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# Radiation effects in GaN materials and devices

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This article reviews the effects of radiation damage on GaN materials and devices such as light-emitting diodes and high electron mobility transistors. Protons, electrons and gamma rays typically produce point defects in GaN, in contrast to neutron damage which is dominated by more extended disordered regions. Regardless of the type of radiation, the electrical conductivity of the GaN is reduced through the introduction of trap states with thermal ionization energies deep in the forbidden bandgap. An important practical parameter is the carrier removal rate for each type of radiation since this determines the dose at which device degradation will occur. Many studies have shown that GaN is several orders of magnitude more resistant to radiation damage than GaAs, *i.e.* it can withstand radiation doses of at least two orders of magnitude higher than those degrading GaAs with a similar doping level. Many issues still have to be addressed. Among them are the strong asymmetry in carrier removal rates in n- and p-type GaN and interaction of radiation defects with Mg acceptors and the poor understanding of interaction of radiation defects in doped nitrides with the dislocations always present.

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## I Introduction

GaN based devices have significant market applications in lighting, displays and power electronics. AlGaN light emitting diodes (LEDs) and laser diodes (LDs) are the predominant semiconducting light sources for the UV/green/blue range. AlGaN/GaN High Electron Mobility Transistors (HEMTs) and GaN diodes are well-suited to high-power/high-frequency/high-temperature electronic applications, while AlGaN/GaN UV photodetectors have matured in recent years. Many of these devices are used in satellite and military systems where radiation tolerance is critical. The Earth's magnetosphere is bombarded by a nearly isotropic flux of energetic charged particles – 85% protons, 14% α-particles, and 1% heavier ions covering the full range of elements. There are also the van Allen belts, an inner belt extending to 2.5 Earth radii and comprising energetic protons up to 600 MeV together with electrons up to several MeV, and an outer belt comprising mainly electrons extending out to 10 Earth radii. In the years around solar maximum, the sun is an additional sporadic source of lower energy particles accelerated during certain solar flares and/or in the subsequent coronal mass ejections. These solar particle events last for up to several days at a time and comprise both protons and heavier ions with variable composition from event to event. Energies typically range up to several hundreds of MeV and have most influence on high inclination or high altitude systems. Irradiation also provides a versatile method for controllably introducing defects in GaN for fundamental studies.

In this review we summarize the results for various types of radiation damage effects in GaN and the most common devices fabricated with this material system.

## II Fundamental studies of radiation defects in GaN

The first experimental study of displacement threshold energy for GaN showed a Ga displacement energy of 19 ± 2 eV.[1] Comparison of defect production efficiency for 2 MeV protons and 2.5 MeV electrons showed that the protons were 1000 times more effective. The measured displacement threshold in GaN is much higher than in Si and GaAs due to the difference in the bond strength. In undoped n-GaN irradiated with 700–1000 keV electrons,[2] nitrogen vacancies were introduced with a rate close to 1 $cm^{-1}$ at such energies. Molecular dynamics calculations of displacement effects in GaN[3] showed that there exists a wide distribution of threshold energies for both Ga and N sublattices, and that effects of recombination induced by self-annealing caused by athermal local energy transfer are very important. The minimal energies of defect formation were 18 ± 1 eV for Ga and 22 ± 1 for nitrogen, but the average displacement energy was higher, 45 ± 1 eV (Ga) and 109 ± 2 (N). This modeling predicts about 5 times higher radiation tolerance of GaN compared to GaAs.

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Radiation defects in GaN are mobile even at low temperatures and the doses of ions producing amorphization are more than an order of magnitude higher than for GaAs.[4] Recombination of radiation defects during irradiation is prominent even at low temperatures and increases with increase in irradiation temperature.[4] At room temperature, the doses of irradiation producing amorphization increased from $\sim 10^{14}$ cm$^{-2}$ for heavy ions to $\sim 10^{16}$ cm$^{-2}$ for light ions because of the difference in relative energy loss by nuclear collisions, and increasing the implantation temperature increased the amorphization threshold by several orders of magnitude.

The most basic primary defects produced in GaN by any type of irradiation are Frenkel pairs in the Ga and N sublattices. Theoretical estimates predict that nitrogen vacancies in GaN are electronic resonances with levels in the conduction band. By capturing electrons they are turned into effective-mass-like (EM) shallow donors[5] and measurements of electrical properties of undoped n-AlGaN films as a function of Al composition suggest that such EM $V_N$ donors have an ionization energy of 40–60 meV in GaN and the Al concentration at which respective resonance levels emerge in the forbidden gap is ~20%.[6] For nitrogen interstitials $N_i$, theoretical calculations predict the existence of a deep acceptor near 1 eV from the conduction band bottom.[7,8] Ga vacancies $V_{Ga}$ in the doubly charged state produce an acceptor state near $E_v + 1$ eV, whilst Ga interstitials form negative-U type donors whose +/0 transition level is close to the conduction band edge and 3+/2+ charge transition level is located near $E_v$ + 2.6 eV ($E_c - 0.8$ eV).[7,8] Electron irradiation with energies of 0.7–1 MeV introduced new donors with an ionization energy of ~0.06 eV with an introduction rate of 1 cm$^{-1}$. The net electron concentration in GaN subject to electron irradiation hardly changed while the mobility of electrons decreased with increasing dose. Acceptor centers were introduced at a rate similar to the rate of the 0.06 eV donors. The observed effects were explained by the formation of Frenkel pairs in the nitrogen sublattice, with the nitrogen vacancies, $V_N$, responsible for the donors with thermal ionization energies at 0.06 eV and nitrogen interstitials, $N_i$, associated with the compensating acceptors.

