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Ion damage buildup and amorphization processes in $\text{Al}_x\text{Ga}_{1-x}\text{As}$

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The nature of keV ion damage buildup and amorphization in $\text{Al}_x\text{Ga}_{1-x}\text{As}$ at liquid-nitrogen temperature is investigated for various Al compositions using Rutherford backscattering channeling, transmission electron microscopy, and in situ time-resolved-reflectivity techniques. Two distinct damage buildup processes are observed in $\text{Al}_x\text{Ga}_{1-x}\text{As}$ depending on Al content. At low Al content, the behavior is similar to GaAs whereby collisional disorder is "frozen in" and amorphization proceeds with increasing dose via the overlap of damage cascades and small amorphous zones created by individual ion tracks. However, some dynamic annealing occurs during implantation in AlGaAs and this effect is accentuated with increasing Al content. For high Al content, crystallinity is retained at moderate ion damage with disorder building up in the form of stacking faults, planar, and other extended defects. In the latter case, amorphization is nucleation limited and proceeds abruptly when the level of crystalline disorder exceeds a critical level. The amorphization threshold dose increases with increasing Al composition by two orders of magnitude from GaAs to AlAs. Dynamic annealing and damage creation processes during implantation compete very strongly in $\text{Al}_x\text{Ga}_{1-x}\text{As}$ even at liquid-nitrogen temperatures. This behavior is discussed in terms of both the availability of very fast mobile defects and bonding configurational changes related to the Al sublattice in $\text{Al}_x\text{Ga}_{1-x}\text{As}$ of high Al content. © 1995 American Institute of Physics.

I. INTRODUCTION

Ion beam processing is a well-established technique for doping of semiconductor devices and more recently, for electrically isolating III-V compounds and compositional disordering of multilayer structures. In doping applications, one of the main drawbacks of ion implantation is the resulting damage or even amorphization of an, otherwise, good crystalline material, thereby making its electrical and/or optical properties undesirable. For electrical isolation and superlattice disordering applications, ion beam disordering processes can be used to advantage to controllably introduce defects. Although ion beam interaction processes with Si and other III-V binary compounds such as GaAs and InP have been studied quite extensively, there remains a lack of understanding of ion disordering processes in ternary alloys such as AlGaAs, which is an integral component for optoelectronic devices. For example, what is the relationship between damage formation (or amorphization) and the fraction of Al in AlGaAs? What are the dynamic annealing mechanisms that control stable defect formation during ion implantation in AlGaAs? We have recently shown that by comparing the damage behavior of AlGaAs, GaAs, and Si, the amorphization processes can be decidedly more complex when irradiation takes place in a critical temperature regime where defect annihilation processes compete with defect generation by ion interactions. Apparently the final defect state depends on defect migration, annihilation, and trapping processes at surfaces, interfaces, and other defects. Furthermore, it has also been reported that resistance to amorphization in AlGaAs increases with Al content. In fact AlAs has been shown to be extremely resistant to damage even at liquid-nitrogen temperature. In this study, a systematic investigation is carried out on damage buildup and amorphization produced by keV Si ion beams in bulk AlGaAs of seven different Al compositions.

II. EXPERIMENTAL

All structures in this work were grown on epi-ready LEC semi-insulating GaAs, with the (100) axis $\theta^\circ$ off normal towards (110), using the ANU (metalorganic chemical vapor deposition (MOCVD) reactor. The growth temperature was at 750 °C and the reactor pressure was at 76 Torr. Trimethyldichloroaluminum (TMA) and trimethylgallium (TMG) were used as group III sources with hydrogen as the carrier gas while arsine was used as the group V source. An initial GaAs buffer layer of ~$0.5-1 \mu m$ was first grown on all samples followed by AlGaAs layers of thicknesses ~$8000 \AA$ which were then terminated with thin GaAs capping layers of 30–150 $\AA$ to prevent oxidation of Al-rich layers. The thickness of the capping layers were carefully chosen so that a protective layer still existed after high-dose implants ($>10^{15} \text{ cm}^{-2}$), where a significant amount of sputtering occurs and yet not thick enough to significantly effect ion beam interaction with the AlGaAs. Al$_x$Ga$_{1-x}$As of seven Al compositions were grown ($x=0, 0.26, 0.49, 0.71, 0.83, 0.95$, and 1). Implantation was carried out using the ANU 1.7 MV tandem accelerator with 90 keV Si$^+$ ions. During implantation, the samples were held at liquid-nitrogen (LN2) temperature and tilted 7° from the beam axis to minimize channeling effects. A large range of ion doses was used (from ~$2\times10^{13}$ to ~$2\times10^{16} \text{ cm}^{-2}$) and the dose rates were in the range of 0.01–10 $\mu \text{A/cm}^2$. During implantation and warming up to
FIG. 1. RBS-C spectra showing the damage buildup for 90 keV Si irradiation of GaAs at liquid-nitrogen temperature.

