 7105 (1982).

8. S. Pearton, J. Kahn, W. Hansen, and E. Haller, J. Appl. Phys. 55, 1464 (1984).

9. R. Singh, S. J. Fonash, A. Rohatgi, P. Rai-Choudhury, and J. Giganti, J. Appl. Phys. 55, 867 (1984).

10. R. Singh, S. Fonash, A. Rohatgi, and P. Rai-Choudhury, Proc. European Photovoltaic Conference, Athens, Greece (1983).

11. D. Meier, A. Rohatgi, R. Campbell, P. Alexander, S. Fonash, and R. Singh, Photovoltaic Specialists Conference, Orlando, Florida, May, 1984.

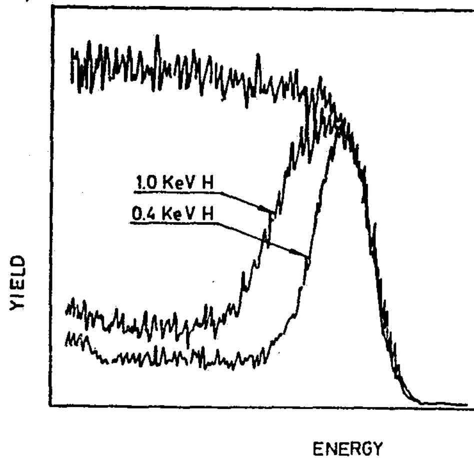

Figure 1. Rutherford backscattering channeling data for 0.4 keV and 1.0 keV H^+ beam exposures. Also shown is the random yield.

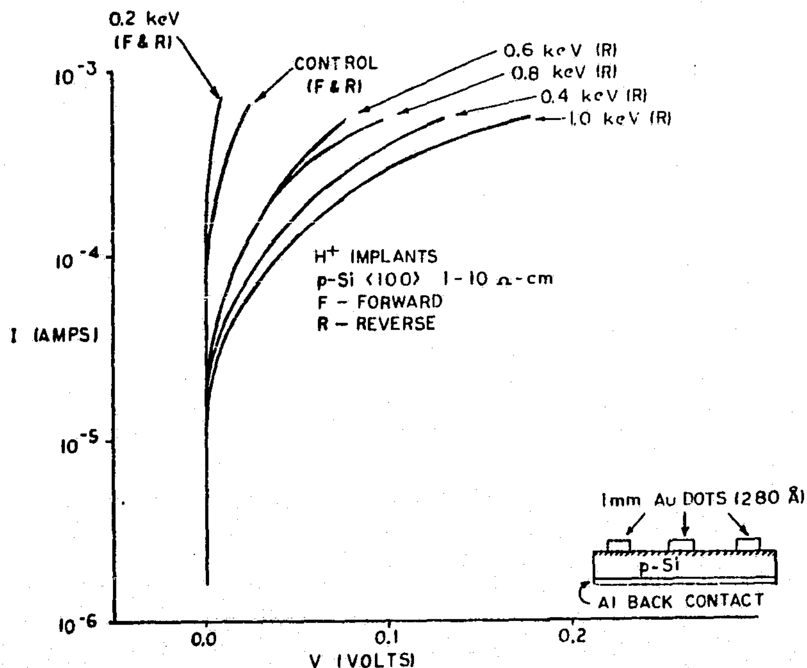

Figure 2. Current-voltage (I-V) data for Au dots evaporated onto p-Si surfaces subjected to various hydrogen ion beam exposures. The Au dots were deposited after the H^+ exposures and after an HF etch and DI water rinse. This etch and rinse is used to remove a high resistance Si:H or S:H:O layer that forms for the lower energy beams. See Figure 6.

