Improved electrical performance of MOCVD-grown GaN p-i-n diodes with high-low junction p-layers

Jennifer Howell-Clark, Zhibo Guo, Christian Wetzel, T. Paul Chow, Piao Guanxi, Yoshiki Yano, Toshiya Tabuchi, Koh Matsumoto

A R T I C L E   I N F O
Keywords:
Gallium nitride
Quasi-vertical p-i-n diodes
High-low junction
Hole injection

A B S T R A C T

We report that using a high/low p-type junction anode structure results in improved hole injection over a uniformly-doped p-anode layer in GaN pin junction diodes. Quasi-vertical diodes with a 20 nm thick, magnesium concentration of $2 \times 10^{20} \text{cm}^{-3}$ layer on top of a 480 nm thick layer with a magnesium concentration of $10^{19} \text{cm}^{-3}$ show greatly increased forward current density ~$100\times$ higher than those with a 500 nm thick uniformly $3 \times 10^{19} \text{cm}^{-3}$ Mg-doped p-layer. Forward knee voltage and ideality factor are reduced by a factor of more than two, and reverse leakage current density is also reduced. Additionally, the specific differential series resistance is reduced significantly. With photoluminescence measurements, we found that these improvements are largely due to improved p-GaN material quality of the high/low junction sample with lower average Mg concentration.

1. Introduction

Electronic devices derived from gallium nitride (GaN) and related III-Nitride materials such as aluminium gallium nitride (AlGaN) and indium gallium nitride (lnGaN) have attracted intense research interest in recent years due to their potential in multiple areas of semiconductor technology. III-Nitride based light-emitting diodes emitting in blue and green wavelengths are already widely commercially available, but GaN has also been examined for high-voltage power device applications due to its high critical field (> 3 MV/cm) [1]. Power rectifiers often use a p-i-n type structure, since a thick layer of lightly doped – usually n-type material is needed to support a high voltage (this type of structure is still commonly referred to as “p-i-n” even when the central drift region is not actually intrinsic). Owing to its ultraviolet (~3.4 eV at 300 K) band gap, GaN p-i-n structures have also been proposed for use as solar-blind photodiodes [2].

To date, the most commonly used acceptor element for producing p-type conductivity in GaN is Mg. However, the tendency of Mg to acceptors to complex with interstitial H atoms complicates sample processing (because the p-type conductivity must be activated ex situ), and since the acceptor ground state is relatively far (~170 meV) from the valence band edge, the hole concentration will be relatively low at room temperature [3]. It has been previously established that the maximum hole concentrations are typically achieved at Mg concentrations of $2\sim3 \times 10^{19} \text{cm}^{-3}$; the hole concentration is typically observed to decrease if the Mg concentration is increased significantly above this level, due to increased prevalence of compensating defects [4,5].

Commercially, p-i-n rectifiers are more commonly manufactured using indirect bandgap materials such as silicon. Since the carrier lifetimes in indirect gap materials are typically quite long (on the order of microseconds for Si), ambipolar carrier transport allows for conductivity modulation of the lightly doped center region, reducing its effective resistivity [6]. For direct gap materials like GaN, carrier lifetimes are liable to be much shorter, resulting in different physics of operation, since the carrier diffusion lengths will be much shorter than the thickness of the lightly doped drift layer.

Consequently, GaN p-i-n rectifiers are liable to have entirely different optimizations – from an electrical performance point of view – compared to similar rectifiers fabricated from indirect gap semiconductors such as Si or 4H-SiC, necessitating new experiments to determine the optimal device structures for GaN.

Several research groups have investigated the properties of GaN p-i-n structures in recent years; Gupta, et al. investigated the effect of the

---

* Corresponding authors.
E-mail address: chowt@rpi.edu (T.P. Chow).

https://doi.org/10.1016/j.sse.2019.107646
Received 2 August 2019; Accepted 2 September 2019
Available online 03 September 2019
0038-1101/ © 2019 Elsevier Ltd. All rights reserved.
thickness of the lightly doped drift layer on GaN p-i-n diodes grown on sapphire substrates [7], and Zhang and co-workers demonstrated similar devices on [111] Si substrates [8]. Such structures have also been successfully transferred from the growth substrates to metallic contact layers to obtain fully-vertical diodes [9]. On conductive bulk GaN substrate, Ohta et al. reached breakdown voltages as high as 4 kV [10]. However, specific studies of the p-layer doping profile in MOCVD-grown p-i-n diodes have not been previously reported. In all the reported work above, p-GaN layers are uniformly Mg-doped at higher than $1 \times 10^{19}$ cm$^{-3}$ to maximize hole concentration; however, maximizing the hole concentration in the p-layer alone does not necessarily guarantee the best electrical performance in a bipolar device such as a junction diode. Our study shows that an Mg-doped high/low junction p-layer delivers increased forward current density as well as reduced ideality factor and forward voltage compared to a more conventional, uniformly doped p-layer.

