JOURNAL OF APPLIED PHYSICS VOLUME 84, NUMBER 2 15 JULY 1998

Aluminum-implantation-induced deep levels in *n*-type 6H–SiC

S. Fung,[a] M. Gong, and C. D. Beling
Department of Physics, The University of Hong Kong, Hong Kong

G. Brauer, H. Wirth, and W. Skorupa
Institut fur Ionenstrahlphysik und Materialforschung, Forschungszentrum Rossendorf, Postfach 510119, D-01314 Dresden, Germany

(Received 30 January 1998; accepted for publication 31 March 1998)

Deep-level defect centers on the n-side of p^+n junction diodes formed by low and elevated temperature aluminum-ion implantation into n-type 6H–SiC have been studied using deep-level transient spectroscopy. Two shallow Al-acceptor levels have been observed in the n region just beyond the implantation depth through their minority-carrier emission signatures. The dominant level is situated at 0.26 eV above the valence band and is accompanied by a shallower level of small intensity. Comparison with photoluminescence results suggests the dominant level (labeled A_k) and the shallower level (labeled A_h), are associated with the cubic and hexagonal lattice sites, respectively. Unlike previously reported results, which show many different implantation-induced donors within the implantation region, only one deep donor level at $E_C-0.44$ eV is found to occur in the postimplantation region, indicating that the various crystal damage sites occur with different spatial distributions. © *1998 American Institute of Physics.* [S0021-8979(98)05213-X]

Ion implantation is the only practical method of obtaining selective area doping in SiC, not only because of a very low diffusion coefficient of the impurities but also because no dielectric masking layers are available that can withstand the temperatures for significant diffusion.[1] A problem with the implantation procedure is that a high-temperature annealing is necessary to remove irradiation damage and electrically activate the implanted dopant. This processing, however, causes impurity redistribution.[2–4] Recently, an interesting review of Al and B implantation studies has been presented by Troffer *et al.*[5] Generally, as observed in other semiconductors, two kinds of redistributions were noted. The first is an out-diffusion, drawing the dopant towards the surface, and thereby decreasing its concentration, while the second is an in-diffusion, which introduces a dopant tail extending into the deeper bulk region.[4–7] These processes are both associated with diffusion mechanisms that have been enhanced by implantation damage. In the specific case of boron implantation, a number of secondary ion mass spectroscopy (SIMS) studies have revealed a long in-diffusion boron tail in n-type 6H–SiC.[2–4] Our previous work[4] has shown that the boron tail, with a total boron atom concentration larger than that of the donor in the substrate, overlaps the depletion region of the pn junction formed by boron implantation, and is D-center dominant. In that work, we pointed out that the existence of a high concentration defect region overlapping the depletion layer is probably harmful for devices.

With regard to aluminum implantation, most SIMS experiments in 6H–SiC presented no obvious aluminum redistribution caused by in-diffusion during annealing.[2,3] However, it is noted that the aluminum profile resulting from primary implantation is broader than that of boron. The importance of this is that, since only a small part of the implanted Al atoms act as acceptors at room temperature,[5] the concentrations of the nonactive aluminum and the other residual damage can still be considerable in the n region behind the implantation region. The purpose of this communication is to present deep-level transient spectroscopy (DLTS) measurements, which reveal that this region does indeed contain both acceptor and donor states induced by Al implantation.

The experimental procedure was as follows: N-type epilayers with a donor concentration of $7\times10^{15}\ \mathrm{cm}^{-3}$ were grown on $5\times10^{18}\ \mathrm{cm}^{-3}$ nitrogen-doped substrates obtained from CREE Research Inc. Aluminum implantations were carried out with various energies so as to form a box implantation profile. A set of such implantations was performed at different substrate temperatures from 20 to 1200 °C. The final mean concentration of the dopant was $7\times10^{19}\ \mathrm{cm}^{-3}$ as obtained by TRIM code simulation and confirmed by SIMS. After Al implantation, all samples were annealed at 1700 °C and cut into $3\times3\ \mathrm{mm}^2$ samples. These were then etched by reactive-ion etching so as to bring the Al-implanted layer to the surface and to decrease the edge leakage current for the DLTS measurements. Ohmic contacts on n- and p-type sides were made as described in detail in our previous work.[4] Finally, the capacitance–voltage and the current–voltage measurements, carried out at both room temperature and liquid-nitrogen temperature, showed that the pn junctions had good rectifying properties.