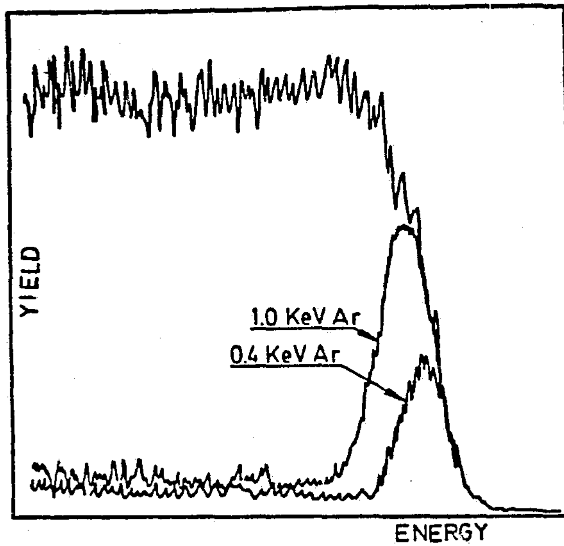

Figure 3. Rutherford backscattering channeling data for 0.4 keV and 1.0 keV Ar^+ beam exposures. Also shown is the random yield.

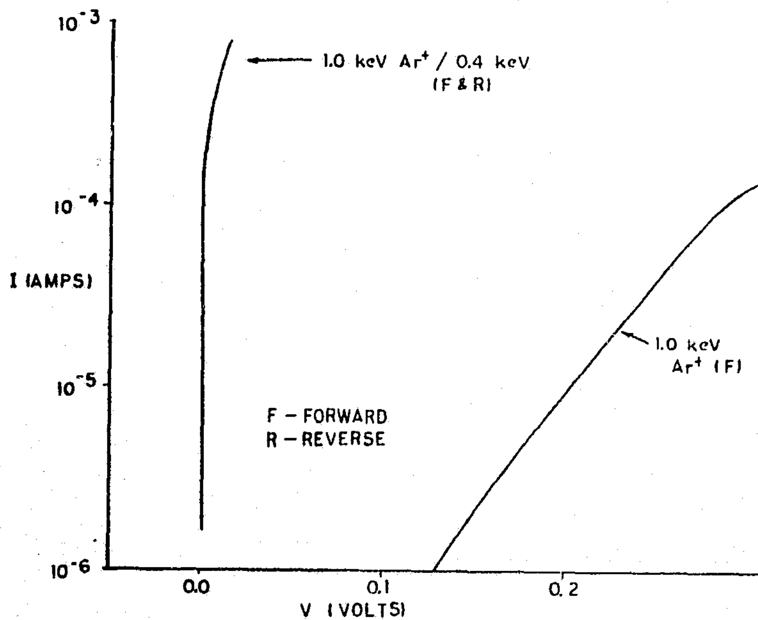

Figure 4. Current-voltage (I-V) data for Au dots evaporated onto p-Si surfaces subjected to 1.0 Ar^+ exposure. In one case the Ar^+ exposure was followed by an H^+ exposure. In each case the Au dots were deposited after the ion exposures and after an HF etch and DI water rinse.

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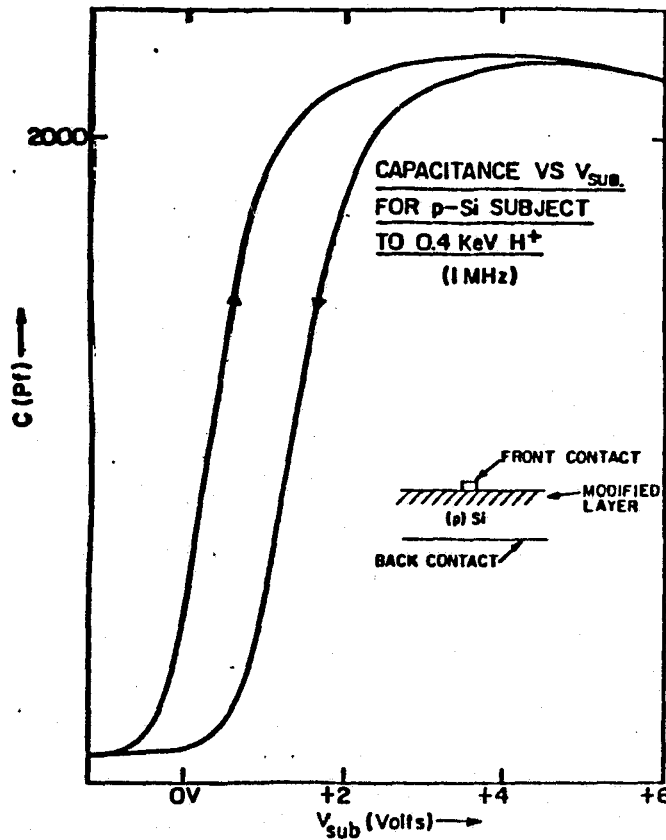

Figure 5. Capacitance-voltage data for p-Si which was implanted with H^+ (0.4 keV beam energy) but which was not given the subsequent HF etch and DI rinse used for the samples of Fig. 2. Accumulation and inversion are clearly occurring due to the presence of an insulating surface film.