2. Experimental procedure

2.1. Epitaxial structures

Two GaN p-i-n epilayers – Sample A and Sample B – were grown on c-plane sapphire substrates via metal organic chemical vapor deposition (MOCVD). Both samples nominally consist of a 3 μm thick unintentionally doped (UID) GaN buffer layer, 1 μm thick n+ ([Si] = $3 \times 10^{18}$ cm$^{-3}$), a 4.2 μm thick drift layer, and an 0.5 μm thick Mg-doped layer (Fig. 1). Sample A has a uniform Mg concentration of $3 \times 10^{19}$ cm$^{-3}$ in the Mg-doped layer, whereas sample B has a high/low junction consisting of a 480-nm thick layer with [Mg] = $1 \times 10^{18}$ cm$^{-3}$ capped by a 20 nm-thick layer with an Mg concentration of $1 \times 10^{20}$ cm$^{-3}$. Similar layer arrangements have been noted as p/p$+$ sequence, e.g., in Ref. [10]. The average concentrations of Si, Mg, O, and C in the lightly doped drift layers were each determined to be $10^{16}$ cm$^{-3}$ or less via secondary ion mass spectrometry (SIMS).

2.2. Device fabrication

The fabrication of quasi-vertical p-i-n diodes was preceded by a p-GaN activation anneal at 650 °C for 30 min in oxygen ambient. Deep mesa etching of 5 μm was performed in BCl$_3$/Cl$_2$ plasma with a nickel hard mask. After sidewall etching, a treatment in 85 °C tetramethylammonium hydroxide (TMAH) for 15 min to remove etching damage was carried out followed by sidewall passivation by 1 μm of SiO$_2$ using plasma enhanced chemical vapor deposition (PECVD) with TEOS as precursor. The SiO$_2$ was annealed at 900 °C for 30 min in nitrogen. Subsequently, Ti/Al/Ni/Au Ohmic contact to n + GaN was deposited by e-beam evaporation. Prior to Ni/Au contact deposition on p-GaN, the p-GaN surface was treated by diluted HCl (HCl:H$_2$O = 1:3) for 3 min. The fabrication was completed by depositing Ti/Mo as probing pads (Fig. 2). The contacts were annealed at 500 °C for 1 min under O$_2$ in a rapid thermal annealer.

Optical microscope images of fabricated devices (including Ti/Mo pads) are shown in Fig. 3. To study the potential effects of active area on device performance, devices with three different diameters (112, 226, and 394 μm) were prepared. The Ni/Au contacts to the p-type layer on top of the mesas are semi-transparent to allow observation of electroluminescence through the contact, which is not obstructed by the Ti/Mo final metal.

2.3. Optical & material characterization

Prior to any fabrication, the epitaxial samples were characterized by photoluminescence (PL) and X-ray diffraction (XRD), in order to evaluate the material properties.

PL tests uses a He-Cd laser emitting at a peak wavelength of 325 nm. Samples were cooled to 5 K, and PL spectra were collected using a charge-coupled device (CCD) spectrometer.

XRD measurements were conducted using a commercial apparatus equipped with a Cu Kα (0.15 nm) X-ray generator. Samples A and B were characterized by obtaining the θ rocking triple full width at half maximum (FWHM) values for the (0 0 2) and (1 0 2) reflections in GaN,

![Fig. 1. Nominal thicknesses and doping profiles of the epitaxial film samples, Sample A (left) and Sample B (right).](image)
which are frequently used as proxies for the prevalence of extended defects.

2.4. Electrical characterization

After the fabrication of quasi-vertical p-i-n diode devices, the devices were characterized to determine their electrical performance. Current-voltage curves were obtained for all mesa sizes in both forward and reverse bias at room temperature. Forward bias current-voltage curves were also obtained at high temperature via controlled heating of the probe station vacuum chuck. Forward bias curves were used to obtain the forward voltage and ideality factors for devices in the two samples, and the reverse bias curves used to determine leakage current and breakdown properties.

