Phase degradation in $B_xGa_{1-x}N$ films grown at low temperature by metalorganic vapor phase epitaxy

Brendan P. Gunning, Michael W. Moseley, Daniel D. Koleske, Andrew A. Allerman, and Stephen R. Lee

Using metalorganic vapor phase epitaxy, a comprehensive study of $B_xGa_{1-x}N$ growth on GaN and AlN templates is described. BGaN growth at high-temperature and high-pressure results in rough surfaces and poor boron incorporation efficiency, while growth at low-temperature and low-pressure growth (750-900°C and 20 Torr) using nitrogen carrier gas results in improved surface morphology and boron incorporation up to ~7.4% as determined by nuclear reaction analysis. However, further structural analysis by transmission electron microscopy and x-ray pole figures points to severe degradation of the high boron composition films, into a twinned cubic structure with a high density of stacking faults and little or no room temperature photoluminescence emission. Films with <1% triethylboron (TEB) flow show more intense, narrower x-ray diffraction peaks, near-band-edge photoluminescence emission at ~362 nm, and primarily wurtzite-phase structure in the x-ray pole figures. For films with >1% TEB flow, the crystal structure becomes dominated by the cubic phase. Only when the TEB flow is zero (pure GaN), does the cubic phase entirely disappear from the x-ray pole figure, suggesting that under these growth conditions even very low boron compositions lead to mixed crystalline phases.

Keywords: B1. Nitrides; A3. Metalorganic vapor phase epitaxy; A1. Crystal structure; A1. X-ray diffraction; A1. Characterization; A1. Defects

1. Introduction

Gallium nitride (GaN) and its related semiconductor alloys have garnered substantial interest in recent years for high-power electronics owing to their large band-gap energies and high critical electric fields.[1, 2] However, due to the lack of lattice-matched substrates, III-nitride based power-electronic devices such as p-i-n diodes have primarily been focused on GaN alone where bulk substrates have become more available.[3-5] One way to achieve lattice match to alternative substrates, such as AlN or SiC, is through the addition of boron to create a $B_xGa_{1-x}N$ alloy. At boron compositions of 12% and 17%, BGaN can be lattice-matched to AlN and SiC, respectively, potentially enabling low-defect density ultra-wide bandgap power electronics.[6] Moreover, forming a quaternary alloy of $B_xAl_yGa_{1-x-y}N$ can reduce the boron composition required for lattice matching while simultaneously increasing the band gap to achieve higher breakdown voltages.

To date there are several reports of BGaN, BAlGaN, and BAlN growth; however, the topic remains relatively unexplored due to the substantial challenges in the growth of these materials.[7-12] The boron incorporation efficiency into GaN is generally low, particularly at growth conditions desirable for high quality films.[6, 13] Moreover, phase separation by spinodal decomposition is expected to occur at boron compositions above 5%.[14] Thus, reports of boron compositions in BGaN are typically limited to the 0-3% range. Nonetheless, some groups have achieved higher boron compositions with quantum well or superlattice structures by using pulsed growth techniques or through the use of BAlN or BAlGaN alloys.[7, 8, 15-17] In this study, we explore boron incorporation in BGaN under wide-ranging growth conditions and report on the resulting surface morphology and crystal structure.

2. Experimental

All films in this study were grown by metalorganic vapor phase epitaxy (MOVPE) in a short-jar Veeco D-125 reactor. The precursors for the growth of the BGaN films were trimethylgallium (TMG), triethylboron (TEB), and ammonia. The nominal growth conditions included substrate temperatures between 750-1040°C, ammonia flow of 7-14 standard liters per minute (SLPM), H₂ and N₂ flows of 0-14 SLPM (total H₂ + N₂ = 14 SLPM), TMG molar flow of 83-205 $\mu\text{mol/minute}$, and TEB molar flow of 0-9 $\mu\text{mol/minute}$. Growth pressure, carrier gas flows, substrate temperature, and V/III ratio were all varied in order to study the boron incorporation and the resulting structural and morphological properties. The substrates used for this study were 1-3 μm thick AlN-on-sapphire templates or ~ 5 μm thick GaN-on-sapphire templates, both grown by MOVPE. The BGaN films were monitored *in situ* by multi-wavelength reflectance and emissivity-corrected pyrometry (kSA ICE system). All substrate temperatures discussed herein are those measured by a near-infrared pyrometer corrected for the susceptor and window emissivities. Post-growth morphology analysis was performed using optical microscopy as well as atomic force microscopy (AFM, Veeco Dimension V), with the AFM operated in tapping mode. Photoluminescence (PL, Nanometrics RPM2000) was performed at room temperature using a 266 nm or 325 nm laser.

