AIP Applied Physics Letters

Implantation-produced structural damage in In x Ga 1−x N

S. O. Kucheyev, J. S. Williams, J. Zou, S. J. Pearton, and Y. Nakagawa

Citation: Applied Physics Letters **79**, 602 (2001); doi: 10.1063/1.1388881

View online: http://dx.doi.org/10.1063/1.1388881

View Table of Contents: http://scitation.aip.org/content/aip/journal/apl/79/5?ver=pdfcov

Published by the AIP Publishing

Implantation-produced structural damage in $In_xGa_{1-x}N$

S. O. Kucheyev[a)] and J. S. Williams
Department of Electronic Materials Engineering, Research School of Physical Sciences and Engineering, The Australian National University, Canberra, ACT 0200, Australia

J. Zou
Australian Key Center for Microscopy and Microanalysis, The University of Sydney, Sydney, NSW 2006, Australia

S. J. Pearton
Department of Materials Science and Engineering, University of Florida, Gainesville, Florida 32611

Y. Nakagawa
Nichia Chemical Industries, Anan, Tokushima 774, Japan

(Received 23 March 2001; accepted for publication 1 June 2001)

The influence of In content on the accumulation of structural damage in $In_xGa_{1-x}N$ films (with $x = 0.0-0.2$) under heavy-ion bombardment is studied by a combination of Rutherford backscattering/channeling spectrometry and transmission electron microscopy. Results show that an increase in In concentration strongly suppresses dynamic annealing processes and, hence, enhances the buildup of stable lattice disorder in InGaN under ion bombardment, A comparison of the damage buildup behavior and defect microstructure in InGaN with those in GaN is presented. Results of this study may have significant technological implications for estimation and control of implantation-produced damage in InGaN/GaN heterostructures. © *2001 American Institute of Physics.* [DOI: 10.1063/1.1388881]

The family of III-nitrides (in particular, GaN, AlGaN, and InGaN) is currently a subject of extensive research. Significant interest in these materials is driven by important technological applications of III-nitrides in the fabrication of a range of (opto)electronic devices, as has been discussed in detail in a number of recent reviews.[1] In the fabrication of such devices, ion implantation is a very attractive technological tool. However, ion implantation inevitably produces lattice disorder. Because ion-beam-produced damage affects all the properties of the material under bombardment, studies of implantation disorder are technologically important.

Given such technological importance, considerable work has recently been done to understand ion-beam-damage processes in GaN.[2] However, ion-beam-produced structural damage in AlGaN and InGaN has not been studied in any detail. These materials are of great importance because of their role as active and cladding layers in optoelectronic devices.[3,4] Hence, in this letter, we report on studies of the influence of In content (x) on the buildup of implantation-produced damage in $In_xGa_{1-x}N$. Interestingly, results show that an increase in x suppresses dynamic annealing processes and enhances the accumulation of stable lattice damage during ion bombardment.

About 1400-Å-thick $In_xGa_{1-x}N$ films ($x = 0.00$, 0.03, 0.07, 0.12, 0.16, 0.18, and 0.20, with $\Delta x = 0.01$, as measured by Rutherford backscattering) on the top of ~2-μm-thick wurtzite undoped GaN epilayers were grown on *c*-plane sapphire substrates by metalorganic chemical vapor deposition in a rotating disk reactor at Nichia Corp. Samples were implanted with 300 keV $^{197}Au^+$ ions at 20 °C with a beam flux of $\sim 3.1 \times 10^{12}\ cm^{-2}\ s^{-1}$ over the dose range from 4×10^{13} to $2 \times 10^{15}\ cm^{-2}$ using an ANU 1.7 MV tandem accelerator (NEC, 5SDH-4). During implantation, samples were tilted by ~7° relative to the incident ion beam to minimize channeling.

After implantation, all samples were characterized *ex situ* by Rutherford backscattering/channeling (RBS/C) spectrometry using an ANU 1.7 MV tandem accelerator (NEC, 5SDH) with 1.8 MeV $^4He^+$ or 3.3 MeV $^4He^+$ ions incident along the [0001] direction and backscattered into detectors at 98° and 110° relative to the incident beam direction. The 8° glancing-angle detector geometry with a 1.8 MeV $^4He^+$ ion beam was used to provide enhanced depth resolution for examining near-surface damage accumulation, while the 20° glancing-angle detector geometry with 3.3 MeV $^4He^+$ ions was used to separate In and Ga peaks in the RBS/C spectra to study the accumulation of In and Ga displaced atoms separately. RBS/C spectra acquired using 3.3 MeV He ions were analyzed using one of the conventional algorithms[5] for extracting depth profiles of the effective number of scattering centers. For brevity, such a number of scattering centers, normalized to the atomic concentration, will be referred to below as "relative disorder." Selected samples were also studied by cross-sectional transmission electron microscopy (XTEM) in a Philips CM12 transmission electron microscope operating at 120 keV. XTEM specimens were prepared by 3 keV Ar^+ ion-beam thinning using a Gatan precision ion-polishing system.