room temperature, selected samples were monitored in situ by time-resolved reflectivity (TRR) by using a red HeNe laser emitting at 6328 Å. Analysis was carried out at room temperature by the Rutherford backscattering-channeling (RBS-C) technique using 2 MeV He$^+$ ions. The Si surface barrier detector was set at a scattering angle of 100° (10° grazing exit angle to the surface) to optimize the depth resolution in the region of interest. Selected samples were then analyzed by plan-view or cross-sectional transmission electron microscopy (TEM). For plan-view TEM, the samples were jet thinned using a 5% bromine in methanol solution until perforated. Cross-sectional TEM (XTEM) specimens were prepared by ion beam thinning using a cold stage. A 3 kV Ar ion beam was used during thinning to avoid preferential thinning of AlGaAs compared with Si.

III. RESULTS

Figures 1 and 2 show the RBS-C spectra for GaAs and three Al$_{1-x}$Ga$_x$As samples of different Al compositions. As observed by channeling, the damage buildup in AlGaAs appears similar to that previously reported for GaAs, in so far as the amount of disorder increases with increasing ion dose until a buried damage peak is formed at a depth which corresponds to the maximum energy deposition at around 550–750 Å from TRIM90 (Ref. 17) calculations, depending on the Al composition. In Fig. 2, the variation in the surface area is related to the different thickness of GaAs cap used for each Al composition, ~30 Å for Al$_{0.26}$Ga$_{0.74}$As and Al$_{0.49}$Ga$_{0.51}$As, and ~80 Å for Al$_{0.83}$Ga$_{0.17}$As. This buried damage peak could consist of clusters of point defects, a band of extended defects and possibly an overlap of small amorphous zones created by individual ion tracks. With increasing ion dose, the density of these defects increases until a level when the aligned signal just coincides with the random level. One interpretation of the channeling data is that, at this stage, a buried amorphous layer is just formed. Although TEM studies indicate that a continuous amorphous layer is not necessarily formed at such a dose, they show that the amount of damage is extensive and that the dose is very close to a threshold for producing amorphous material. In fact, a slight increase in dose does create amorphous material as indicated by the diffuse “amorphous rings” in the inset TEM diffraction patterns. It is thus reasonable to take the dose at which RBS-C spectra just reach the random level as a convenient measure of the onset of amorphization. Therefore, for Al$_{0.26}$Ga$_{0.74}$As, the onset of amorphization is at ~6×10$^{13}$ cm$^{-2}$, for Al$_{0.49}$Ga$_{0.51}$As, the dose required is ~1×10$^{14}$ cm$^{-2}$ while for Al$_{0.83}$Ga$_{0.17}$As, the threshold is ~8×10$^{14}$ cm$^{-2}$. It is also worth mentioning that an intermediate AlGaAs composition (Al$_{0.7}$Ga$_{0.3}$As) shows the same damage buildup behavior.

Figure 3 shows the damage profiles for GaAs, Al$_{0.26}$Ga$_{0.74}$As and Al$_{0.49}$Ga$_{0.51}$As at two different ion doses. The damage profiles are obtained from the channeling spectra in Fig. 1 and Fig. 2 by a linear subtraction of the component due to dechanneling. During the initial stage of im-
plantation, some channeling effect do take place in the good crystalline sample, resulting in an overall damage distribution which extends deeper into the substrate. However, when more damage is created (with the increment of ion dose), this effect becomes less significant. Additionally, the disorder peaks are slightly shifted towards the surface due to a channeling effect during analysis where the stopping power for channeled ions is lower than the random case. It is possible that some slight "annealing" occurs in GaAs and AlGaAs on warming up to room temperature. Bearing these effects in mind, the damage profiles for GaAs [Fig. 3(a)] fit reasonably well with the vacancy distribution as calculated by TRIM90. This suggests that insignificant dynamic annealing is taking place during irradiation in GaAs; presumably most of the defects created by ion beam damage are being "frozen in" at LN2 temperatures. However, for Al0.26Ga0.74As and Al0.49Ga0.51As, the damage profiles deviate quite significantly from TRIM90 calculations in regions towards the ion’s end of range [Figs. 3(b), 3(c)]. Thus, the damage profiles are much narrower than that predicted by theory and are confined to the region of maximum energy deposition density (as calculated by TRIM90).