Deep level transient spectroscopy (DLTS) measurements of 1 MeV electron irradiated n-GaN indeed showed deep electron traps with an activation energy of 0.9 eV that could be attributed to $N_i$ acceptors.[9] GaN irradiated with $^{60}$Co γ-rays also showed the presence of electron traps G1 with activation energy 80 meV (ref. 10) close to the one observed for $V_N$ donors.[2] In undoped n-GaN samples irradiated with 10 MeV electrons, deep electron traps with an activation energy of 1 eV that were shown to be acceptors and attributed to $N_i$ acceptors are created.[11] 2.5 MeV electron irradiation at 4.2 K produced in GaN a strong defect photoluminescence PL band centered near 0.95 eV, for which optically detected electron paramagnetic resonance ODEPR spectra could be obtained.[12] The PL line was attributed to Ga vacancy, $V_{Ga}$, with a level near $E_v + 1$ eV and the ODEPR process was interpreted as interaction with two different $Ga_i$ interstitial centers with levels close to $E_v$ + 2.6 eV. The quenching of the ODEPR signal for annealing to room temperature was then attributed to moving of the $Ga_i$ defect away from $V_{Ga}$.

If, however, only Frenkel pairs were produced by irradiation of GaN one would expect that, for n-GaN, the carrier removal rate at the initial stage of irradiation, when the density of radiation defects is lower than the concentration of dopant donors, will be equal to the production rate of $V_{Ga}$ and $N_i$, whilst for high irradiation doses the Fermi level will be pinned by the shallower of the two EM-like native donors, $V_N$ or $V_{Ga}^{1+}$. In p-type GaN the initial carrier removal rate should be close to that in the n-type until the aggregate density of $V_N$ and $Ga_i$ donors exceeds the density of acceptor dopants after which the Fermi level should jump to the level of $V_{Ga}$ near $E_v + 1$ eV and get pinned there. Generally this is not the case because primary defects recombine, form complexes with each other, with dopants and with extended defects, and because the energy of the primary recoils becomes so high that they produce collision cascades and form heavily disordered regions with a very high defect density in the core. DLTS measurements on n-GaN irradiated with electrons, γ-rays and protons with MeV energies to produce predominantly point radiation defects showed formation of electron traps with thermal ionization energies of 0.13 eV, 0.16 eV and 0.18–0.2 eV (ref. 10 and 13–18) (ER1, ER2, and ER3 according to the defect nomenclature introduced in ref. 15). These traps produce a broad feature in DLTS due to the proximity of emission rates of components, but this broad peak can be deconvoluted into separate defect contribution due to the large difference in electron capture cross-sections.[9,10,13–16] For the ER3 trap, the apparent activation energy is the sum of the trap ionization energy which is 0.06 eV, *i.e.* close to that of $V_N$, and the capture activation energy of about 0.14 eV. Based on this observation it was suggested that the ER3 traps are complexes of nitrogen vacancies with other native defects, such as $V_N$–$N_i$ or $V_N$–$N_{Ga}$–$Ga_N$–$N_i$.[9] The shallower defects ER1 and ER2 have been observed in a variety of surface treatments bound to produce nitrogen vacancies and are also thought to be $V_N$-related.[9] It was assumed that the 0.13 and 0.2 eV electron traps were donors as well as the 0.06 eV $V_N$ donors.[9] By implication, the carrier removal in irradiated n-GaN should then be controlled by the balance between the relatively shallow $V_N$-related donors, deep $N_i$ acceptors, deep $Ga_i$ donors and $V_{Ga}$ acceptors. However, this picture is negated by experimental measurements of the ionization energy dependence of ER3 traps on the applied electric field that indicates that these are acceptors.[14] An alternative suggested identity of these traps in ref. 14–16 associates them with $V_{Ga}N_i^{2-}$ complexes. In addition, DLTS spectra of n-GaN irradiated with light particles (γ-rays, MeV protons or electrons) are dominated by the relatively shallow ER1–ER3 electron traps. With rare exceptions, only for higher electron energies,[11] heavier ions (He, N),[15,19] neutron irradiation[20] or higher density of defects in proton implanted samples (as for 150 keV protons with the dose over $5 \times 10^{14}$ cm$^{-2}$ (ref. 21)) are deeper traps commonly detected. Electron traps that are generally observed in these cases show activation energies of 0.75–0.8 eV and 0.95–1.2 eV.[11,15,19–21] The 0.8 eV traps show a measurable decrease of the apparent ionization energy with increasing electric field whereas the 1 eV trap energy does not vary with the electric field. The first type of behavior is expected for donors because of the Poole–Frenkel effect,[22] whilst

 

the second type of behavior is characteristic of acceptors that are neutral when they emit an electron.[23] The attribution based mostly on theoretical calculations tends to ascribe the first to $Ga_i^{2+}$ deep donors and the second to $N_i^{2+}$ deep acceptors.[9,11,15,20] In addition to these deep centers, implantation of n-GaN with 150 keV protons to doses higher than $5 \times 10^{14}$ cm$^{-2}$ created deep electron traps whose energies increased from 0.2 eV at low dose to 0.25 eV, 0.32 eV and 0.45 eV at higher doses, suggesting that these centers are larger complexes produced by addition of new radiation defects to the more simple radiation defects formed at low doses.[21]