The small-signal junction capacitance of fabricated devices was also obtained at thermal equilibrium and in reverse bias, using a measurement AC voltage amplitude of 25 mV and a frequency of 1 MHz.

p-type contact resistivities were also estimated on both samples, using circular transmission line measurement (TLM) structures fabricated using the same metallization as the p-type contacts to the p-i-n diodes.

3. Results & discussion

3.1. Optical characterization

Photoluminescence was performed using 5 mW of the 325 nm line of a HeCd laser. In GaN, the absorption coefficient $\mu_a$ at this wavelength is high ($\sim 10^5$ cm$^{-1}$). Since the fraction $T$ of an incident beam that successfully transmits through a path length $L$ in a material with absorption coefficient $\mu_a$ is given by

$$T = e^{-\mu_a L}$$  

over 99% of the excitation laser will be absorbed in the top 500 nm of the samples, i.e. the Mg-doped layers [11]. As a result, the PL spectra can be associated exclusively with the Mg-doped layers, with contributions from the underlying n-type and UID layers being negligible. The spectra collected at $T = 5$ K before and after the p-layer activation anneal are shown in Fig. 4. The spectra of the two Mg-doped layers are very different. Prior to the activation anneal, Sample A is seen to exhibit a single broad peak at 3.15 eV, while Sample B exhibits several sharp peaks at 3.49, 3.28, 3.19, and 3.09 eV. The sharp peaks in the Sample B spectrum are frequently observed in unannealed Mg-doped GaN samples, and the 3.19 and 3.09 eV peaks are considered to be longitudinal optical (LO) phonon replicas of the 3.28 eV line, which is itself typically associated with a donor-acceptor pair (DAP) recombination involving a shallow donor and an acceptor [12]. The feature observed in Sample A is less typical, but similar features have been associated with a different DAP transition, in particular one involving deep donors which compensate the Mg-doped material, suggesting that the p-layer of Sample A is highly compensated [13,14]. In another common explanation, the band orginates in a shallow donor to shallow Mg acceptor transition in the presence of large potential fluctuations [15].

Since the band gap of GaN is 3.5 eV at 5 K, the 3.49 eV feature seen in Sample B may be attributed to the direct recombination of excitons, either free or bound to shallow donor states slightly below the conduction band edge, such as the +1/0 transition levels of O$_{Si}$ or Si$_{Ga}$, which may be present in at least small concentrations in the Mg-doped layers [16,17]. Acceptor bound exciton (ABX) lines are also sometimes observed in Mg-doped GaN samples, but at lower energies than 3.49 eV [18].

After annealing, the only feature visible in Sample A is a broad, low-intensity feature at approximately 2.5 eV. Often the appearance of a so called “blue band” around 2.9 eV is observed in Mg doped GaN. Here however, we find no contribution in the range of 2.8–3.3 eV. This, is in agreement with our observation that the blue band only appears once Mg doping is too high for highest hole concentration and compensating defects start to form. In Sample B we observe a feature at $\sim 3.4$ eV. Since the band gap of GaN is 3.5 eV at 5 K, and not 3.4 eV, this line is not due to direct band-to-band recombination. A 3.4 eV luminescence at low temperature has also been attributed to the presence of stacking fault...
defects; however, this is unlikely to be the cause of the feature in Sample B, since XRD measurements did not detect a substantive difference in extended defect concentration between the two samples (Table 1). Instead, we speculate that this 3.4 eV feature might be due to the formation of inversion domains in the heavily doped “contact layer” in Sample B; such domains have been shown to form in high concentrations at Mg concentrations in the 1020 cm⁻³ range due to N-Mg-N bond formation [19]. 3.4 eV lines have also been associated in literature with LO phonon replicas of a shallow-bound exciton line, since the difference between the peak energy of the 3.4 eV line and the 3.49 near band edge (NBE) line is very close to the LO phonon energy in GaN, which is approximately 92 meV [20].

For GaN, the presence and intensity of LO phonon replicas is correlated with reduced density of point defects, since defects tend to scatter phonons and reduce their coupling to electronic transitions [21]. As a result, we associate the prevalence of these replicas in Sample B – and their absence in sample A – with a difference in material quality in the Mg-doped layers.

By contrast to PL, which couples to point defects, XRD measures the reciprocal space of the crystal lattice, and thus is more sensitive to extended defects such as threading dislocations (TDs) and stacking faults, which produce more long-range distortions in the lattice constants (Table 1).