Figure 1 shows RBS/C spectra illustrating lattice damage in $In_xGa_{1-x}N$ films bombarded at 20 °C with 300 keV Au ions to a dose of $4 \times 10^{13}\ cm^{-2}$. It is clearly seen from Fig. 1 that the level of implantation-produced lattice damage increases with increasing In concentration. It should be noted

a)Electronic mail: sergei.kucheyev@anu.edu.au

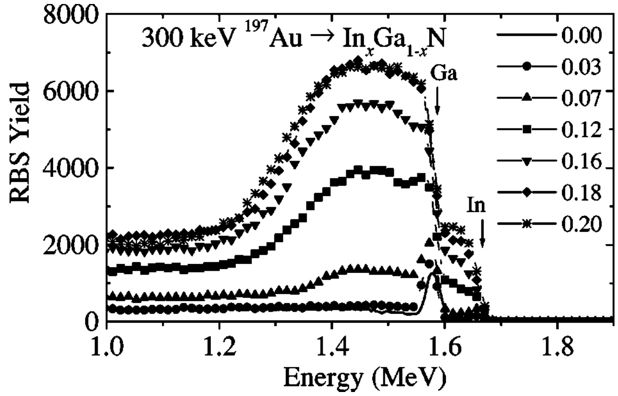

FIG. 1. RBS/C spectra (acquired using 1.8 MeV He ions) illustrating lattice damage in $In_xGa_{1-x}N$ films bombarded at 20 °C with 300 keV Au ions to a dose of 4×10^{13} cm^{-2} with a beam flux of $\sim3.1\times10^{12}$ $cm^{-2}\,s^{-1}$. Indium content (x) in different samples is indicated in the legend. The positions of the surface peaks of In and Ga are shown by arrows.

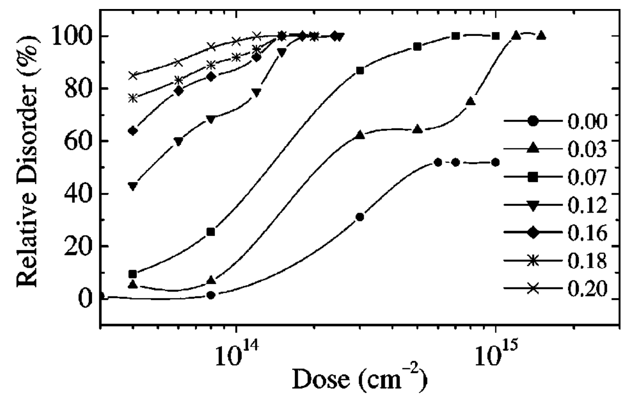

FIG. 3. The dose dependence of relative disorder (extracted from RBS/C data for the Ga peak) in the bulk defect peak region for $In_xGa_{1-x}N$ samples bombarded with 300 keV Au ions at 20 °C with a beam flux of $\sim3.1\times10^{12}$ $cm^{-2}\,s^{-1}$. Indium content (x) in different samples is given in the legend.

that, for a given dose, an increase in relative disorder with increasing In content is actually larger than that suggested by Fig. 1 because the RBS random level of Ga (which is used to normalize damage) decreases with increasing In concentration. Figure 1 also shows that the damage buildup in InGaN does not result in a bimodal distribution with a strong surface disorder peak, in contrast to the situation for GaN.[2] Indeed, even in the sample with the lowest In concentration studied ($In_{0.03}Ga_{0.97}N$), surface disordering is not strong.[6]

Typical RBS/C spectra illustrating the buildup of lattice disorder in $In_xGa_{1-x}N$ with increasing dose of 300 keV Au ions at 20 °C are shown in Fig. 2 for the case of $x=0.12$. In these spectra, acquired using 3.3 MeV He ions, the Ga and In peaks are well separated, which allows the accumulation of Ga and In displaced atoms to be monitored separately. Figure 2 shows that, with increasing ion dose, the number of In and Ga displaced atoms accumulates at similar rates for both species. Moreover, an analysis of RBS/C data reveals that, in all InGaN samples studied, this accumulation of Ga and In disorder proceeds at essentially the same rate. Indeed, for a given dose, RBS/C yields (normalized to random levels) for Ga and In peaks are the same (within experimental error) in all InGaN samples studied. Hence, the reduced efficiency of dynamic annealing in InGaN appears not to be caused by reduced effective diffusivities of only Ga or In interstitials.