The DLTS spectra were measured under majority-carrier injection conditions by applying a reverse bias of $V_r=-6$ V with a forward filling pulse of $V_p=6$ V. The experimental results of the samples implanted at room temperature and 1200 °C are presented in Fig. 1. Two strong deep-level peaks can be easily seen within the temperature range from 80 to 400 K. The positive peak is due to majority-carrier trapping while the negative peak is due to minority-

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[a] Electronic mail: sfung@hkucc.hku.hk

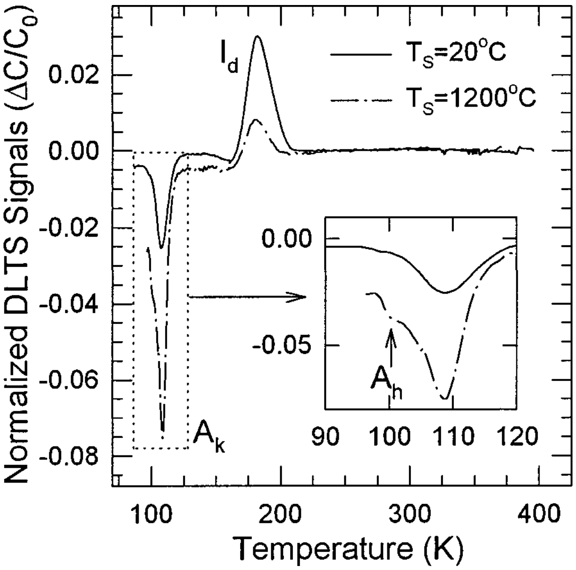

FIG. 1. Typical normalized DLTS spectra of 20 and 1200 °C Al-implanted n-type 6H–SiC samples. Inset shows the shoulder peak of the hexagonal lattice site (A_h).

carrier trapping. The other samples formed at intermediate substrate temperatures gave the same basic DLTS spectra but with different peak amplitudes. The positions of the majority and minority deep levels in the band gap, as determined by the Arrhenius plots shown in Fig. 2, are $E_C - 0.44$ eV and $E_V + 0.26$ eV, respectively.

Neither the majority nor the minority peaks seen in the Al-implanted DLTS spectra appeared in our recent study of B-implanted samples formed under similar conditions.[4] The negative signal, which is labeled as A_k in Fig. 1, is unusual in that it appears without the condition of minority-carrier (hole) injection in exactly the same manner as observed for the D center in B-implanted SiC.[4] As discussed in our recent study,[4] this indicates that the center responsible for this level comes from the minority-carrier tail region that exists in the ionized n region of the depletion layer, which lies just beyond the ion-implantation (p-type) range. The essential point is that there exists on the n side of the pn junction a narrow region in which hole capture (emission) from a deep-level acceptor can occur as the intersection position of the hole Fermi energy and the trap level changes under zero (reverse) bias. Because the minority-carrier tail region is basically much smaller than that of the n-side depletion layer, from which the majority signals arise, this defect must have a much higher concentration compared to that of the shallow donor in this region. The results also imply an aluminum-related involvement for this defect since it exists only and always in Al-implanted samples. The position of this level, at $E_V + 0.26$ eV, being so close to the Al-acceptor level at 0.23–0.28 eV above the valence band measured by photoluminescence (PL),[8] admittance spectroscopy,[9] and Hall effect[10] in p-type SiC, strongly suggests that it is the shallow Al-acceptor. In particular, there is always a small shoulder (A_h) on the low-temperature side of the A_k peak in all the DLTS spectra as shown in the inset in Fig. 1. This phenomenon probably supports the observation of PL,[8] namely, that the Al atoms occupy both h (hexagonal) and k (cubic) lattice sites. According to the Ikeda's *et al.* results,[8] the aluminum atom with the shallower level is at the h site, while the deeper level results from the atom at the k site. The amplitude of A_k, in our work, is larger than that of A_h, which also agrees with the photoluminescence data.[8]

