Comparison of low- and room-temperature damage formation in Ar ion implanted GaN and ZnO

E. Wendler a,*, W. Wesch a, A.Yu. Azarov b, N. Catarino c, A. Redondo-Cubero c, E. Alves c, K. Lorenz c

aFriedrich-Schiller-Universität Jena, Institut für Festkörperphysik, Max-Wien-Platz 1, D-07743 Jena, Germany
bUniversity of Oslo, The Department of Physics, P.O. Box 1048 Blindern, NO-0316 Oslo, Norway
cIST/ITT-INstituto Tecnológico e Nuclear, Instituto Superior Técnico,Universidade Técnica de Lisboa, Estrada Nacional 10, P-2686-953 Sacavém, Portugal

1. Introduction

GaN and ZnO are wide bandgap semiconductors suitable for applications in optoelectronics and high temperature, high frequency and high power devices. Ion implantation is a well established technique for device fabrication. Prototypes of GaN-based field effect transistors and LEDs employing processing by ion implantation were reported [1]. However, despite their strong radiation resistance (see [2,3] and references therein) ion implantation is not yet routinely applied in these materials. Therefore, further work is necessary to better understand the mechanisms of ion induced damage formation and the effects of the implanted ions to finally find out suitable implantation and annealing conditions for certain applications. Furthermore, the study of ion-beam induced damage formation is also of scientific interest, because the origin of the high radiation resistance of these materials is still not fully understood.

In this paper new data on damage formation at 295 K in Ar implanted GaN and ZnO are presented. The aim of the paper is to compare the corresponding results with those obtained previously for Ar implantation in these materials performed at 15 K [4–6]. This comparison reveals that in both materials the effect of the target temperature during implantation is rather small. From that, in turn, it can be concluded that the high radiation resistance of GaN and ZnO is not predominantly related to thermally enhanced processes.

2. Experimental conditions

(0001) Oriented GaN and ZnO were implanted with 300 keV and 200 keV Ar ions, respectively. Implantation was performed 7° off-axis in order to minimise channelling effects. Rutherford backscattering spectrometry (RBS) in channelling configuration with 1.4 MeV He ions and a backscattering angle of 170° was used to register the damage formation. To quantify the damage produced, we use the difference in minimum yield \( \Delta Y_{\text{min}} \) versus depth \( z \) which is calculated from the yield \( Y \) of backscattered ions in aligned (al) and random (ra) direction of virgin (vir) and implanted (impl) samples \( \Delta Y_{\text{min}} = (Y_{\text{impl}} - Y_{\text{vir}}) / Y_{\text{vir}} \). RBS analysis was performed quasi-in situ, i.e. each implantation step was followed by an immediate measurement, whilst temperature and environment of the samples did not change (for details of the experimental setup see [7]). For both materials, implantation series are performed at 15 and 295 K and the ion fluence \( N_i \) varies between \( 2 \times 10^{12} \) and \( 7 \times 10^{16} \) cm\(^{-2}\). For comparison the number of displacements per implanted ion and unit depth, \( N_{\text{disp}} \), is calculated with the code SRIM (version...
2008, 04) [8]. The density of GaN and ZnO is taken to be 6.15 g/cm³ and 5.609 g/cm³ corresponding to an atomic density of N₀ = 8.846 × 10²² at./cm³ and 8.295 × 10²² at./cm³, respectively. The displacement energies are taken to be 45 eV for Ga, 109 eV for N in GaN [9] and 65 eV for Zn, 50 eV for O in ZnO (see[5] and references therein). A cross section σ_{SRIM} representing the primary displacements of lattice atoms is calculated by σ_{SRIM} = N_{d}/N₀ (see Table 1).

3. Results and discussion

Figs. 1 and 2 show the difference in minimum yield Δχ/Δz versus depth z for various ion fluences N_I implanted in GaN and ZnO, respectively. In both materials independent of temperature the Δχ/Δz spectra exhibit a similar shape over a wide range of ion fluences. This shape is characterised by a certain contribution of direct backscattering (appearing as a peak) and a relatively high dechannelling background. It was shown previously that this characteristic shape can be explained assuming the existence of both defect clusters (characterised by randomly displaced lattice atoms) and extended defects (characterised by correlated displaced lattice atoms) [6]. The same conclusion was drawn when analysing RBS channelling spectra of ion implanted GaN with the help of Monte Carlo calculations. In that study the existence of extended defects is taken into account by a bent-channel model [10]. It is shown that for GaN both extended defects and randomly displaced lattice atoms have to be assumed in order to represent the measured spectra [10,11]. Furthermore, for implantation at room temperature, the existence of extended defects was proven by transmission electron microscopy (TEM) in GaN [11,12] and ZnO [13]. Therefore, the similarity in the shape of the RBS spectra at 15 K and at 295 K indicates that extended defects do form not only during implantation at room temperature but also at 15 K. At this temperature the influence of thermal effects can be widely excluded. Therefore, our results suggest that the formation of extended defects is not driven by the thermal mobility of intrinsic defects.