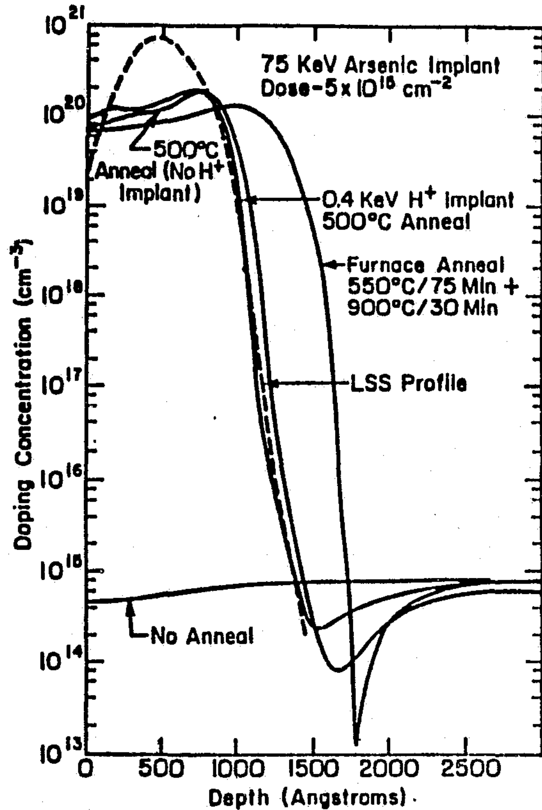

Figure 6. Spreading resistance plots for 75 keV As^+ implanted Si. Treatments following the As^+ implants are shown on the figure as is the theoretical LSS profile.

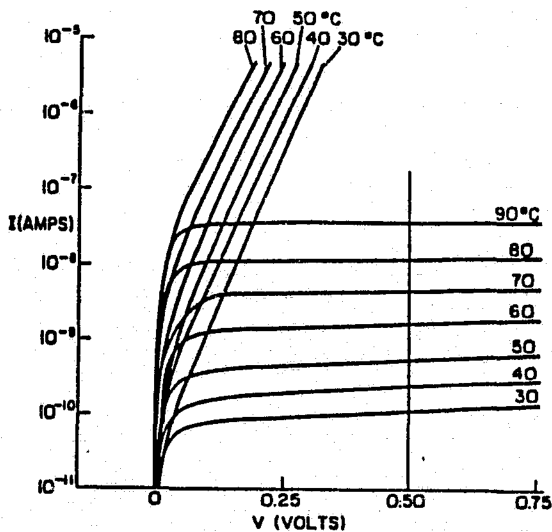

Figure 7. Current-voltage-temperature data for As^+ implanted Si. This n^+/p diode structure has had a 0.4 keV H^+ implant for 1 minute followed by a 600°C 1 hour furnace anneal in N_2. The diode area is 3.7 x 10^{-2} cm^{-2}.

DISCUSSION

TURNER: I would propose an experiment by which you might be able to differentiate between those two models. That is, to shine blue light and red light -- as long as it is not infrared -- on the back of the cell, cut a window through the back metallization and shine light on that, and that will be sensitive to the diffusion length through the base of the material. The emitter won't make any difference, except for minor reflectivity.

FONASH: That is a good suggestion.

KAZMERSKI: Or you could try implanting through the rear.

FONASH: We haven't tried the former suggestion, the latter suggestion is being tried right now. We are trying implanting through the rear.

LESK: In your one-minute implantations, with your system, what dose are you getting?

FONASH: The dosages are on the order of $10^{18}/cm^2$.

LESK: And that is your general dose you are using for all this work?