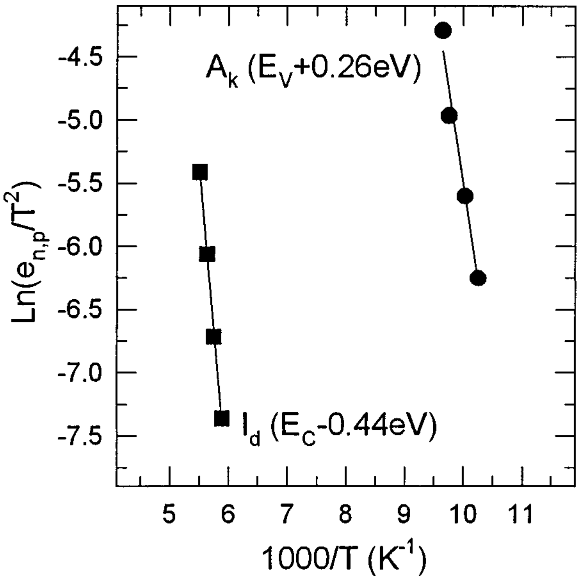

FIG. 2. Arrhenius plots of electron or hole ($e_{n,p}$) emission rates as a function of 1000/T, for deep levels corresponding to minority-carrier peak (Al-acceptor A_k ●) and majority-carrier peak (Al-induced defect I_d ■).

A question that arises from our observation of the Al acceptor in n-type SiC is why other workers have not seen this level under minority injections in this material.[6] In this respect, we point out that in the present study a much greater concentration (7×10^{19} cm^{-3}) of aluminum atoms was implanted (cf. 4×10^{16} cm^{-3} in Ref. 5). Indeed, the concentration was sufficiently high that under annealing activation a pn-junction diode was manufactured. Moreover, implantation profile broadening due to ion range straggling, produces in our samples an ionized aluminum-acceptor region, with a concentration only slightly less than that of the donors, on the n side of the postimplantation region. This coupled with the fact that the nitrogen donor is much shallower ($\sim$0.08 eV) than the Al acceptor ($\sim$0.26 eV) ensures that the free-electron tail is not frozen out at the temperatures required to observe hole emission from the acceptor state. The sample thus remains DLTS active and with the high concentration of acceptor levels within the n-region measurement of the minority-carrier trap is possible.[4] We thus conclude that in the work of Troffer *et al.*[5] the concentrations were too low for the Al-acceptor level to be observed.

As previously mentioned, our samples showed an Al-implantation-induced deep-level donor signal at $E_C - 0.44$ eV, which we label as I_d (induced donor). The observation of this single level is to be compared with the work of Troffer *et al.*[5] who find at least six different Al-induced deep levels (labeled ID$_3$, ID$_4$, ID$_8$, ID$_9$, RD$_1$, and RD$_2$ in Fig. 7 of Ref. 5) in both 4H and 6H–SiC, using DLTS in the temperature range of 100–400 K. It is not clear whether the I_d level we see corresponds to one of the levels seen by Troffer *et al.* Their ID$_9$ peak has a very similar energy but a vastly different cross section (some four orders of magnitude larger). A more likely candidate is their low intensity peak ID_8 that, although not matching well in energy, does appear at the correct temperature. It is thus possible that I_d is a new defect or possibly the ID_8 level. Irrespective of whether this is so or not, in comparing the spectra of Troffer with ours one may arrive at an important conclusion, namely, that various damages with different distributions have been introduced during implantation. Specifically, it seems that I_d is abundant in the deeper tail region beyond the implantation and is relatively dilute near the surface. The latter follows, in that, in the work of Ref. 5, a Schottky contacted sample was employed thus yielding information only on defect centers within the near-surface implantation region.[5] This phenomenon is similar to that which has been observed by us, boron-implanted n-6H–SiC, where none of the implantation-induced donor levels were found in the postimplantation range.[4] The presence of a different spectrum of induced defects close to the surface and deeper into the implantation range is most likely to be a consequence of the various selective properties of surface defect gettering and damage-enhanced diffusion. It is also of interest that unlike the defects observed by Rutherford backscattering/channeling spectroscopy in which the damages can be annealed out,[5,11,12] the I_d level is very thermally stable once it is formed even after undergoing a 1700 °C heat treatment.