The knowledge of hole trap behavior in irradiated n-GaN is less satisfactory compared to electron traps. DLTS studies on n-GaN have mostly been carried out on Schottky diodes where probing of the traps in the lower half of the bandgap can only be achieved by optical injection of holes (by techniques such as deep level optical spectroscopy (DLOS)[24] or optical deep level transient spectroscopy (ODLTS)[25]). The results of these measurements critically depend on complete recharging of deep traps within the space charge region.[26–29] When the lifetime strongly decreases after irradiation, the ability to fully recharge hole traps becomes an issue and could lead to erroneous conclusions. Also, these traps are believed to be related to the yellow recombination band in GaN that is attributed to a donor–acceptor pair DAP transition involving a shallow donor and the $E_v$ + 0.9 eV hole trap.[6,30–33] These defects are stable up to 500 °C, even though the ODEPR signal vanishes because of the increased spatial separation of $V_{Ga}$ and $Ga_i$.[12]

Measurements of deep trap spectra of irradiated p-GaN are scarce. Standard DLTS measurements on Schottky diodes on p-GaN are difficult because of the high series resistance of the films and strong freeze-out of relatively deep Mg acceptors even at moderately low temperatures. Studies of radiation defects in p-GaN were carried out for 100 keV protons in ref. 34 and for fast reactor neutrons in ref. 35. Electron traps near $E_c$ − 0.5–0.6 eV and deep hole traps with levels near $E_v$ + 0.3 eV and $E_v$ + 0.85 eV were introduced by proton irradiation. The electron traps at $E_c$ − 0.5–0.6 eV are believed to be complexes of Mg acceptors with native defects and were shown to also give rise to an intense blue defect luminescence band centered at 2.9 eV.[36–39] The $E_v$ + 0.85 eV traps are probably the same as the $V_{Ga}$-related hole traps in n-type films. Their in-grown concentration in p-type GaN should be very low because of the high formation energy.[7] However, irradiation being a very nonequilibrium process, can introduce such defects and as a result produce the yellow luminescence band in heavily irradiated p-GaN films while yellow luminescence is manifestly absent in virgin p-GaN.[34]

Neutron irradiation was performed for two types of p-GaN films, one grown by hydride vapor phase epitaxy HVPE and the other by MBE.[35] HVPE films differed from MBE films by a very slight temperature dependence of conductivity, much lower mobility and the presence of additional acceptors with an activation energy of 0.12 eV, shallower than the ordinary Mg acceptors with an activation energy of 0.15 eV. Irradiation of both types of samples compensated acceptors and slightly but measurably increased the activation energy of major acceptors from 0.15 eV to 0.18 eV. Both types of samples remained p-type up to the dose of $1.7 \times 10^{17}$ cm$^{-2}$ and in both we observed the emergence of the yellow luminescence band due to the formation of the $E_v$ + (0.8–0.9) eV $V_{Ga}$-related acceptors after high doses. At a neutron fluence of $10^{18}$ cm$^{-2}$ both samples got converted to high-resistivity n-type samples with the Fermi level pinned near $E_c$ − (0.9–1) eV. Clearly, except for the $E_c$ − (0.5–0.6) eV Mg-related centers, the radiation defects introduced in p-type GaN are the same as in the n-type, but the carrier removal rate is about 20 times higher than n-GaN.[35]

## III Carrier removal and deep trap introduction rates

Basic questions in radiation physics of semiconductors include the mechanism for changes of electrical properties of an irradiated material and where the Fermi level is stabilized after high doses of radiation. In Fig. 1 (top) we present the carrier concentration changes for n-GaN samples irradiated with 10 MeV electrons: the 0.18 eV electron traps and the 1 eV electron traps. The carrier removal rate is 0.4 cm$^{-1}$, while the traps with thermal ionization energies of 0.18 eV and 1 eV have introduction rates of 0.2 cm$^{-1}$ and 0.8 cm$^{-1}$, respectively. The contribution from the other acceptors, the $V_{Ga}$ related $E_v$ + 0.9 eV hole traps in ODLTS spectra of irradiated samples, is underestimated by the interference of the 1 eV electron traps due to $N_i$ acceptors. Deconvolution of the 0.9 eV hole trap ODLTS feature gives the upper limit of the $V_{Ga}$ introduction rate as about 0.4 cm$^{-1}$. Thus, if we assume that the initial carrier removal rate comes from the difference in introduction rates of all these acceptors and the introduction rate of the 0.06 eV $V_N$ donors, the latter should be close to 1.4 cm$^{-1}$, *i.e.* ~seven times

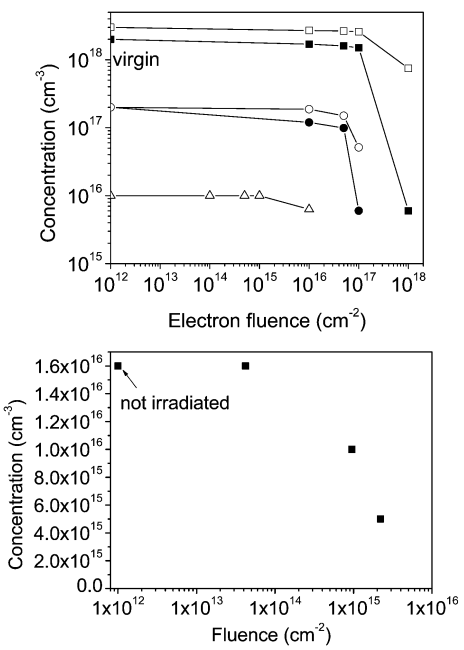


**Fig. 1** Decrease in electron concentration in n-GaN films of different doping concentration irradiated with (top) electrons or (bottom) fast neutrons to different doses.