The results for Al0.95Ga0.05As are presented by the RBS-C spectra in Fig. 4. The damage buildup is now quite different to the previous, lower Al content samples. In this case, no damage peak near the region of maximum energy deposition is observed for doses up to 4 × 10^{15} cm^{-2}, consistent with the higher dechanneling level observed in the RBS-C spectra (Fig. 4). Instead, the dechanneling level in the Al0.95Ga0.05As layer increases with increasing dose up to a saturation level. (The 100 Å or so on the surface corresponds to the GaAs capping layer and for all doses shown this layer has been rendered amorphous). Upon further bombardment the disorder peak then grows towards both the substrate and the surface to form a continuous layer of heavily damaged crystal or possibly even amorphous material.

To better understand the different damage behavior in AlGaAs of high Al composition, (110) XTEM was employed on the Al0.95Ga0.05As samples above, corresponding to three different ion doses of 4 × 10^{15}, 5 × 10^{15}, and 8 × 10^{15} cm^{-2} (Fig. 5). At the lowest dose [Fig. 5(a)], a layer (up to ~900 Å in depth) with extensive stacking faults is formed below an amorphous GaAs cap of ~100 Å. This region of stacking faults corresponds closely in depth with the depth of maximum energy deposition (~750 Å from TRIM90). Below this region of stacking faults is a band of mixed amorphous (as confirmed by diffraction pattern) and heavily damaged material, and region of crystalline damage consisting of difficult to identify damage clusters extending towards the end of the ions range distribution. At the next highest dose in Fig. 5(b), four distinct regions are identified:

1. an amorphous GaAs cap layer (~100 Å),
2. region with extensive stacking faults,
3. a band of mixed amorphous (as confirmed by diffraction pattern) and heavily damaged material, and
4. region of crystalline damage consisting of difficult to identify damage clusters extending towards the end of the ions range distribution.
FIG. 5. 220 dark-field XTEM micrographs for samples in Fig. 4. Implant doses are (a) $4 \times 10^{15} \text{ cm}^{-2}$, (b) $5 \times 10^{15} \text{ cm}^{-2}$, and (c) $8 \times 10^{15} \text{ cm}^{-2}$. The surface of each sample is arrowed. All micrographs have the same magnification.

Region (3) is presumably formed from the collapse of the stacking faults into a dense band of defects with further irradiation. This band, upon increasing the ion dose, subsequently “nucleates” a buried amorphous layer which extends towards both the surface and substrate as shown in Fig. 5(c). At this dose of $8 \times 10^{15} \text{ cm}^{-2}$, a continuous amorphous layer is just about to form. However, in this case, islands of crystallites, which are decorated with many extended stacking faults, are observed within the amorphous layer [Fig. 5(c)]. This strongly suggests that the stacking fault regions act as a precursor to the formation of the amorphous phase.

Similar damage behavior is observed in AlAs as shown in Fig. 6. Indeed, we have shown that an amorphous phase can form at LN2 temperature for AlAs as confirmed by the inset selected area diffraction pattern. Also shown is the damage profile of AlAs (linear subtraction as before) which is compared to the vacancy distribution as calculated by TRIM90 for a dose of $8 \times 10^{15} \text{ cm}^{-2}$. The much narrower than predicted damage profile indicates that dynamic annealing is competing very strongly with damage creation during irradiation, particularly in the low damage density wings of the distribution.

The damage buildup data are summarized in Fig. 7(a) for materials of seven different Al compositions. These data are for energy deposition and stopping of 90 keV Si+ wholly within the AlGaAs layers at LN2 temperatures. The $X_{\text{min}}=100\%$ level corresponds to the aligned signal level just reaching the random level at a particular dose. Based on TEM observations, we take this as the threshold dose for amorphization. It is obvious from this figure that the amorphization threshold increases with increasing Al content by more than two orders of magnitude; from $-4 \times 10^{15} \text{ cm}^{-2}$ for GaAs to $-8 \times 10^{15} \text{ cm}^{-2}$ for AlAs. This contrasts with TRIM90 simulations where the peak vacancy production rate for AlAs is only $-35\%$ lower than in GaAs. It is interesting to note that the threshold dose does not vary linearly with Al content but by a power law as shown in Fig. 7(b). For comparison, the data for amorphization threshold from Cullis et al. are plotted onto the same figure. These points are essentially scaled by the peak nuclear energy deposition density from 150 keV Si beams (as used in their experiments) to the threshold needed with 90 keV beams. Although the trend
in the Cullis et al. data is in agreement with our data, the absolute threshold doses do not closely agree. One possible reason is that the dose steps used by Cullis and co-workers were larger, making accurate determination of amorphization threshold dose difficult. Perhaps, more significantly, bi-layer structures were used in their experiment (compared with our thick bulk layers) and the presence of adjacent GaAs or AlGaAs has a dramatic effect on the amorphization threshold as previously indicated. We will report further on multilayer effects elsewhere.