Although it did not appear in the postimplantation region of boron-implanted n-type 6H–SiC,[4] we point out that the center responsible for the deep-level I_d is still possibly a non-Al-related defect. The reasoning is that in the case of boron implantation an interaction between substitutional boron atoms and a native vacancy occurs to form the well-known D center.[4,11] It is quite possible that no similar interaction between the Al atom and induced vacancies exists in the case of Al implantation. This would mean no damage-enhanced aluminum in-diffusion, and moreover, no removal of the primary vacancy damage sites by vacancy–Al formation. Under this picture, the I_d level would be associated with primary radiation damage, and thus probably vacancy related. It is clear that to understand this complicated case, further studies are needed, such as implantation with heavier atoms to see if the same I_d level can be formed.

As mentioned, similar DLTS spectra were measured in the low- and elevated-temperature Al-implanted samples but with different peak intensities. Slight variations in sample contacting made absolute determinations of the A_k and I_d concentrations unreliable. The ratio of the A_k level to the I_d level in the samples of different implantation temperature, however, is not subject to this systematic error. This ratio is presented in Fig. 3. It is obvious that the relative intensity of Al acceptors (A_k) is smaller than that of I_d in the room-temperature implanted sample, whereas their relation is reversed under high-temperature implantations above 400 °C. This is a technologically useful observation since by using a higher-temperature implantation procedure, more electrically active aluminum can be obtained with comparatively less induced damage.

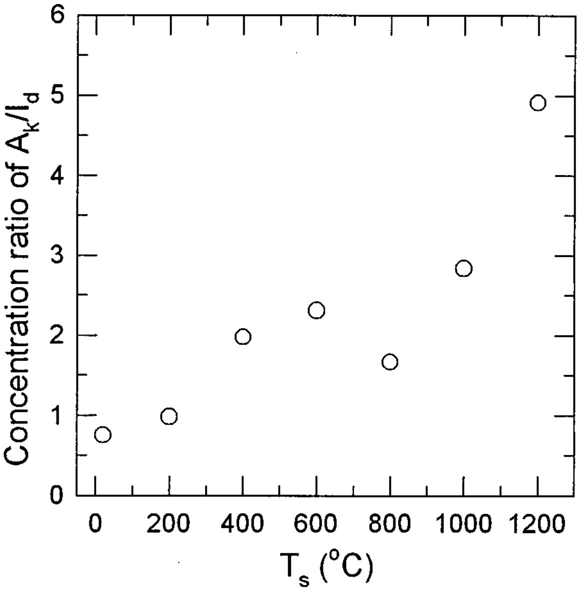

FIG. 3. The proportion of the A_k-level and I_d-level concentrations vs the substrate temperature (T_S) of implantation.

In conclusion, shallow Al acceptors in n-type 6H–SiC have been observed at about 0.26 eV above the valence band by DLTS measurements. Two closely spaced DLTS signals arising from the Al acceptors support photoluminescence results that suggest aluminum atoms can occupy inequivalent hexagonal and cubic sites in the lattice of 6H–SiC. The existence of an Al-implantation-induced deep trap I_d ($E_C - 0.44$ eV) implies a complicated family of implantation damages. At this stage, without additional information available, the structures and the properties of these defects cannot be ascertained, and thus, the need for more studies is indicated.

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