that of the 0.18 eV traps. A similar relationship between the concentrations of these traps was determined from the electron concentration and mobility fitting and DLTS measurements for n-GaN irradiated with 1 MeV electrons in ref. 40 (1 $cm^{-1}$ for the $V_N$ centers *versus* 0.2 $cm^{-1}$ for the 0.18 eV centers). Hence, the data on electron removal by relatively low energy electrons can be consistently explained by the introduction rates of well documented radiation point defects.

This is not the case for fast neutron irradiated GaN. The main deep traps formed are 0.18 eV ER3 electron traps and 0.8 eV $Ga_i$ electron traps. However, the introduction rate of the shallower ER3 traps is very low and much lower than for electron irradiation. The introduction rate of the 0.8 eV traps is below 1 $cm^{-1}$ and is much lower than the electron removal rate of 5 $cm^{-1}$ (Fig. 1(b)). Besides, since these traps are deep donors they cannot contribute to carrier removal. The studies of hole traps are summarized in Fig. 2. In ref. 20 we suggested that the removal rate observed in neutron irradiated n-GaN can be explained by the formation of disordered regions (DRs) of the type described by Gossick.[41] Neutron irradiation produces a broad hole-trap-like feature in DLTS spectra at temperatures of 100–300 K. The apparent activation energy of this peak is 0.6–0.7 eV. If this were a true hole trap or a band of hole traps, it would be located near $E_v$ + 0.6 eV and its ODLTS signal should not be produced by optical excitation with photon energy <2.8 eV. However, the signal is effectively generated even for a photon energy of 1.4 eV. For neutron irradiated samples we see a strong persistent photocapacitance/photoconductivity PPC signal with an optical threshold close to 1 eV. Both the PPC phenomena and the appearance of the quasi-hole-trap signal in ODLTS stem from the existence of regions in which the bands are bent upwards by about 1 eV so that the electrons released from deep centers inside these regions are swept out by the built-in electric field of the region and have to overcome the barrier of about 1 eV to be recaptured by their host traps. It is natural to associate the regions in question with Gossick-like DRs,[41] *i.e.* heavily disordered core regions surrounded by the space charge region with a strong band bending. For very high neutron irradiation doses, the outer regions of the DR's overlap and the Fermi level pinning position in such materials give some idea about the Fermi level position in the core of the DR.

In heavily neutron irradiated GaN, irrespective of the starting conductivity type and doping, the Fermi level is pinned near $E_c$ − (0.9–1) eV.[42] There is a correlation between the Fermi level pinning position in the bulk of the GaN and the Fermi level pinning at the surface of n-GaN Schottky barriers. This is closely linked to the Fermi level stabilization, $F_s$, or the charge neutrality, CNL, concept that is introduced to explain a similar correlation in many other III–V materials.[43,44] Several theoretical models have been presented to account for the observed $F_s$ position in various materials. In one class of models, the Fermi level is believed to be trapped between the levels of the major native defects.[43] In GaN it is located between the levels of $N_i$ acceptors and $Ga_i$ donors. Moreover, lattice parameter measurements in heavily neutron irradiated GaN show an increase indicating that the dominant defects could be interstitials.[20,35,42] Rutherford backscattering experiments on neutron irradiated GaN also point to a very high density of interstitials, predominantly $Ga_i$.[45] In addition, neutron irradiation of undoped ELOG GaN resulted in a much lower effective removal rate than for the standard MOCVD material, 1 $cm^{-1}$ *versus* 5 $cm^{-1}$, as shown in Fig. 3.

Much needs to be done to build a more quantitative model. For example, the dependence of the carrier removal rate on the starting donor density is not easily explained by the classical Gossick model.[46–49] Fig. 4 shows such a dependence for n-GaN films with donor concentrations of $10^{15}$ $cm^{-3}$ and 1.6 × $10^{16}$ $cm^{-3}$. The carrier removal rate decreases by about 5 times when the donor concentration decreases. The carrier removal rate in neutron irradiated p-GaN is about 20 times higher than for

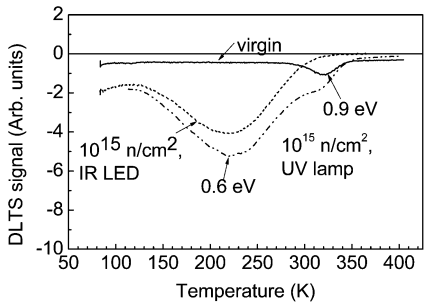


**Fig. 2** ODLTS spectra measured on virgin and neutron irradiated undoped n-GaN; reverse bias −1 V, injection with a pulse of UV deuterium lamp excitation, 5 s long, time window 300/3000 ms; also shown is the spectrum taken with 1.4 eV LED excitation.