Figure 3 summarizes RBS/C data for all samples studied, showing the dose dependence of relative disorder (extracted from RBS/C data for the Ga peak) at depths corresponding to the bulk defect peak region for $In_xGa_{1-x}N$ samples with different x bombarded with 300 keV Au ions at 20 °C. It is clearly seen from Fig. 3 that an increase in In concentration strongly enhances implantation-produced disorder. For example, Fig. 3 shows that, for a dose of 4×10^{13} cm^{-2}, the level of relative disorder in $In_xGa_{1-x}N$ samples dramatically increases from ~2% to ~85% with increasing x from 0.0 to 0.2. This behavior is consistent with a previous report comparing the degree of dynamic annealing in different semiconductors, including GaN and InN.[7] In addition, Fig. 3 reveals that an increase in In concentration apparently suppresses the effect of damage saturation in the crystal bulk. Such defect saturation is very pronounced in GaN under heavy-ion bombardment at room temperature (see Ref. 2 and Fig. 3). Figure 3, however, shows that the effect of damage saturation is significantly less pronounced in InGaN samples, with apparent saturation levels increasing with an increase in x.

In selected samples, ion-beam-produced defects have been studied by XTEM. Figures 4(a) and 4(b) show dark-field XTEM images of $In_{0.16}Ga_{0.84}N$ implanted at 20 °C with 300 keV Au ions to a dose of 4×10^{13} cm^{-2}. These XTEM images reveal that defect structures in ion implanted InGaN

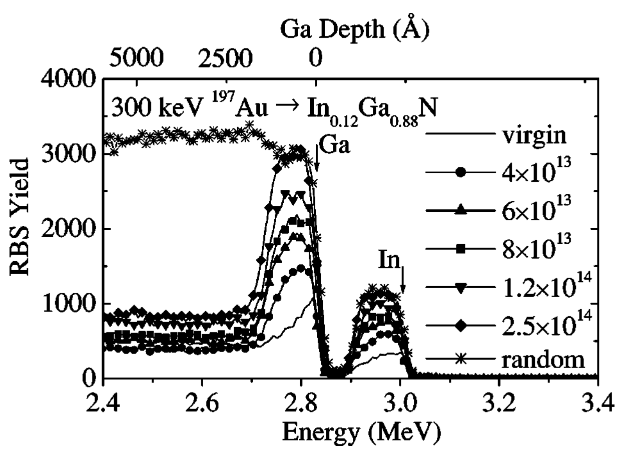

FIG. 2. RBS/C spectra (acquired using 3.3 MeV He ions) showing the damage buildup for 300 keV Au ion bombardment of $In_{0.12}Ga_{0.88}N$ at 20 °C with a beam flux of $\sim3.1\times10^{12}$ $cm^{-2}\,s^{-1}$. Implantation doses (in cm^{-2}) are indicated in the figure. The positions of the surface peaks of In and Ga are shown by arrows.

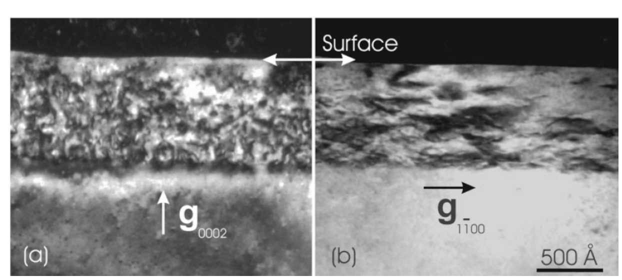

FIG. 4. Dark-field XTEM images [(a) $\mathbf{g}=0002^*$ and (b) $\mathbf{g}=1\bar{1}00^*$] of an $In_{0.16}Ga_{0.84}N$ epilayer bombarded at 20 °C with 300 keV Au ions with a beam flux of $\sim3.1\times10^{12}$ $cm^{-2}\,s^{-1}$ to a dose of 4×10^{13} cm^{-2}. Images (a) and (b) are of the same magnification.

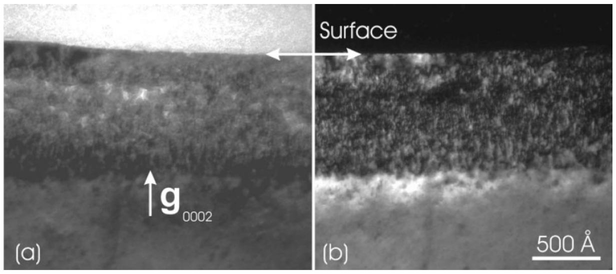

FIG. 5. Bright-field (a) and dark-field (b) XTEM images (**g**=0002*) of an $In_{0.18}Ga_{0.82}N$ epilayer bombarded at 20 °C with 300 keV Au ions with a beam flux of $\sim 3.1\times 10^{12}$ cm^{-2} s^{-1} to a dose of 2×10^{14} cm^{-2}. Images (a) and (b) are of the same magnification.