Only for implantation of GaN at 15 K with ion fluences N_I > 1.2 × 10¹⁵ cm⁻² the channelling spectra reach the random level corresponding to Δχ/Δz ≈ 1 (see Fig. 1a). This is commonly taken as an indication for amorphisation. This state is not obtained for ZnO up to the highest fluence applied of 7 × 10¹⁵ cm⁻². This demonstrates that also at 15 K ZnO is more radiation resistant than GaN.

In Fig. 3, the difference in minimum yield Δχ/Δz taken at the depth of maximum energy deposition in the displacement of lattice atoms is plotted versus the ion fluence N_I. Amorphisation of GaN at 15 K proceeds via five steps (see also [4]). In GaN at 295 K and in ZnO no amorphisation is observed and damage saturation occurs in four stages. Here we focus on these stages of damage formation and will not discuss possible mechanisms of amorphisation in GaN at 15 K.

Stages I–IV are characterised by an increase followed by a plateau-like nearly constant value of Δχ/Δz. Commonly, the first two stages are attributed to the formation and recombination of point defects and the latter two stages can be interpreted as to

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Table 1

<table>
<thead>
<tr>
<th>Parameter</th>
<th>GaN</th>
<th>ZnO</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>15 K</td>
<td>295 K</td>
</tr>
<tr>
<td>σ₉ (cm²)</td>
<td>9.5 × 10⁻¹⁵</td>
<td>8.5 × 10⁻¹⁶</td>
</tr>
<tr>
<td>σᵣ (cm²)</td>
<td>1.4 × 10⁻¹⁴</td>
<td>1.95 × 10⁻¹⁶</td>
</tr>
<tr>
<td>σ_{SRIM} (cm²)</td>
<td>7 × 10⁻¹⁶</td>
<td>8.5 × 10⁻¹⁶</td>
</tr>
</tbody>
</table>

Fig. 1. Difference in minimum yield Δχ/Δz versus depth z for Ar ion implanted GaN with implantation and subsequent measurement being performed at 15 K (a) and at 295 K (b). Solid lines are results for intermediate fluences which are not included in the legend for clarity purposes.

Fig. 2. Difference in minimum yield Δχ/Δz versus depth z for Ar ion implanted ZnO with implantation and subsequent measurement being performed at 15 K (a) and at 295 K (b). Solid lines are results for intermediate fluences which are not included in the legend for clarity purposes.

Fig. 4 shows the results for the first two stages of damage formation in a different scale to highlight the changes in this regime.
The lines in Fig. 4 are fitted to the experimental data assuming production and recombination of point defects with the cross sections \( \sigma_p \) and \( \sigma_r \), respectively, which yields [15]

\[
\Delta \chi_{\text{min}} = \frac{\sigma_p}{\sigma_r} (1 - \exp(-\sigma_r N_1)).
\]

The values of \( \sigma_p \) and \( \sigma_r \) used to calculate the curves in Fig. 4 are given in Table 1. In GaN independent of temperature the first increase of damage formation occurs for \( N_1 < 1.5 \times 10^{14} \text{ cm}^{-2} \). Above this value, \( \Delta \chi_{\text{min}} \) tends to saturate at a value which is lower at the higher temperature of implantation. This meets the expectation that at higher temperatures a stronger recombination of defects occurs (see \( \sigma_p \) in Table 1). In ZnO at 15 K the first increase is observed for ion fluences below \( 5 \times 10^{13} \text{ cm}^{-2} \). This fluence is lower than that for GaN which again reveals the higher radiation resistance of ZnO since saturation occurs at lower fluences and lower damage levels. No clear statement can be made for implantation at 295 K in this material because of the statistical fluctuations of the data in the corresponding range of ion fluences. Here further experiments are needed.