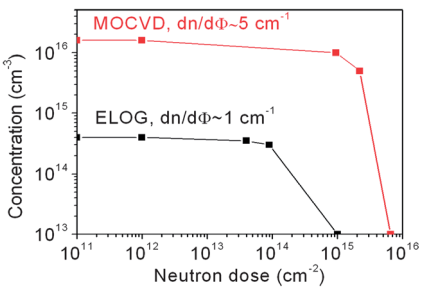


**Fig. 3** Decrease in carrier concentration in n-GaN films grown by MOCVD or the ELOG process as a function of fast neutron dose.

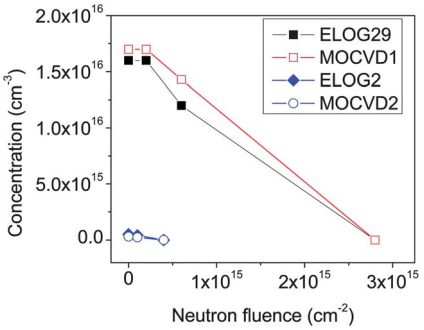


**Fig. 4** Changes in electron concentration induced in n-GaN films grown by MOCVD or ELOG by fast reactor neutron irradiation.

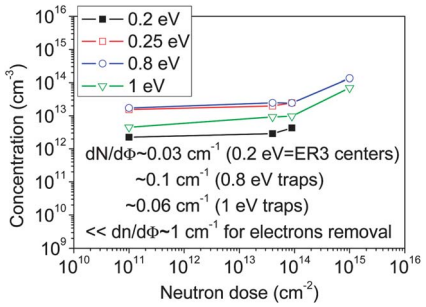


**Fig. 5** Changes in concentrations of electron traps and hole traps in neutron irradiated undoped ELOG n-GaN.

n-GaN despite the much higher concentration of acceptors in p-GaN than donors in n-GaN that should have overpowered the effect of increased barrier height in p-GaN DRs.[20,35] The same asymmetry of carrier removal rates was observed for proton implanted p-GaN and n-GaN[21,34] and it suggests that interaction of primary defects with Mg ions present in very high concentrations could be an additional factor in both cases.

DLTS spectra of neutron irradiated ELOG samples differed from MOCVD GaN in that the 1 eV $N_i$-related acceptor state could be clearly seen. However, as with MOCVD, the introduction rates of all traps were several times lower than the electron removal rate (Fig. 5) again suggesting the dominant role of DRs in carrier removal.

## IV Thermal stability of radiation defects in GaN

Annealing of defects introduced by particles with light mass (2 MeV protons, 0.2–2.4 MeV electrons)[15] showed that the shallow radiation defects ER1, ER2 and ER3 start annealing at 540 K and the annealing is complete after 620 K. Deeper electron traps ER5 associated with $N_i$ start annealing also at 540 K, but a higher annealing temperature of 660 K was needed for complete removal.[15] For $Ga_i$ deep donors, they start moving at room temperature.[12] The $V_{Ga}$ centers responsible for the 0.95 eV PL band were found to be stable up to 500 °C.

In GaN samples with a high density of radiation defects (high doses of ~100 keV implanted hydrogen or heavier ions, neutron irradiated material) the thermal stability of radiation damage is much higher.[50–61] After heavy proton implantation the band edge luminescence intensity is not restored to the pre-irradiation value even after annealing at 800 °C.[21] For activation of ion implanted donors (Si) or acceptors (C) annealing to temperatures exceeding 1000 °C was necessary and yet the degree of impurity activation was relatively low.[62] Fig. 6 presents the evolution of the sheet resistivity of an undoped GaN sample irradiated with fast and thermal neutrons to a fast neutron fluence of $1.5 \times 10^{17}$ $cm^{-2}$ (the ratio of fast and thermal fluences is 1 : 1).[23] The as-irradiated resistivity was high, decreased at the 150–250 °C stage, increased strongly at 250–450 °C and then gradually decreased in a very broad stage of 500–1000 °C.[23] The first stage corresponds to reconstruction of the ER3 and ER5 acceptors[15] which explains the decrease in resistivity. The reverse annealing stage at 250–450 °C is likely due to movement of the $N_i$ and $Ga_i$ centers, forming new deep compensating centers. The onset of the third stage of recovery at 500 °C correlates with the $V_{Ga}$ acceptor annealing stage[12] which explains the decrease of the resistivity. Still, even after annealing at 800 °C, the pre-irradiation resistivity was not reached; the Fermi level was pinned at relatively deep centers with an activation energy of 0.45 eV; the sample's series resistance was quite high which resulted in the appearance of DLTS peaks of the wrong sign. The most prominent electron traps were the 0.9 eV and the 1 eV traps that are likely to be related to the $Ga_i$ donors and the $N_i$ acceptors, but with a very high binding energy, possibly trapped within disordered regions. The ODLTS spectra were dominated by the hole traps with activation energy close to the $V_{Ga}$ 0.9 eV centers, but with a much smaller capture cross-section. After 1000 °C annealing the Fermi level was pinned near $E_c - 0.2$ eV, DLTS spectra were dominated by the 0.6 eV and 0.9 eV traps in high concentration. The total concentrations of the 0.45 eV traps pinning the Fermi level after 800 °C annealing and of the 0.2 eV traps dominant in the 1000 °C annealing are close to each other and equal to the number of donor Ge atoms converted from Ga by interaction with thermal neutrons ($2 \times 10^{16}$ $cm^{-3}$).[23] These results show that, even after low doses of neutron irradiation, it is hard to break down the disordered regions and to restore the virgin conductivity.