appear to be similar to those previously reported for GaN bombarded with heavy ions.[2] Indeed, some coarse defect clusters [see Fig. 4(a)] and a band of extended (planar) defects [see Fig. 4(b)] are observed in ion implanted GaN. However, Fig. 4(b) shows that, in the case of InGaN, implantation-produced planar defects seem to be coarser and less regular than those in ion implanted GaN. This observation is consistent with the fact that, as discussed above, dynamic annealing processes in InGaN are suppressed as compared to those in GaN. Indeed, it is dynamic annealing which is responsible for the formation of extended defects during ion bombardment.

It is also interesting that a rather sharp border of implantation-produced damage is seen in Fig. 4, where extended defects are observed only within the top InGaN film, while the underlying GaN epilayer is free from such extended defects. This is also consistent with the fact that dynamic annealing processes in GaN are extremely efficient, which results in a considerably lower level of lattice disorder in GaN as compared to that in InGaN.

Finally, Figs. 5(a) and 5(b) show bright- and dark-field XTEM images of an $In_{0.18}Ga_{0.82}N$ film implanted at 20 °C with 300 keV Au ions to a dose of 2×10^{14} cm^{-2}. These images clearly illustrate that, even in the case of relatively high dose ion bombardment, when the RBS/C yield reaches the random level (see Fig. 3), the implanted InGaN film may not be completely disordered (i.e., amorphous). Rather, XTEM strongly suggests that the $In_{0.18}Ga_{0.82}N$ film shown in Fig. 5 consists of amorphous regions embedded into a heavily damaged (but still crystalline) matrix. Hence, under heavy-ion bombardment at room temperature, the amorphization behavior of InGaN is markedly different from that of GaN. Indeed, in GaN, under room-temperature heavy-ion bombardment, amorphization proceeds layer-by-layer from the surface.[2] It should be noted that amorphization which proceeds via the formation of amorphous regions embedded into a heavily damaged crystalline matrix (such as shown in Fig. 5) is typical for materials in which dynamic annealing, although being present, is not significant compared with the rate of accumulation of stable lattice disorder.[8]

In conclusion, the structural characteristics of $In_xGa_{1-x}N$ layers bombarded with keV heavy ions have been studied by RBS/C and XTEM. Results show that an increase in In content (i) suppresses dynamic annealing processes, (ii) enhances the accumulation of stable lattice defects, and (iii) weakens the effect of damage saturation in the bulk during room temperature bombardment. In addition, we have found that, with increasing ion dose, the number of In and Ga displaced atoms accumulates at essentially the same rate (within experimental error). XTEM has revealed similar defect structures in ion implanted InGaN as compared to those in GaN. However, extended defects appear to be coarser and less regular in InGaN than those in GaN, which is consistent with a reduced level of dynamic annealing in InGaN.

[1] See, for example, S. J. Pearton, J. C. Zolper, R. J. Shul, and F. Ren, J. Appl. Phys. **86**, 1 (1999); S. C. Jain, M. Willander, J. Narayan, and R. Van Overstraeten, *ibid.* **87**, 965 (2000), and references therein.
[2] See, for example, B. Rauschenbach, in *III-Nitride Semiconductors: Electrical, Structural and Defects Properties*, edited by M. O. Manasreh (Elsevier, Amsterdam, 2000); S. O. Kucheyev, J. S. Williams, and S. J. Pearton, Mater. Sci. Eng., R. **33**, 51 (2001).
[3] H. Morkoç, *Nitride Semiconductors and Devices* (Springer, Berlin, 1999).
[4] S. Nakamura, in *Properties, Processing and Applications of Gallium Nitride and Related Semiconductors*, edited by J. H. Edgar, S. Strite, I. Akasaki, H. Amano, and C. Wetzel (INSPEC–IEE, London, 1999).
[5] K. Schmid, Radiat. Eff. **17**, 201 (1973).
[6] It should be noted that somewhat pronounced surface defect peaks observed in RBS/C spectra of $In_{0.03}Ga_{0.97}N$ and $In_{0.07}Ga_{0.93}N$ samples shown in Fig. 1 are most likely due to an overlap of ion-beam-produced bulk and intrinsic (i.e., present in as-grown samples) surface RBS/C peaks. Indeed, for larger ion doses and, hence, for higher levels of lattice disorder, no preferential surface disordering has been revealed, and RBS/C spectra are unimodal even for samples with low In content.
[7] J. S. Williams, Mater. Sci. Eng., A **253**, 8 (1998).
[8] See, for example, J. F. Gibbons, Proc. IEEE **60**, 1062 (1972).