In the last work you talked about, the emitter band-gap widening, were your measurements taken before the AR coat or after, and if after, did the temperature involved in the AR coat deposition have any effect on the annealing?

FONASH: The measurements that I just showed for the web material -- all the data that I showed was before AR coating. There was no temperature processing involved at all in that. Now Westinghouse has subsequently taken these cells and put on AR coatings to see what would happen and the performance continues to be enhanced to the degree that one would expect from an AR coating. They have also done -- and again, I think Ajeet (Rohatgi) will probably discuss this -- Westinghouse has also taken these cells that we have implanted and they have done some temperature stability studies on them, and what have you, and the answer to your general question is that they don't find any destabilization on the cells with moderate temperature cycling.

SCHRODER: Have you done any DLTS measurements on the damage layers themselves?

FONASH: Yes we have.

SCHRODER: The data you showed was after it was removed, on the gold sample?

FONASH: Yes it was.

SCHRODER: On the previous samples?

FONASH: We have done DLTS, and we see a broad damage signal around the middle

of the gap, a little bit above the middle of gap. We have also implanted silicon into silicon and done DLTS to compare the damage signals, and they compare. But that is work that we have just been into for about a month or so. I would not really like to comment on that too much, other than to say that we are doing precisely what you have suggested.

SCHRODER: OK. Another question is, do you think your hydrogen implant conditions are similar to what I assume you referred to -- proton work, plasma, or even recently, I think, they use an electron technique?

FONASH: That is right.

SCHRODER: Are the conditions similar, you think?

FONASH: In a plasma, as you know, the governing quantity that one needs to look at is the voltage across the sheet, because that is the energy that the ions would impact the surface with. So I really don't know; given the power, there is no way to convert that directly into what the energy of the ions is, so I can't comment on whether the situation is the same. I don't know the energy of their ions when they impact the surface. They are also talking about a process where they do the hydrogen treatment at $300^{o}C$, as I remember. It is done for something like two hours. To summarize my answer, I don't know the energy of their ions. Theirs is a process done at elevated temperature, and it is a process that is done for some two hours. Ours is a process that is done at essentially room temperature, at one minute, for very carefully controlled ion energy.

LOFERSKI: I was going to suggest perhaps a third mechanism that would account for the increase of diffusion length in the Westinghouse web samples. Perhaps, also, for what is going on in that gold-doped cell. You know, for one thing, the implantation certainly produces a lot of vacancies in the silicon, at interstitials and vacancies. You know the work by George Watkins -- and the ESR studies in silicon -- and of other people, that shows that vacancies form complexes and these complexes are what result in the deep levels that control lifetime. Now vacancies can also -- some of the complexes can be in a direction where they wipe out something that was a lifetime killer, and, you know, just neutralize it, so that it is possible that what is happening in both cases is that you have generated vacancies in excess of what you would have had at that temperature and they go down into the complexes; the vacancies are very mobile. Watkins showed that you don't see any free vacancies, and if you produce them at temperatures higher than, I think something at 1K, they move rapidly even down there at 20K, and so forth.

FONASH: I realize that.

LOFERSKI: So, anyway, that is another possibility.

FONASH: I think it is a possibility. That is one of the reasons why we are going to the implant from the back. But there is always a possibility that that experiment, implant from the back, could be clouded by the vacancies coming all the way up to the emitter. One could always argue that that is not the definitive experiment. I am aware of the high

mobility of vacancies, and it does indeed further cloud the issue. Getting back to the gold, I don't think it is the explanation there, because I think we have pretty well established that the gold is captured in that damaged layer and we remove it when we etch off the damaged layer.

LOFERSKI: I was wondering about the gold, whether anybody has studied gold and vacancies, complexes; whether Watkins has done that or not?

FONASH: I don't know.

MILSTEIN: As you may be aware -- I am sure aware -- we also looked at hydrogen implants at 300 eV in web materials and our RBS data and some other things differ a little bit from yours. At 400 eV for 1000 minutes of passivation, just to make sure we saw the damage if it was going to be there. We saw no amorphous material, and that is consistent with the fact that we saw no change in the reflectance under those implant conditions. However, when we went to 900 eV or 1500 eV, we saw several hundred Å of amorphous material, and in fact we saw a very dramatic change in the absorption, which is consistent with the appearance of amorphous material. In that sense, I would question whether at 400 eV, which is a mere 100 eV above where we were, you are seeing the kinds of things you are seeing.