For very high neutron fluences, the resistivity of GaN passes through a maximum related to the onset of hopping conductivity (Fig. 6). The activation energy for the temperature dependence of resistivity for doses before the maximum showed a value of 0.9–1 eV. After the fluence corresponding to the maximum, the temperature dependence was much weaker. Annealing of heavily irradiated samples showed a strong reverse annealing stage up to 300 °C where the density of radiation defects decreased and the activation energy returned to 0.9 eV. A complete recovery could not be attained even after annealing at 1000 °C.

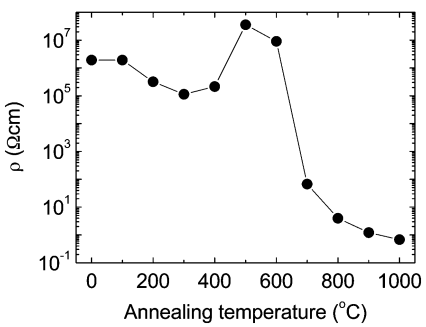


**Fig. 6** Sheet resistivity as a function of annealing temperature for an undoped GaN sample irradiated with fast and thermal neutrons to a fluence of $1.5 \times 10^{17}$ $cm^{-2}$ (reprinted with permission from the American Institute of Physics, Polyakov *et al.*, *J. Vac. Sci. Technol. B.*, 2010, **28**, 608).

## V Radiation effects in other III-nitrides

150 keV protons and $^{60}Co$ γ-ray irradiation of undoped n-InN films showed, in contrast to n-GaN, that irradiation increases electron concentration. This is linked to the difference in the

position of the Fermi stabilization level in GaN and InN that determines the band offsets in the respective heterojunction and the difference in the Schottky barrier height between GaN and InN.[63] With increase in the In composition of $In_xGa_{1-x}N$ solid solutions, the Fermi stabilization level moves upwards to the conduction band edge. The cross-over point is close to $x = 0.34$ and this composition separates solid solutions in which the electron concentration decreases with irradiation from those in which irradiation increases the electron concentration. Blue LEDs are built on GaN/InGaN QW structures with In mole fraction in the QW close to 0.2 and their behavior should be reasonably close to GaN.

For AlGaN there are only a few papers describing the effects of proton and neutron irradiation. n-$Al_{0.12}Ga_{0.88}N$ with a free electron concentration of $10^{17}$ $cm^{-3}$ irradiated at room temperature and at 300 °C with 2 MeV protons[18] showed that the carrier removal rate in AlGaN was about twice as high as that for GaN and decreased approximately by two times for high temperature irradiation. For undoped n-AlGaN with an Al mole fraction of $x = 0.4$, fast reactor neutron irradiation[64,65] led to compensation of the 0.25 eV traps and introduced deeper states with an activation energy of 0.28 eV at neutron fluences of $10^{15}$ $cm^{-2}$ to $2.5 \times 10^{16}$ $cm^{-2}$. For higher neutron fluences, deeper traps with activation energies of 0.35 eV and 1 eV were formed. After irradiation with $1.7 \times 10^{17}$ $cm^{-2}$ neutrons the films became semi-insulating with the Fermi level pinned near 0.35 eV from the conduction band edge.[66] Irradiation with higher neutron fluences increased the sheet resistivity of the layers in excess of $10^{14}$ Ω per square.

The introduction rate for compensating defects for neutron irradiated n-AlGaN was very much higher than for undoped n-GaN (about 500 $cm^{-1}$ *versus* 5 $cm^{-1}$) and even higher than for p-GaN (~100 $cm^{-1}$ (ref. 21)). Measurable changes of electrical properties started at a neutron fluence of $10^{15}$ $cm^{-2}$, *i.e.* similar to the undoped n-GaN,[20] even though the concentration of centers to be compensated was 2 orders of magnitude higher for n-AlGaN.

The effects of proton implantation were similar to the effects of neutron irradiation, but the 100 keV protons started to change the electrical properties of AlGaN after a dose of $10^{12}$ $cm^{-2}$, again two orders of magnitude lower than for the undoped n-GaN, even despite a much higher donor density in n-AlGaN.[21] As for neutron irradiation, the resistivity of the samples rapidly increased with increasing the proton fluence and after irradiation with $10^{14}$ $cm^{-2}$ of protons the sheet resistivity was $10^{13}$ Ω per square.

For p-AlGaN with an Al mole fraction of $x = 0.12$, in the unirradiated material the electrical properties were determined by Mg acceptors with an activation energy of 0.17 eV and a concentration of $3 \times 10^{18}$ $cm^{-3}$. The concentration decreased after irradiation with a low proton fluence of $10^{12}$ $cm^{-3}$. After irradiation with $10^{13}$ $cm^{-2}$, the apparent activation energy of the dominant acceptors increased to 0.2 eV while the concentration further decreased. Irradiation with a fluence of $10^{14}$ $cm^{-2}$ totally compensated the p-AlGaN film down to the depth corresponding to the range of 100 keV protons. The Mg-related MCL band intensity decreased by about 10 times after irradiation with $10^{14}$ $cm^{-2}$ protons and after irradiation with higher proton doses we observed, alongside with the decrease of the intensity of this band, the emergence of the yellow band. The observed changes, both qualitatively and quantitatively, are similar to proton irradiation effects in p-GaN.

