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# Effect of growth conditions on the conductivity of Mg doped p-type GaN by Molecular Beam Epitaxy

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The performance of III–V nitride heterostructure bipolar transistors has been limited by highly resistive p-type layers, in addition to difficulties associated with a precise p–n junction placement at the emitter-base heterojunction due to the Mg-memory effect during growth by Metal-Organic Chemical Vapor Deposition. The problem of precise p–n heterojunction placement can be solved by Molecular Beam Epitaxy growth. To investigate this possibility, in this work, we present a comprehensive study of the effect of III/V ratio, growth temperature, and Mg doping on the resistivity of Mg doped GaN layers grown by Molecular Beam Epitaxy. $N_2$-rich growth conditions are found to lead to a temperature- independent low hole mobility, as opposed to Ga-rich growth conditions that lead to higher hole mobilities that vary with temperature. In addition, the growth temperature, Ga flux, and Mg flux windows leading to the highest p-type conductivity are identified.



**1 Introduction** III–V nitride semiconductors have attracted a lot of interest for their applications in high-speed high-power electronics and visible and short wavelength optoelectronics. However, the performance of nitride heterojunction bipolar transistors (HBTs) has been limited by the highly resistive p-type layers. This is due to the low hole mobilities combined with a large acceptor activation energy of Mg atoms (~120 meV–200 meV) [1–5], the most common p-type dopant. Recently, most commercial development of nitride has concentrated on growth by Metal Organic Chemical Vapor Deposition (MOCVD). However, long transient times of the Mg precursors in MOCVD result in non-abrupt doping profiles, resulting in difficulties in accurate placement of the emitter base junction in HBTs [6]. Growth of Mg doped GaN by molecular beam epitaxy (MBE) eliminates this particular problem. This is due to the ultra high vacuum used during growth allowing for a hydrogen-free environment, and the ability to create abrupt doping profiles [7].

Resistivities as low as 0.2 Ω cm have been reported by Nakamura et al. for Mg-doped p-type GaN by MOCVD [8]. For similar MBE-grown samples, resistivities as low as 0.6 Ω cm were recently reported [9]. In order to enhance the p-type conductivity in GaN, a detailed investigation of the effects of growth conditions on Mg doping is needed. Haus et al. qualitatively showed that the surface morphology and electrical conductivity of p-type GaN is affected by the III/V ratio and substrate temperature during growth [10]. Smorchkova et al. showed how the Mg concentration in p-type GaN can affect the electrical conductivity for Ga-rich growth conditions [7]. However, there have been no quantitative investigations on the effect of MBE growth conditions on the hole mobilities and concentrations, and their dependence on temperature. In this work, we present a comprehensive study of the effect of III/V ratio, growth temperature and Mg doping on surface morphologies and temperature-dependent hole mobility, hole-concentration, and p-type conductivity of Mg doped GaN. The optimal growth conditions that lead to highest p-type conductivities are identified.

**2 Experiment** Mg-doped GaN samples were grown by plasma-assisted MBE on commercially available Fe-doped semi-insulating GaN templates grown on sapphire.

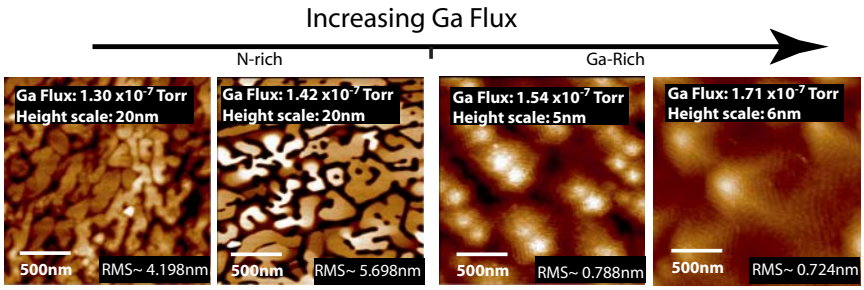


**Figure 1** (online colour at: www.pss-a.com) AFM scans of GaN:Mg samples with different Ga BEP. III/V ratios larger than 1 result in smoother surface morphologies.