The microstructure of select BGaN films was analyzed by transmission electron microscopy (TEM, Evans Analytical Group). Several samples were also characterized by secondary ion mass spectrometry (SIMS, Evans Analytical Group) to determine boron composition in films with less than 1.5% boron incorporation, or by Rutherford backscattering with nuclear reaction analysis (RBS-NRA, Evans Analytical Group) for a film with a boron composition of more than 7%. X-ray diffraction (XRD, Panalytical X'Pert Pro MRD) was used

extensively in this study to determine crystalline phase and material quality. The XRD detector configuration was varied depending on the scan type, with the scans using a 3-bounce analyzer crystal, a detector-slit collimator, or an open detector. Coupled 2θ - ω scans were used in both symmetric and asymmetric geometries to measure lattice constants, infer strains and compositions, and identify phases. Peak widths measured in ω scans were used to determine crystal quality.[18] The c and a lattice constants extracted from combined (0004) and (10-11) 2θ - ω scans, were used to estimate the boron composition and strain in the BGaN films grown on AlN templates.[19, 20] It should be noted that the poor diffracted-beam intensity for the (10-11) reflections, the sparse data for wurtzite-phase boron nitride elastic coefficients,[21] and the possible departure of BGaN-alloy lattice constants from Vegard's law combine to produce significant uncertainty in the boron compositions determined from the XRD; composition results should therefore be treated as approximate. Finally, because the (0002) wurtzite and the (111) zinc-blende reflections in (0001)/(111)-oriented GaN produce nearly identical scattering geometries, traditional XRD analysis is insufficient for separating these two crystal phases. Thus, x-ray pole figures were employed to unambiguously elucidate the formation of cubic and hexagonal phases.

3. Results and Discussion

3.1. BGaN growth on GaN/sapphire templates

The preliminary growth of BGaN on GaN/sapphire templates was performed using a well-optimized GaN template-like condition with a pressure of 500 Torr, a substrate temperature of 1040°C, and gas flows of 15 SLPM H₂, 6 SLPM N₂, and 7 SLPM NH₃. The TEB/(TEB+TMA) molar flow ratio in the gas phase was ~1-3%, and the total group III molar flow was

approximately 205 $\mu\text{mol}/\text{min}$ resulting in a V/III ratio of ~ 1500 . Based on the XRD analysis (not shown), the BGaN films grown at this high temperature, high-pressure condition contained little to no boron and were visibly rough under an optical microscope.

To improve boron incorporation, the group III flows were decreased and the substrate temperatures were reduced to 800-900°C while maintaining a growth pressure of 500 Torr. At these lower temperatures, the resulting films demonstrated shoulders to the right of the (0002) GaN XRD reflection, as shown in Figure 1. Several representative samples were analyzed with SIMS and the boron composition was found to be approximately 0.3% to 1.1%, in agreement with XRD analysis of the (0002) reflection using Vegard's law for wurtzite GaN and BN assuming complete strain relaxation in the BGaN film. While these growth conditions resulted in specular films, morphological degradation was observed above $\sim 2\%$ boron incorporation. A reduced growth pressure of 75 Torr was then used to achieve a boron composition of approximately 3% with improved morphology.

3.2. BGaN growth on AlN templates on sapphire

Using the improved surface quality and boron incorporation developed on GaN templates, we then adopted AlN templates with the goal of achieving a lattice-matched condition with $\sim 12\%$ boron composition. Starting from the best conditions above (800°C, 75 Torr, 15 SLPM H_2 , 6 SLPM N_2 , 7 SLPM NH_3 , and $\sim 90 \mu\text{mol}/\text{min}$ total group III flow), a more systematic approach was taken to increase the boron concentration and improve the surface morphology and crystal quality. The first series of growths kept the same conditions as mentioned above and used a TMG flow of 83.3 $\mu\text{mol}/\text{min}$, but varied the TEB flow from 1.75 to 7 $\mu\text{mol}/\text{min}$ to deliver