Very little dose rate effect is observed in these bulk AlGaAs materials at LN2 temperatures over one to two orders of magnitude. This is illustrated in Fig. 8(a) for Al0.49Ga0.51As and in Fig. 8(b) for AlAs. However, if the dose rate is high enough to cause local heating, dynamic annealing dominates over damage production, thus resulting in a lower residual damage level. This is true for dose rates exceeding 10 μA/cm² in our case.

It has been reported that GaAs and AlGaAs recrystallize when taken up to room temperature after irradiation at LN2 temperature if a continuous amorphous layer is not formed. To study this phenomenon, in situ TRR was carried out during implantation and subsequent warming up to higher temperatures. Figure 9(a) shows the TRR plot for an Al0.49Ga0.51As sample undergoing this sequence. The ion dose used is 3×10¹⁴ cm⁻² which is sufficient to render the alloy amorphous. Basically, three stages are observed in this reflectivity curve. The first stage (during irradiation) shows that the reflectivity of the material changes in a continuous manner until it reaches a saturation level when an amorphous phase is formed. In the second stage, the spectrum remains essentially “flat” during initial heating. The minor undulations are due to the difference in expansion coefficients of the material and the holder on which it is mounted, causing slight changes to the point of impact of the reflected signal onto the photodetector. The final stage occurs at ~250 °C when there is a sudden dip in the TRR signal suggesting that recrystallization is taking place. This result indicates that no recrystallization whatsoever has taken place at temperatures between LN2 and room temperature. This is also true for AlAs as shown in Fig. 9(b). In this case the ion dose is 2×10¹⁶ cm⁻² which is enough to amorphize the material. The interpretation of these TRR data has been confirmed by channeling and TEM on selected samples.
IV. DISCUSSION

Consistent with other studies,7-11 AlGaAs becomes much more difficult to amorphize with increasing Al content. However, an amorphous phase can be ultimately formed for all Al compositions at LN2 temperatures if the implantation dose is sufficiently high. The discrepancy between TRIM90 calculations, where the peak nuclear energy deposition density is only 35% lower in AlAs compared with GaAs, and the experimental data, where over two orders of magnitude change is observed in the amorphization threshold from GaAs to AlAs, indicates that dynamic annealing is playing a major role in suppressing amorphization in AlGaAs of high Al content. This dynamic annealing effect is clearly illustrated by the comparison between TRIM90 calculations and the damage distributions obtained from RBS-C which show strong dynamic annealing for AlGaAs in the wings of the distribution away from the peak in energy deposition density. The effect is accentuated for alloys with higher Al content. It has been proposed12 that this dynamic annealing behavior is controlled by mobile defects in AlAs (AlGaAs) which mediate in situ annealing during implantation. Our data provide further insight into the annealing processes. The observed power law of the amorphization threshold may suggest a volume-like effect. Also, the insignificant dose rate effect observed over one to two orders of magnitude at LN2 temperatures suggests that dynamic annealing is very fast indeed in AlGaAs compared with the various reports for GaAs during room-temperature bombardment.24,25 These observations, in addition to the interesting damage buildup prior to amorphization for high Al content AlGaAs, suggest that multiple competing processes may be contributing to dynamic annealing and hence suppressing amorphization. For example, we propose that the high Al content greatly facilitates local bonding rearrangement which can act to restore local crystallinity even at low temperatures (somewhat analogous to bonding in metals). As we indicate below, this process, which does not require long range defect migration, can also play a major role in suppressing amorphization.