The samples consisted of 400 nm Mg-doped GaN followed by 5 nm p++ GaN cap layers for ohmic contacts. $N_2$ plasma power was kept constant at 275 W and the growth rate was 243 nm/hr. In the first series of samples (A), the Mg Beam Equivalent Pressure (BEP) was kept constant at $8.60 \times 10^{-10}$ Torr, the thermocouple temperature ($T_C$) at 600 °C, and the Ga BEP was varied from $1.30-1.71 \times 10^{-7}$ Torr. This corresponds to a change from $N_2$-rich to Ga-rich growth conditions. A second series (B) of samples were grown at the same $T_C$ of 600 °C, a constant Ga BEP of $1.54 \times 10^{-7}$ Torr, and the Mg BEP was varied from $3.2-12.1 \times 10^{-10}$ Torr in order to study the effect of different doping levels on the p-type conductivity. A third series (C) was grown to examine the effect of growth temperature on the p-type resistivity. The BEPs were fixed at $8.60 \times 10^{-10}$ Torr and $1.54 \times 10^{-7}$ Torr for Mg and Ga respectively, and $T_C$ was varied from 540–630 °C across different samples.

Reflection High Energy Electron Diffraction (RHEED) and Atomic Force Microscopy (AFM) were used to study surface morphology during and after growth. Mesa patterns, in the Van der Pauw geometry (inset Fig. 2) were etched into the samples for Hall-effect measurements with a $Cl_2$ plasma in a Reactive Ion Etcher (RIE). Ni/Au metal ohmic contacts were then deposited, and temperature dependent Hall-effect measurements were performed from 140 K to 300 K using a magnetic field of 0.5 Tesla.

**3 Results** Figure 1 shows the surface morphologies of $2 \times 2$ μm regions as measured by AFM for Series A samples grown under different Ga-fluxes. A III/V ratio <1 results in a rough sample surfaces and spotty RHEED patterns. When the Ga BEP is increased above $1.50 \times 10^{-7}$ Torr, the RHEED patterns become streaky, the surface becomes smooth revealing atomic steps, and excess Ga metal droplets were found on the sample surfaces after growth, which were removed in HCl before processing and fabrication. This observation is similar to those reported earlier by Haus et al. [10]. From 300 K Hall-effect measurements, resistivities as low as 1 Ω cm at RT were obtained for Ga-rich growth conditions (Fig. 2), with mobilities $\mu_p \sim 15$ cm$^2$/Vs and carrier concentrations between $p \sim 2.5-3.3 \times 10^{17}$ cm$^{-3}$. For $N_2$-rich conditions, the resistivities are an order of magnitude larger than those grown under Ga-rich conditions. Though similar behavior has been reported earlier, there have been no studies aimed at understanding its origin. To identify the reasons for such behavior, temperature-dependent Hall-effect measurements were performed and compared for the Ga-rich and $N_2$-rich grown p-type layers in series A. The results are shown in Fig. 3.

All samples exhibit carrier freeze out as the temperature is lowered, as seen in Fig. 3(a). Below 140 K, the sample resistivities were too high to obtain accurate resistivity and Hall-effect measurements. Acceptor activation energies $E_A \sim 167$, 120 meV was obtained for the Ga-rich and $N_2$-rich samples respectively. The temperature-dependence of hole mobilities were markedly different for different growth conditions. For Ga-rich samples, $\mu_p$ increased from ~15 cm$^2$/Vs at 300 K to ~25–35 cm$^2$/Vs at 160 K. This clearly indicates the dominance of optical phonon scattering. However, hole mobilities were much lower, in the