2.1% to 7.7% TEB in the gas phase. For the BGaN films grown on AlN templates, the significant separation between the BGaN and AlN XRD reflections allowed for a more robust XRD compositional analysis from the (0004) and (10-11) peak positions and assumed elastic coefficients.[19, 20] The resulting BGaN film compositions are shown in Figure 2 versus the TEB flow. For increasing TEB flow, a sub-linear increase is observed in the boron composition. This observation suggests that either the boron incorporation efficiency decreases, or the boron incorporates into a different phase such as BN inclusions/crystallites or cubic BGaN as discussed later in section 3.4. Even so, doubling of the ammonia flow increases the boron incorporation, which suggests a possible NH_3 dissociation deficiency which limits N incorporation due to the lower growth temperature used for these films. With an increase in reactive nitrogen, more boron can be kinetically trapped into the growing film, similar to what is seen with InGaN growth at higher V/III ratios.[22]

While the boron compositions achieved above are high compared to most reports in the literature, the surface quality was very poor for boron compositions in excess of ~2.5% when using 7 SLPM NH_3 flow as shown by the microscope images in Figure 2. In contrast, a lower, ~2% boron concentration gave a specular surface with sparse pitting, while a higher, ~3.1% boron concentration gave a degraded, visibly rough surface marked by large yellow clusters and many small dark spots. Increasing the NH_3 flow to 14 SLPM not only increased the achievable boron concentration, but also improved the surface morphology such that a smooth but slightly pitted surface was present even for a high boron composition up to ~3.7%.

3.3. *Effect of H_2/N_2 carrier gas on surface morphology and boron incorporation*

Having achieved smooth surfaces with boron compositions up to ~4.2% at 800 °C and 75 Torr, the H_2/N_2 gas flows were then varied. Using a TMG flow of 83.3 $\mu\text{mol}/\text{min}$, a TEB flow of 7 $\mu\text{mol}/\text{min}$, and a NH_3 flow rate of 14 SLM, the H_2 flow was varied from 2 to 14 SLPM with the corresponding N_2 flow varied from 12 to 0 SLPM such that the combined $H_2 + N_2$ flow was fixed at 14 SLPM. Figure 3 shows the resulting boron composition as a function of the H_2/N_2 gas mixture along with selected optical microscope images. The figure shows that the gas flow configuration did not dramatically affect the boron incorporation. However, variation of H_2/N_2 flow did strongly influence the surface morphology, with increased H_2 resulting in a rougher surface and increased N_2 resulting in a smoother but still heavily pitted surface. This observation might be due to H_2 reactivity with the growth surface, resulting in roughening at higher H_2 flows due to possible gas phase transport of Ga atoms and pitting at lower H_2 flows from reduced gas-phase and surface adatom transport.[23]

As H_2 is reduced in the H_2/N_2 flow mixture, we observed that the (0002) rocking curve (omega scan) became better defined, as shown in Figure 4. With a high H_2 flow of 11 SLPM, the (0002) omega scan shows a very broad, poorly defined peak with a FWHM exceeding 1° . On the other hand, with a low H_2 flow of just 2 SLPM, the omega scan shows a narrower, better-defined peak and reduced background intensity. The comparison suggests that high H_2 flow leads to BGaN grains poorly aligned to the c-axis.

To further increase boron concentration, the substrate temperature was next reduced from 800 to 750°C and the H_2/N_2 gas flow study was repeated. As shown in Figure 5, two different pressures were also used with slightly higher boron incorporation observed at 30 Torr compared to the previous 75 Torr. All films were predominantly smooth and specular as shown by the

images in Figure 5. At both pressures, the H_2/N_2 gas flows had only a small effect on the boron incorporation. By comparing the microscope results at 800 °C (Figure 3) to those at 750 °C (Figure 5), we also see that both lower temperatures and lower pressures tended to improve the final surface morphology.

A H_2 flow of 5 SLPM at 30 Torr resulted in a featureless surface and a boron composition of 5.0% which represents, to our knowledge, one of the highest reported for BGaN single layers (i.e. not superlattice or quantum wells) grown by MOVPE. Continuing the H_2 flow trend in Figure 4, the samples grown with low H_2 flows in this series similarly exhibited sharper, more clearly defined peaks in the (0002) omega scans suggesting improved crystalline alignment to the c-axis. The full width at half maximum (FWHM) of the (0002) omega scan was ~650 arcsec for the growth at 750°C and 30 Torr with 5 SLPM H_2 flow. By further reducing the pressure to 20 Torr and eliminating hydrogen flow altogether, the surface remained smooth and featureless under optical microscope while the omega scan FWHM narrowed to just 370 arcsec. Maintaining 20 Torr with no hydrogen flow, but with a slight increase in substrate temperature to 800°C, the (0002) omega scan FWHM narrowed to just 250 arcsec while maintaining a smooth and featureless surface. A 350 nm thick sample grown under this condition was analyzed by RBS-NRA and the boron composition was found to be ~7.4%. The discrepancy in boron composition between XRD and RBS-NRA is most likely due to a change in the crystal orientation or phase, as will be discussed later, leading to differences in boron incorporation