FONASH: Well, I don't question it, because we have reproduced it several times, so I know it's real. And I think these other data -- there is a group at IBM for example, that has used Rutherford back-scattering to study damage incurred in reactive ion etching, and they have found that for comparable energies, 300 eV, 400 eV, the RBS does indeed detect a substantially damaged layer. So I don't doubt the data, that there is a damaged layer that is produced there at 400 eV.

MILSTEIN: Well, we saw latticed damage, 1600 Å of lattice damage, we did not see amorphous at 400 eV.

FONASH: Oh, you mean the RBS yield was not up to the amorphous level?

MILSTEIN: That is correct.

FONASH: Well, I don't know the explanation for that other than to suggest that perhaps you did some thermal treatment. It could be that your energies are not the same.

MILSTEIN: No.

FONASH: Well, we find this reproducible.

SAH: I would like to ask you if you have tried it on an encapsulated silicon?

FONASH: Yes.

SAH: What happens?

FONASH: Well, what do you want to discuss?

SAH: For instance, you called this level a donor level.

FONASH: That is right, it is symmetric. It lowers the barrier in n-type silicon. We have published that in a number of places.

SAH: It is still donor, or is it not donor?

FONASH: It is still donor. It lowers the barrier in n-type silicon. In other words, you get an n^+n type of layer on n-type silicon. You see, it's one of the first things we did to see if the effect is symmetric between p and n silicon, and indeed it is. It is not a special case of a donor that shows up for p material and mysteriously disappears for n type material.

HANOKA: The stuff we've done leaves no doubt that the first of your two explanations is the right one. The second one is not, because we have done a lot of diffusion-length measurements using, let's say, a liquid-junction technique, which does not change the sample at all. It is totally nondestructive. It doesn't heat it or anything. We take a sample with no junction, just measure diffusion length for an implanted sample. Diffusion length definitely goes up, and it's definitely a bulk effect. We etch off samples and measure it again. The same sort of thing. Etch off some of the material and measure again.

FONASH: What energies are you using now?

HANOKA: These are higher energies. I think 1500 to 1700 eV.

FONASH: I think we have to keep that in mind -- that you are talking about energies that are 1500 eV.

HANOKA: But I don't see where that makes a difference here.

FONASH: Well, just last week we did some implants for Westinghouse. We took a sample -- Westinghouse had measured the diffusion length by surface photovoltage -- we did the implant and gave them the samples back. They measured the diffusion length by surface photovoltage and it didn't change. But we are down around 400 eV.

HANOKA: We do find the diffusion length changes from sample to sample. It is a function of the kind of defects you have in the sample.

FONASH: Well, I think an important point is, you are looking at your material, are you not?

HANOKA: This is only on EFG now.

FONASH: Right. And we are looking at web material, and the nature of the defects is quite different, is it not? In your material don't you have a lot of defects that intersect the surface, and so one could think of conduits that could carry the hydrogen down into the bulk.

HANOKA: That's right.

FONASH: We have a different kind of material, so I think we have to watch an apples-and-oranges kind of comparison.

HANOKA: Well, I still think it is apples, but I think it is McIntosh and Delicious.

LOFERSKI: I just wonder if the difference between your results and his could relate to temperature control.

FONASH: That's what I think.

LOFERSKI: Dose up to 10^{17} cm^{-3} in one minute at 400 eV. That is a big jolt of energy, and the temperature may be fluctuating.

FONASH: Well, of course it works the other way though, doesn't it? You would think that we might get some regrowth if we are dumping all that hydrogen in in such a short time, but yet he is claiming that perhaps he's got some regrowth. So I think it is going in the opposite direction.

LOFERSKI: If he has the same dose as you have for that long he is going to explode the surface.

FONASH: I think he is using the same dose.

HANOKA: A factor of three lower.

FONASH: Oh, is it a factor of three lower?

LOFERSKI: A lot less.

FONASH: I think temperature is the key to that difficulty. I don't think there is anything fundamentally significant in that.