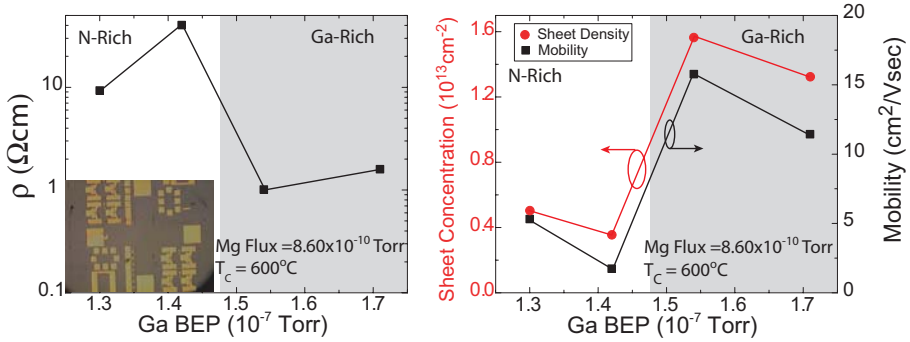


**Figure 2** (online colour at: www.pss-a.com) Room temperature Hall data for Mg doped sample grown under different Ga fluxes. Samples grown under Ga-rich conditions exhibit resistivities one order of magnitude lower than samples grown under N-rich conditions.

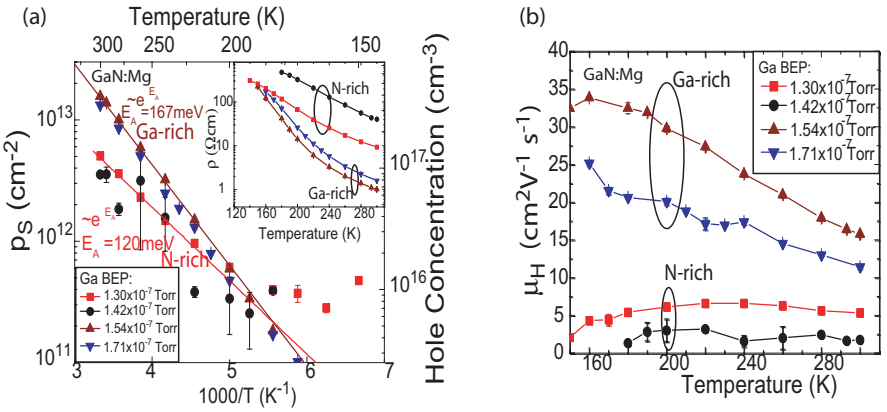


**Figure 3** (online colour at: www.pss-a.com) (a) Temperature dependent carrier concentration of Mg doped GaN. Room temperature concentrations are larger for samples grown under Ga-rich conditions resulting in resistivities an order of magnitude lower (insert). (b) Temperature dependent hole mobilities. N-rich growth resulted in lower mobilities.

range of 2–5 $cm^2/Vs$, and were relatively temperature independent for $N_2$-rich growth. The temperature independence indicates Coulombic scattering by a very high density of charged defects. It has been found earlier that $N_2$-rich growth can lead to the formation of Ga-vacancies [12], and it is also known that Ga-vacancies can be charged to $2^-$ to $3^-$ states [13]. The low temperature-independent hole mobility in p-type GaN grown under $N_2$-rich conditions is attributed to scattering by charged Ga-vacancies. Quantitative calculations are not attempted here to prove this conjecture due to lack of precise experimental knowledge of the hole effective masses.

Changes in effective Mg flux (series B) resulted in no significant change in the surface morphology, as determined by both AFM (Fig. 4), and RHEED. RT resistivities show a minimum of 1 Ω cm for a Mg BEP of $8.60 \times 10^{-10}$ Torr (Fig. 4). Mg BEP higher than this results in lower carrier mobilities due to increased neutral-impurity scattering, as well as increased crystalline disorder caused by Mg clustering [7], whereas Mg doping lower than this BEP results in higher resistivities due to lower hole concentrations. Thus, an optimum Mg BEP exists for achieving the lowest resistivity at any given growth temperature.