## VI Radiation effects in GaN-based devices

AlGaN/GaN HEMTs irradiated with 1.8 MeV protons showed decrease of transconductance, threshold voltage and drain saturation current occurred after proton fluences of $10^{14}$ $cm^{-2}$.[67–80] This dose is about two orders of magnitude higher than for AlGaAs/GaAs HEMTs.[81] Annealing at 800 °C was shown to be efficient in partially restoring the electrical characteristics.[80] For higher proton energies of 40 MeV both DC characteristics (transconductance, threshold voltage, drain saturation current) and AC characteristics were little affected by proton fluences of up to $5 \times 10^{10}$ $cm^{-2}$.[81–85] Proton irradiation of

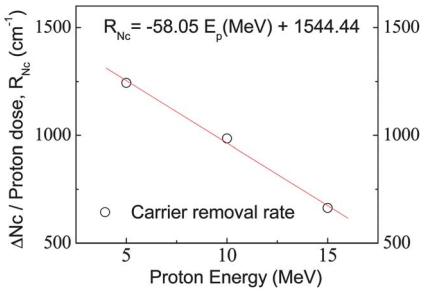


**Fig. 7** Carrier removal rate as a function of proton energy in AlGaN/GaN HEMTs.

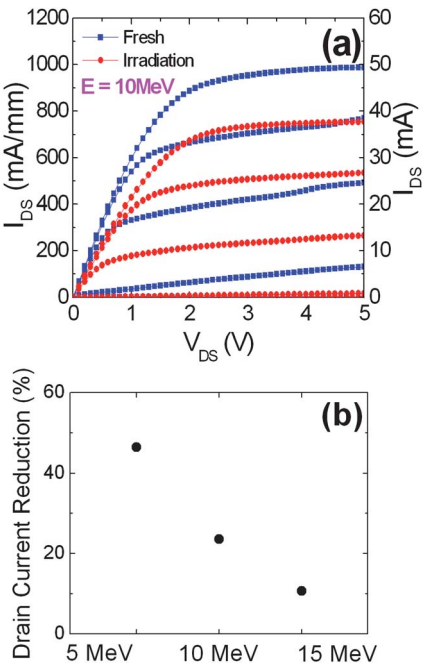


**Fig. 8** (Top) Drain *I–V*s of AlGaN/GaN HEMTs pre- and post-proton irradiation with a proton energy of 10 MeV. (Bottom) Saturation current at $V_{DS} = +5V$ as a function of irradiation energies.

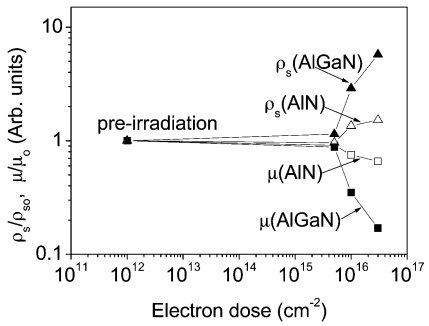


**Fig. 9** Relative changes of 2DEG resistivity and mobility after 10 MeV electron irradiation of AlGaN/GaN and AlN/GaN HEMT structures (reprinted with permission from the American Institute of Physics, Polyakov *et al.*, *Appl. Phys. Lett.*, 2008, **93**, 152101).

AlGaN/GaN HEMTs with energies 1.8 MeV, 15 MeV, 40 MeV, and 105 MeV, and proton fluences up to $10^{13}$ $cm^{-2}$ showed the strongest changes for the lowest proton energy and the effect was explained by the decrease of the energy transferred to Al, Ga and N atoms in elastic collisions occurring within the active region of devices as the range of protons increased with increasing energy. Gamma irradiation of AlGaN/GaN HEMTs up to a dose of 600 MRad did not substantially change the characteristics.[86]

It was reported that the carrier removal rate was about four times higher in InAlN/GaN HEMTs compared to the AlGaN/GaN devices shown in Fig. 7. Fig. 8(a) shows the drain current–voltage (*I–V*) curves of AlGaN/GaN HEMTs before and after 10 MeV proton irradiation. Although all the proton-irradiated HEMTs exhibited good pinch-off characteristics, the amount of saturation drain currents that were degraded was dependent on the irradiation energy. For the 10 MeV irradiated HEMTs, the reduction of saturation drain current at $V_G = 0$ V was 23.6%. Much larger saturation drain current reduction, 46.4%, was observed for the HEMTs irradiated with proton energy at 5 MeV and only 11.5% drain current reduction was exhibited for the HEMTs irradiated with proton energy at 15 MeV, as illustrated in Fig. 8(b).

Under electron irradiation, we found that AlN/GaN HEMTs suffered less degradation in carrier concentration and mobility than their AlGaN/GaN counterparts, as shown in Fig. 9. This is consistent with the higher average bond strength of the former.