Interestingly, two distinct damage buildup processes are observed in AlGaAs depending on Al content. Consider, firstly, GaAs irradiated at low temperatures by medium mass ions (such as Si for example). The amorphization process can be best described by the heterogeneous model26 in which the disorder created consists of damage cascades and even small pockets of amorphous zones created by individual ion tracks. Very little dynamic annealing takes place at these temperatures due to the defects being "frozen in." With increasing ion dose, the density of these defects increases and the cascades overlap until a buried continuous amorphous layer is formed. Subsequent bombardment then causes the amorphous layer to increase in thickness, towards both the surface and substrate. In AlGaAs of low Al content the damage buildup appears similar to that of GaAs. However, the main difference between these alloys and GaAs is that, in the former, the defects created can annihilate to some extent to restore crystallinity even at LN2 temperatures. This dynamic annealing effect becomes stronger with increasing Al content (as evident in the increase in amorphization threshold dose with increasing Al content). Amorphization still occurs in these alloys, most probably by a heterogeneous process of overlap of disordered regions, but dynamic annealing results in a higher threshold dose for amorphization.

On the other hand, amorphization of AlGaAs of high Al content proceeds via a homogeneous process27,28 except that the route to amorphization is via the accumulation of extended defects and complexes rather than via the simple accumulation of point defects. In this case, both defect migration and local bonding rearrangements can contribute significantly to suppressing amorphous phase formation. Nevertheless, disorder can build up in the form of stacking faults and other extended defects which are the result of imperfect dynamic annealing. Ultimately a critical damage density level is reached, after which this heavily disordered crystal appears to suddenly "collapse" into an amorphous band with the slightest increment in dose. However, it should be noted that for high Al content AlGaAs, irradiation at LN2 temperatures constitutes a critical temperature regime where damage creation and dynamic annealing processes are closely balanced. In this regime, a slight change in either ion dose (to change the residual disorder level) or temperature can favor one process over the other. This critical effect is observed in certain localized areas of the AlGaAs sample where pockets of crystallinity are maintained within region (3) of Fig. 5(b) and is most probably due to slight inhomogeneity of the ion beam dose and/or a local variation in temperature. On the average, however, region (3) builds

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FIG. 9. In situ TRR spectra during Si irradiation at LN2 temperature and subsequent warming up to higher temperatures for (a) Al0.46Ga0.54As at 3X10^14 cm^-2 and (b) AlAs at 2X10^16 cm^-2. In both cases the dose is sufficient to render the material amorphous.
up as a band of highly defective crystal with increasing dose. This band then acts as a nucleating site for forming and extending the amorphous phase with further irradiation. The sudden transition to an amorphous phase can be best explained by referring to Fig. 10 which shows schematically the free-energy differences between amorphous and crystalline phases in metals and semiconductors. The amorphous phase has a higher free-energy than the crystalline phase. However, the amorphous phase is very unstable in pure metals due to the highly nondirectional bonding nature of such elements. Hence, there is a strong tendency for displaced atoms in metals to locally rearrange to restore crystalline bond angles and preserve local crystallinity. In doing so this lowers the free-energy close to that of the crystal. For a highly covalent material such as Si, there is a high energy barrier (in part due to strong angular bonding rigidity) which inhibits a return to crystallinity. As in Si, there is also a kinetic barrier between the amorphous and crystalline phases for AlGaAs (GaAs and AlAs). However, in the case of AlAs and high Al content AlGaAs, we suggest that local bonding rigidity may be lower than GaAs (more metal-like bonds), allowing displaced atoms to more easily rearrange and preserve crystallinity under ion irradiation. This process may complement higher defect mobility in AlGaAs and make high Al content AlGaAs difficult to amorphize. However, the imperfect dynamic annealing leads to build up of defects and the free-energy of the AlGaAs crystal will increase due in part to strain energy associated with ion damage. Amorphization is then achieved when the free-energy of the highly defective material reaches a critical limit which may even exceed that of the amorphous phase, as illustrated in Fig. 10. At this stage, the defective crystal may “collapse” (relax) into an amorphous state, thereby lowering its free-energy (by relieving strain).

Once a complete amorphous phase is formed, the activation energy for crystallization must now be overcome to partially or completely recover the crystalline structure. This can be achieved by subjecting the amorphous material to heat treatment. No recrystallization is observed during warming up of completely amorphous AlAs to room temperature or even to ~100 °C (from TRR data). However, if a complete amorphous phase is not created, pockets of amorphous AlAs material may partially recrystallize when brought up to room temperature, presumably due to isolated crystalline regions within the amorphous phase acting as seeds for recovery [note the TEM micrograph in Fig. 5(c) which indicates such crystalline regions]. This phenomenon of partial crystallization on warming up is also reported by Turkot et al. This process is well known in Si where continuous amorphous layers recrystallize above 500 °C but isolated amorphous zones in crystalline material can anneal out at much lower temperatures. Our annealing data for amorphous AlGaAs also indicate that, as the Al content increases, AlGaAs becomes more difficult to recrystallize epitaxially and the quality of the recrystallized layer contains an increasing amount of extended defects.