When the growth temperature was varied (series C), drastic changes in the sample morphology and resistivities were observed (Fig. 5). Droplets were observed for all samples indicating Ga-rich growth. Mg incorporation is known to be strongly dependent on the substrate temperature [10], with the Mg concentration decreasing exponentially with increasing substrate temperature. For $T_C < 570$ °C, very rough sample surfaces and high resistivities (~230 Ω cm) were observed due to very high Mg concentrations and clustering, lower carrier mobilities and poor morphologies. Samples grown at $570° < T_C < 600°$

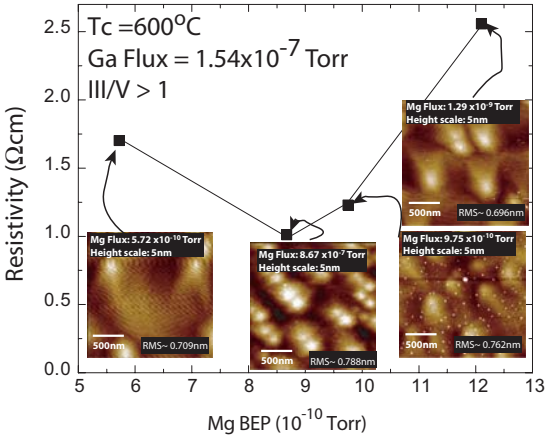


**Figure 4** (online colour at: www.pss-a.com) Room temperature hole resistivities as a function of Mg flux during growth. All samples were grown under Ga-rich conditions. Resistivities as low as 1 Ω cm are obtained for Mg BEP of $8.60 \times 10^{-10}$ Torr. AFM scans for all samples showed atomic steps as evidence of 2-D step growth.

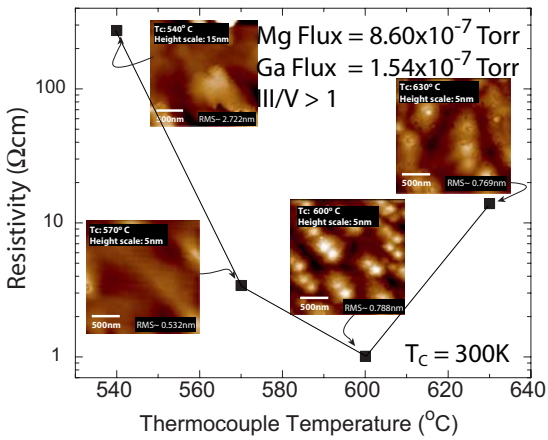


**Figure 5** (online colour at: www.pss-a.com) Room temperature resistivities as a function of thermocouple temperature. 2 × 2 μm AFM scans are shown in insert. Resistivities as high as 230 Ω cm for the rougher samples, and as low as 1 Ω cm for the samples with smooth morphology were obtained.

had atomic steps shown in the AFM and exhibited resistivities as low as 1 Ω cm. When the growth temperature was increased even further, the surface morphology remained smooth and exhibited atomic steps in the AFM (Fig. 5), but the resistivity increased to ~11 Ω cm due to decreasing Mg incorporation. Thus, a growth-temperature window clearly exists for achieving highest p-type conductivities by MBE.

**4 Conclusion** Mg-doped p-type GaN resistivities are shown to have a strong dependence on the III/V ratio, substrate temperature, and Mg doping during MBE growth. The lowest resistivities are found under Ga-rich growth conditions, $T_C \sim 600$ °C and Mg BEP of $8.60 \times 10^{-10}$ Torr, which resulted in a resistivity of ~1 Ω cm. Our results indicate that Ga-rich growth conditions are necessary for highly conductive p-type layers, but it has also been shown that such growth conditions lead to higher leakage currents in vertical devices due to Ga-filled dislocations [14]. Therefore, for vertical devices employing p–n junctions, modulated Ga- or $N_2$ fluxes or intermediate growth regimes would be necessary to prevent leakage, and at the same time maintain a high p-type conductivity.

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