The FWHMs of the (0 0 2) and (1 0 2) reflections are found to be comparable between samples, indicating that the prevalence of extended defects is similar for both samples. This is consistent with the fact that extended defects like TDs typically originate in the lower layers of an epitaxial GaN film, since the lower layers – i.e. the UID buffer layers - of samples A and B are very similar.

### 3.2. Electrical properties

The forward bias J-V curves at T = 298 K are shown in Fig. 5. For both Sample A and Sample B devices, the current density is slightly higher for smaller mesa sizes; this is attributed to current crowding in the larger diode mesas. For all forward voltages, the current density is greater in Sample B than in Sample A; the forward knee voltage (obtained via linear extrapolation of the series resistance dominated region) is approximately 8.8 V for Sample A (uniform p-layer) and 3.5 V for Sample B (high-low junction p-layer).

If the forward diode current in both samples is modelled according to the Shockley diode equation, then at sufficiently large forward bias, the ideality factor n can be determined at a given bias point from

$$J = J_0 e^{\frac{qV}{n kT}}$$  \( (2) \)

Then the ideality factor n can be determined at a given bias point from

$$n(V) = \frac{q}{kT} \left( \frac{d(\ln J)}{dV} \right)^{-1}$$  \( (3) \)

Ideality factors as determined from Eq. (3) are shown in Fig. 6. The minimum ideality factor value near 2.0 is obtained at 1.0–1.2 V bias for both the samples; however, since the current at this bias is very small, the value of n as determined from Eq. (3) does not necessarily provide much information, especially since the assumption of Eq. (2) may not be as accurate (see discussion on parasitic resistance below). However, for almost all forward bias voltages, the values of n are strictly less for Sample B than for Sample A, indicating that for Sample B, the high/low junction p-layer, the forward current increases more quickly. Sample B shows a pronounced local minimum value at 3.2–3.4 V; this minimum is also present, although much less pronounced, in Sample A. This is consistent with the expected built-in potential.

The fact that these minima occur at roughly the same voltage, but with different magnitudes, is instructive. First, it demonstrates that the lower current and higher forward voltage in Sample A are not simply due to debiasing by a parasitic resistance; since the ideality regimes, which are determined by the underlying device physics, depend on the bias on the junction itself. This furthermore becomes apparent when explicitly assuming a parasitic resistance R in Eq. (2) by replacing V by V – J A R with cross-section A and solving for R. Yet that value would
have to vary from 5 Ω to 6 MΩ, i.e. over six orders of magnitude over the voltage range of 1–3.7 V, which is not realistic. So even though GaN is a direct semiconductor and the minority diffusion length is short, it is the minority carrier injection level that determines the apparent resistance by the junction itself.

Furthermore, the coincidence of the ideality regimes between the two samples indicates similar biases on the junctions (at least up to the flatband voltage), demonstrating that the lower current and higher apparent ideality factor in Sample A is not due simply to parasitics. Above the flatband voltage of approximately 3.2–3.4 V (i.e. slightly less than 2 kT/q), the apparent “ideality factor” as determined from Eq. (3) increases rapidly for both samples, as the current becomes limited by the overall series resistance of the device rather than the junction itself.

The forward bias behavior was examined as a function of junction temperature as well; the forward bias current in devices with a diameter of 226 μm are shown in Fig. 7. For purposes of this measurement, substrate temperature was assumed to be equal to the temperature of the underlying chuck. For both samples, the current density is strictly increased by increasing junction temperature. Since bulk mobilities are typically decreased at higher temperature, this indicates that the forward current density is limited by carrier density and/or non-drift transport, rather than by bulk mobility. Due to the relatively high activation energy of the Mg acceptors, it is likely that a higher junction temperature increases hole current by increasing acceptor ionization. Higher junction temperature also increases the thermal voltage $kT/q$, leading to increased rates of carrier transport via diffusion, which tends to dominate at intermediate biases [22].