Table 1 shows that for GaN the cross section of point defect formation \( \sigma_p \) at 15 K and at 295 K are about the same and rather close to the calculated \( \sigma_{\text{SRIM}} \). The latter is true for ZnO implanted at 15 K, too (see Table 1 and [5]). When starting from the definition of \( \sigma_{\text{SRIM}} \) (see Section 2) it is obvious that the cross section of point defect formation is the number of displacements per lattice atom produced by a single ion multiplied with the area (projected to the surface) damaged by one ion. Assuming the number of displacements per lattice atom to be unity means that all atoms are displaced and an amorphous cluster is produced. With this assumption the area damaged by one ion and – assuming a circular shape – the diameter of the cluster can be estimated from \( \sigma_{\text{SRIM}} \). With \( \sigma_{\text{SRIM}} \approx 10^{-15} \text{ cm}^2 \) a diameter of 0.3–0.4 nm is obtained. This is in the order of the distance between nearest neighbours and means that only very few atoms would be involved for which an amorphous cluster cannot be defined. Consequently, the assumption that each atom is displaced cannot be correct, which means that amorphous clusters are not produced. This estimation confirms the assumption mentioned above that within a single ion impact only point defects are produced.

When the impacts of individual ions start to overlap, Eq. (1) assumes the recombination of point defects and their concentration saturates. This yields a constant value of \( \Delta \chi_{\text{min}} \) at higher ion fluences referred to as stage II. However, the defect profiles within the first plateau tend to expand to deeper regions of the sample as discussed for GaN implanted at 15 K in [4] and \( \Delta \chi_{\text{min}} \) is not absolutely constant. Therefore we suspect that the occurring processes are more complex than just being a balance between formation and recombination of point defects. For room temperature implantation of GaN the formation of stacking faults was shown within stage II by TEM [11,12]. Further, it could be shown for GaN implanted with 320 keV Ar ions at 295 K with \( 5 \times 10^{14} \text{ cm}^{-2} \) (which is within the first plateau in Fig. 4) that despite a large amount of extended defects detected by TEM, the representation of the RBS spectrum requires the assumption of a substantial concentration of randomly displaced lattice atoms due to point defects and irregular defect clusters [11,16]. To summarise these observations, it is obvious that independently of temperature, within the first plateau the overlapping of the collision cascades induces point defects.

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Within stage IV almost the same ion fluence for both temperatures (see Fig. 3). In GaN and ZnO point defects recombine and start to form extended defects. So far molecular dynamics simulations of ion irradiation of GaN [9] do not reproduce this behaviour whereas damage formation in GaAs ion implanted at low temperatures can be well described with that technique [18]. Obviously in GaN and ZnO the formation of extended defects is energetically favourable in comparison to the formation of larger randomly ordered agglomerates of point defects which may transform into the amorphous state.

In GaN and ZnO, the third stage of damage formation occurs at almost the same ion fluence for both temperatures (see Fig. 3). Within stage IV $\Delta \chi_{\text{min}}$ again reaches a saturation value which does not significantly depend on the temperature during implantation. In ZnO the saturation value tends to be higher at the higher temperature. This finding may be caused by the enhanced dechanneling of the analysing He ions at 295 K due to the stronger thermal vibration of the lattice atoms. It may also indicate a more efficient formation of extended defects at this temperature than at 15 K (cp. results for $N_i = 3 \times 10^{16}$ cm$^{-2}$ in Fig. 2a and b). However, from the existing data a final statement cannot be made. By TEM on GaN implanted at room temperature it was shown that within stage IV the increasing density of stacking faults leads to the formation of dislocations that coalesce with increasing ion fluence [11]. Also for implanted ZnO TEM reveals the formation of dislocation loops at 275 K [13].

Now the question arises what drives the transition from stage II to stage III. A closer look at the depth distribution of damage shows that the onset of the damage peak within the second stage occurs at the depth where the energy is deposited in the displacement of lattice atoms (see [4]). That means the driving force is not directly connected with the implanted ions themselves. During a further increase of the ion fluence within stage III the damage extends especially towards depths where the implanted ions come to rest (see Figs. 1 and 2 and [4,6]). In Refs. [11,12] X-ray diffraction measurements are reported for GaN implanted at room temperature, which reveal a continuous increase of tensile strain along the c-axis.

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4. Summary

GaN and ZnO are implanted at 295 K with 300 keV Ar and 200 keV Ar ions. Damage analysis is performed with RBS in channeling configuration quasi-in situ at the respective temperature. The difference in minimum yield is taken as a measure of the amount of damage produced and its dependence on ion fluence and depth is discussed. The results obtained are compared with those shown previously for Ar ion implantation at 15 K. The most striking result is that in GaN and ZnO the ion-induced damage formation is only weakly influenced by the implantation temperature. For the discussion of our findings, results on damage formation in these materials obtained by other authors applying TEM are taken into account. Eventually it can be concluded that in these materials extended defects form not only at room temperature but also at a temperature of 15 K. This clearly suggests that the formation of extended defects is not driven by the thermal mobility of point defects. It is supposed that in GaN and ZnO damage-induced strain plays a dominant role and the formation of extended defects seems to be energetically favourable in comparison to the formation of larger randomly ordered agglomerates of defects.

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