One more important thing to be noted in conjunction with AlGaN/GaN HEMTs performance as affected by radiation is the current collapse phenomenon in these devices. This phenomenon is a strong dependence of AC characteristics of transistor on frequency: a substantial loss of current that can be passed through a device at high frequency compared to DC characteristics, a substantial lag in the switching performance when rapidly changing the gate voltage, and long-term drift of parameters after driving the device at high current.[87–91] The most important among those is the loss of current transfer characteristics at high frequencies which is believed to occur because of the trapping of charge carriers in the AlGaN barrier and the formation of a virtual gate with a much increased area compared to the actual metallic gate.[87–92] The effect can be mitigated by deposition of dielectric layers ($Si_3N_4$, $Gd_2O_3$, MgO) on top of the barrier[93–97] due to a decrease of the surface trap density.[98] It has been demonstrated for $Gd_2O_3$ and MgO passivating layers that their beneficial effect is not affected by proton and γ-ray irradiation.[84–86]

Another important class of GaN-based devices is LEDs. For double heterostructure blue GaN/InGaN LEDs measurable degradation of the light output started after irradiation with $10^{14}$ $cm^{-2}$ neutrons.[99] For proton irradiated AlGaN/GaN QW LEDs, the threshold dose for the start of degradation was two orders of magnitude higher than for AlGaAs/GaAs QW LEDs ($10^{12}$ $cm^{-2}$ *versus* $10^{10}$ $cm^{-2}$ for 3 MeV protons). Increasing the proton energy from 3 MeV to 5 MeV measurably increased the dose necessary for the onset of light output degradation, most likely due to a lower energy going into elastic collisions within the active region of devices. Even higher proton doses were found necessary for changing the characteristics of proton-irradiated blue GaN/InGaN LEDs.[100] For green GaN/InGaN LEDs[101] it was reported that 2 MeV protons produce about 40% light output decrease after a fluence of $1.7 \times 10^{12}$ $cm^{-2}$.

## VII Summary and conclusions

Radiation effects in GaN can be reasonably well understood based on a simplistic picture in which the main radiation defects are due to shallow $V_N$ and deep $Ga_i$ donors and deep $V_{Ga}$ and $N_i$ acceptors. This picture places the $V_N$ donors near $E_c - 0.06$ eV, the $Ga_i$ doubly charged donors near $E_c - 0.8$ eV, the $V_{Ga}$ acceptors near $E_v + 1$ eV and the $N_i$ acceptors near $E_c - 1$ eV. A schematic representation of the levels in the bandgap for this model is shown in Fig. 10. Other prominent defects in n-GaN, relatively shallow ER1–ER3 traps, seem to be complexes of these primary defects, mostly of $V_N$, with unidentified species, possibly with donors in the case of ER3 if one considers the results of neutron doping and annealing experiments. In p-GaN there is evidence for formation of defects near $E_c - 0.5$ eV that seem to be Mg acceptor complexes with native defects and also of defects of unidentified nature with a level near $E_v + 0.3$ eV. The carrier removal rate in GaN for light particles is well accounted for by the introduction of these simple defects.

For particles, such as fast neutrons that produce large recoil cascades, the data suggest carrier removal by disordered regions in which the Fermi level in the core is pinned between the $Ga_i$ donor level and the $N_i$ acceptor level. In heavily irradiated

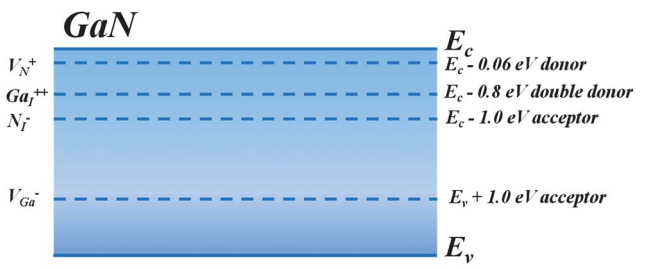


**Fig. 10** Simple model for radiation defects created in GaN by protons and other ionizing radiation.

samples these disordered regions pin the Fermi level near $E_c - 1$ eV irrespective of the initial type of conductivity and doping level. Charge trapping in DRs contributes to strong persistent photoconductivity effects in neutron irradiated GaN. Both theoretical modeling and experiment show a much higher radiation tolerance of GaN compared to Si or GaAs. When comparing the results for different types of particles and different energies of particles one can see that where such comparison with modeling has been done it suggests that modeling provides a reasonable guide for predicting the relative effectiveness of different radiation types in changing the electrical properties of GaN. Thus the work carried out so far can serve as a firm basis for further studies and for developing predictive models of materials and device performance in III-nitrides.

However, many issues still have to be addressed. Among them are: (1) the strong asymmetry in carrier removal rates in n- and p-type materials and possible interaction of radiation defects with Mg acceptors; (2) the scarcity of our knowledge of radiation effects in III-nitrides other than GaN; (3) the poor state of understanding of radiation defects in nitrides with dislocations present in the as-grown material, particularly given the fact that the dislocation density can be very high; (4) the lack of proper understanding of radiation defect interaction with dopants and impurities; (5) the poor understanding of the nature of recombination processes in irradiated nitrides; and (6) the little effort devoted to electrical and recombination effects in homojunctions, GaN/InGaN QW heterojunctions, effects at the AlGaN/GaN, GaN/InGaN interfaces for various compositions of HJs and QWs. In addition, there has been little study of the electronic collisions of high-energy ions which can create damage tracks in materials, including GaN.[102–105] Some results, such as apparently lower radiation stability of n-AlGaN films compared to GaN, do not fit theoretical predictions and suggest checking possible effects of crystalline quality on performance. Much more has also to be done to properly understand defect transformation upon increasing the irradiation temperature and upon annealing. For example, ODEPR experiments on low-temperature electron irradiation of GaN suggest that $V_{Ga}$ acceptors are annealed at 500 °C, yet these acceptors can be clearly seen in the neutron irradiated material even after annealing at 1000 °C.

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