Devices were also examined in reverse bias, in order to determine their reverse leakage and breakdown characteristics. The reverse bias leakage current of both Sample A and Sample B was found to exhibit an exponential voltage dependence, as seen in Fig. 8; the behaviour is well described ($R^2 > 0.95$) by single exponential fitting. Unlike silicon devices, in which space-charge generation current dominates p-n junction leakage, the leakage current in GaN devices does not depend on the generation current. This is due to the fact that the space charge generation current density is given by

$$J_g = \frac{n_i q n_{diff}}{\tau_g} \sqrt{V_{bb} + V_{bi}}$$  \hspace{1cm} (4)

where $n_i$ is the intrinsic carrier density, $\tau_g$ the space charge generation lifetime, and $V_{bb}$ the width of the space charge region, which for an abrupt junction is approximately proportional to the square root of the reverse bias $V_{bb}$ when the applied bias is large compared to the built-in potential $V_{bi}$ [20]. Since in GaN the intrinsic carrier density is very small (on the order of $10^{15}$ cm$^{-3}$), the amount of generation current is typically negligible, even though $\tau_g$ is likely to be short. We therefore find the leakage current of the samples in this work does not exhibit a square root-like behavior, so contributions from space charge generation current can be ignored. Similarly, a diffusion leakage current at the blocking (i.e. p/n-) junction

$$J_{diffusion} \approx \frac{q D_p \tau_p n_i^2}{N_0} \propto n_i^2$$  \hspace{1cm} (5)

will likewise be negligible due to the small value of the intrinsic carrier density, where $D_p$, $\tau_p$, and $N_0$ are the hole diffusion constant, hole lifetime, and drift layer doping, respectively [20].

Instead, the leakage current is thought to be dominated by defect-assisted tunnelling mechanisms, such as variable-range hopping of carriers between defect states [23,24].

Multiple proposed models for variable-range hopping current have been proposed. The current due to variable-range hopping can be modelled as a single exponential dependence on the electric field

$$J = J_0 e^{C_{EF} \left( \frac{E}{T_F} \right)^{1/4}}$$  \hspace{1cm} (6)

Or alternatively as a compressed exponential dependence on the electric field

$$J = J_0 \left( e^{C_{EF} \left( \frac{E}{T_F} \right)^{1/4}} - e^{-\frac{1}{3} \lambda \left( \frac{E}{T_F} \right)^{1/4}} \right)$$  \hspace{1cm} (7)

where $E$ is the magnitude of the electric field, $C$ is a constant, and $T_F$ is a characteristic temperature [23–25]. Since in a p-n diode the drift layer field is trapezoidal, approaching the rectangular field limit at low doping, the field will be proportional to the reverse bias voltage. The leakage current for both Sample A and Sample B appears to exhibit a single exponential dependence on the bias voltage, like that in Eq. (6). Consequently these leakage currents can be quantified by obtaining their parameters – prefactor $J_0$ and exponential constant $C T_F^{1/4}$ – from fits. In particular, the exponential factor for Sample B is more than twice as large compared to Sample A. This discrepancy may be due to an increased concentration of point defects in the drift layer of Sample A, since the concentration of extended defects as determined from the XRD measurements is expected to be similar.

By contrast, as shown in Fig. 9, reverse leakage current density of Sample B is 0.5 mA/cm² at average field of 0.7 MV/cm, which is near the lower end of the state-of-the art GaN-on-sapphire devices [23]. In general, the leakage currents of such structures are observed to exhibit an exponential dependence on voltage, in agreement with the variable-range hopping theory [23–25]. There is significant variation between the samples in terms of both the prefactor $J_0$ and the exponential constant, suggesting that they are not intrinsic material properties. Since the exact mechanisms of different species of defects in carrier hopping in GaN are not precisely understood, further study of its effects
and sample dependence are needed.

Both samples were found to exhibit some degree of electroluminescence (EL) (Fig. 10). Sample B emits substantially more light even at lower forward bias, indicating increased electron-hole recombination. Since luminescence of GaN diodes is typically limited by hole transport more than by electron transport, we conclude that the more intense EL in Sample B (high-low junction p-layer) devices is due to improved injection of holes into the lightly doped drift layer.

The effective doping in the drift layer was also estimated from capacitance-voltage measurements. Since the doping in the drift layer is lighter than both the n+ layer and the Mg-doped layers, it can be estimated from

$$N_{\text{eff}}(V) = \frac{2}{q\epsilon_s} \left( \frac{dC}{dv} \right)$$

(8)

where $N_{\text{eff}}$ is the effective doping concentration on the lightly doped side of a p-n junction, $V$ is an applied (reverse bias) voltage, $\epsilon_s$ is the dielectric permittivity of the semiconductor, and $C$ is the small-signal capacitance per area of the junction (since the alignment tolerance of the Ni/Au p-type contacts is very small compared to the mesa size, the mesa area can be used to normalize the junction capacitance). Using this method, the effective doping concentration was determined to be equal to or less than $4 \times 10^{15}$ cm$^{-3}$ for the drift layers of both Sample A and Sample B.