It is yet to be determined at what Al composition the transition between the above two damage buildup and amorphization processes takes place. Jencic et al. reported that no distinct amorphous zones are created in Al0.55Ga0.45As irradiated at 30 K with 50 keV or 1.5 MeV Xe ions. The only visible damage is some weak-contrast features, presumably clusters of point defects similar to that observed in AlAs by Cullis et al. Our data indicate also that the transition occurs at around this Al concentration. For higher Al concentrations, the lack of amorphous regions at intermediate doses of ~10^14–10^15 cm^-2 can lead one to argue that the dynamic annealing rate is much higher than the rate of defect production. It might be expected that amorphization would then only proceed by increasing the implant dose rate such that damage cascades overlap with existing disorder zones before defects within these zones have fully annealed. This situation may lead to a stable disordered structure and to the formation of an amorphous phase. Hence, the amorphization process in this case would be similar to that of lower Al content AlGaAs except that the high dose rates would be necessary to induce amorphization. However, our data do not support this scenario; there is no dose rate effect and only extended defects and clusters are observed at preamorphous doses. This mode of disorder buildup and the sudden onset of amorphization indicate a distinctly different amorphization mechanism to that of lower Al concentrations. Indeed, our data (particularly the sudden onset of amorphization) suggest that amorphization is “nucleation limited” for high Al content cases, whereby the free-energy of the defective material prior to amorphization actually exceeds that of the amorphous phase (see Fig. 10). When this occurs and the lattice collapses to a lower free-energy, a large volume of material is amorphized for a very small increment in dose. Thereafter the buried amorphous layer extends both towards the surface and into the bulk. This (nucleation-limited amorphization) behavior is similar to that previously observed for silicon.

In most of the previous ion damage work on AlGaAs (AlAs) by various groups, multilayer structures were used. In such cases, the presence of GaAs adjacent to AlGaAs (AlAs) complicates the damaging and annealing processes in both GaAs and AlGaAs (AlAs) and thus the bulk behavior of AlGaAs is obscured. For example, both Klatt and co-workers and Turkot et al. observed asymmetric damage distributions in AlAs sandwiched between two GaAs...
layers and explained this behavior in terms of different damage states existing after irradiation and warm up to room temperature. Cullis et al.\textsuperscript{7,12,13} showed that AlGaAs and AlAs are able to offer some degree of protection against ion-induced damage to adjacent GaAs regions and also to act as a sink for point defects. Multiple competing processes are clearly taking place in GaAs-AlGaAs (AlAs) multilayer structures during ion irradiation. We observed further interesting effects in multilayers which indicate that mobile defects in AlGaAs can both inhibit amorphization to surrounding GaAs layer or enhance amorphization depending on the particular bombardment conditions and nature of the multilayer structures. However, this is the subject of another study and will be reported elsewhere.\textsuperscript{21}

\section*{V. CONCLUSIONS}

Consistent with other studies,\textsuperscript{7-11} our systematic data clearly indicate that AlGaAs is more difficult to amorphize the higher the Al content. More than two orders of magnitude difference is observed in the amorphization threshold dose between GaAs and AlAs. During implantation, dynamic annealing, as controlled by very fast mobile defects and bonding rearrangements, and defect creation processes compete very strongly in AlGaAs even at LN2 temperatures. This effect is accentuated with increasing Al composition. Two distinct damage buildup processes are observed in Al$_x$Ga$_{1-x}$As depending on the Al content. For lower Al content ($x=0.83$), the behavior is similar to GaAs whereby amorphization is via the overlap of damage cascades or even small amorphous zones created by individual ion tracks. For AlGaAs of high Al content ($x>0.95$), disorder builds up in the form of stacking faults and other extended defects until a critical defect density (or system free-energy) is reached. This strong dynamic annealing behavior can be mediated by both the availability of very fast mobile defects and local bonding rearrangement (as a result of high Al content) to restore crystallinity. Nucleation-limited amorphization then proceeds and is best described by a model in which the free energy of the defective crystal exceeds that of the amorphous state.

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