The specific differential series resistances from Sample B are found to be 34 m$\Omega$-cm$^2$, 52 m$\Omega$-cm$^2$, and 71 m$\Omega$-cm$^2$ for the 112 $\mu$m, 226 $\mu$m, and 394 $\mu$m devices, respectively, as obtained from the slope of the linear (i.e. series resistance dominated) regions of the forward biased current-voltage curves. The corresponding specific differential series resistances of sample A are 723 m$\Omega$-cm$^2$, 814 m$\Omega$-cm$^2$, and 852 m$\Omega$-cm$^2$, for the small, medium and large devices respectively, an order of magnitude higher than the corresponding values for Sample B.

In order to ensure that the series resistances of the devices were not due to poor (i.e. highly resistive or highly non-ohmic) contacts, the p-type contact resistivities were estimated from circular TLM measurements. The contact resistivities were found to be approximately 5 m$\Omega$-cm$^2$ for Sample A, and less than 1 m$\Omega$-cm$^2$ for Sample B. Since these values are orders of magnitude smaller than the overall differential resistances of the devices themselves, the parasitic resistance of the contacts is not a significant factor.

Since a direct gap p-i-n diode is not expected to experience conductivity modulation under high injection in the lightly doped drift layer, it is also important to consider the effect of the drift layer resistance on the overall series resistance of the device. If the overall device differential resistances were due solely to the resistivity of the lightly doped drift layers, they would require drift layer resistivities of 81 $\Omega$-cm, 124 $\Omega$-cm, and 168 $\Omega$-cm for the small, medium and large Sample B devices respectively, and 1720 $\Omega$-cm, 1940 $\Omega$-cm, and 2030 $\Omega$-cm for the Sample A devices. However, based on the measured effective doping concentration of $4 \times 10^{15}$ cm$^{-3}$ these resistivities cannot be reached. Assuming a typical bulk electron mobility of $\sim 1000$ cm$^2$/V-s for lightly doped, high-quality MOCVD GaN, the resistivity of Sample B’s drift layer is 7 $\Omega$-cm or less, which corresponds to a specific resistance of only $\sim 3$ m$\Omega$-cm$^2$, far too small to explain the overall series resistance of the device [26–28].

We therefore propose that current crowding in the diode mesas increases the effective series resistances, and that the drift layer resistivity is not the primary contribution to device series resistance. In addition, due to the quasi-vertical nature of the device, the lateral conduction through the bottom thin n+ layer also contributes significantly to the series resistance; even though it is more heavily doped than the drift layer, the n+ layer has lower carrier mobility, smaller cross-section area and longer distance for lateral current conduction.

The breakdown voltage of the fabricated diodes was found to be approximately 300–400 V; however, this breakdown occurs in the dielectric and/or the mesa sidewall, rather than in the semiconductor, due to the lack of field termination structures in the diode design.

4. Conclusion

GaN quasi-vertical p-i-n diodes with p-layers consisting of a 20-nm thick contact layer ([Mg] = 10$^{20}$ cm$^{-3}$) and a 480-nm thick layer with [Mg] = 10$^{18}$ cm$^{-3}$ on heteroepitaxial layers grown by MOCVD on sapphire substrates were fabricated and were found to have superior electrical performance to similar diodes where the p-layer instead had a uniform Mg concentration of $3 \times 10^{19}$ cm$^{-3}$. Low temperature PL suggested that the prevalence of point defects, in particular compensating donors, in the high-low junction sample was less, and XRD measurements showed that the concentration of extended defects was similar between the two samples. The high-low junction diodes have higher forward current and lower forward voltage than the diodes with a uniform p-layer, and a much smaller differential series resistance; this is attributed to increased hole injection from the high-low junction p-layer into the drift region. This conclusion is further supported by the more intense electroluminescence of the high-low junction, indicating increased hole-limited recombination.
Acknowledgements

The authors would like to thank the NSF and New York State NYSTAR, which partially supported this work under cooperative agreement EEC-0812056 and NYSTAR contract C090145. This work made use of Engineering Research Centers Shared Facilities supported by the National Science Foundation under NSF Cooperative Agreement No. EEC-0812056. Additional support was also provided through the Rensselaer Office of Graduate Education. Any opinions, findings, and conclusions or recommendations expressed in this material are those of the author(s) and do not necessarily reflect the views of the National Science Foundation.

References