3.4. *Identification of wurtzite-to-cubic phase transition*

Based on the growth conditions above (800°C, 20 Torr, no H₂) with the narrowest rocking curve linewidths, we repeated the BGaN growth, varying the time from 4 minutes to 2 hours at a growth rate of ~300-350 nm/hour. Figure 6 shows 5x5 μm AFM images for BGaN films grown at thicknesses of ~20 and ~700 nm. Note that the terraces in Figure 6(a) are from step-bunching in the underlying AlN template and are not due to the BGaN growth. At a thickness of 20 nm, the surface morphology is smooth on top of each terrace but exhibits a fine-grained structure with average grain sizes on the order of 40-80 nm in diameter. Figure 6(b) shows that as the film thickness increases the BGaN film undergoes substantial roughening. With increasing thickness from 20 nm to 700 nm, the surface roughness rises from ~2.5 nm RMS on the flat, terraced regions in Figure 6 (a) up to ~20 nm RMS in Figure 6(b). This rapidly increasing roughness with increasing thickness is most likely due to the reduced surface diffusion length produced by the low growth temperature of 800°C. In the 700 nm thick film shown in Figure 6(b), the surface shows a substantially different character with larger, elongated grains exhibiting preferential orientations that may reflect the underlying crystal structure and might suggest the formation of a different crystalline phase.

During XRD characterization of these BGaN films with varied thickness, it was observed that the intensity of the (0002) diffraction peak changed very little with increasing thickness above ~40 nm. An example of this lack of change in XRD intensity and diffraction-peak structure is shown in Figure 7 for the (0002) BGaN and AlN reflections for two films with thicknesses of 40 nm (red) and 700 nm (black). For relatively thin films like these where x-ray absorption and extinction are minimal, the diffracted intensity is expected to increase with thickness. The absence of any increase in the XRD intensity suggests that the additional

thickness does not contribute to the (0001)-oriented wurtzite phase that first nucleates on the template. Since the (111) zinc-blende and (0002) wurtzite peaks for GaN occur at nearly identical angles in the symmetric diffraction geometry, pole figures were performed near the (311) zinc blende and (11-22) wurtzite GaN reflections to determine possible changes to the BGaN phase as the film thickness is increased. Figure 9 shows an x-ray pole figure of intensity vs. χ (radius) and ϕ at a fixed 2θ angle, in this case 69.65° . For this measurement, an open detector configuration was used to boost intensity and increase the acceptance angle such that both the hexagonal and cubic reflections should appear, if present. Using this geometry with an open detector, the strong underlying AlN substrate reflections overlap and obscure any weak hexagonal (0001)-oriented BGaN reflections. Consequently, the described pole figure serves primarily to detect the presence (or absence) of the undesired cubic phase. A pole figure was also performed on a bare AlN-on-sapphire template as a “background” scan to help ensure the correct peak identification.

Figure 8 shows the resulting pole figure for the 700 nm thick BGaN film. The thin concentric circles mark χ angles of 30, 60, and 90 degrees, where the center of the circle represents a χ angle of 0. Comparing to the bare-AlN-template scan (not shown), we find that the peaks at χ angles between 45° and 76° are all attributable to the AlN template layer or the sapphire substrate, as well as any overlapping wurtzite BGaN reflections. Thus, the remaining peaks occurring at χ angles of $\sim 30^\circ$ and $\sim 81^\circ$, combined with the 2θ angle of 69.65° , are attributed to the (3 1 1) and (3 -1 -1) reflections, respectively, for (111)-oriented cubic BGaN. Furthermore, the presence of six-fold rotational symmetry for both reflections points to the presence of twins within the cubic-phase domains of the BGaN film. This (111)-oriented cubic BGaN revealed by the pole figure is unexpected as the cubic structure is not the most favorable

phase of either GaN (wurtzite) or BN (hexagonal or turbostratic).[24, 25] In order to further confirm the presence of the cubic phase and examine the microstructure in greater detail, a 350 nm thick BGaN sample grown under identical conditions was analyzed by cross-sectional TEM. Figure 9 shows two cross-section TEM images for the BGaN film, with Fig. 10(a) showing the overall microstructure of the film and with Fig. 10(b) showing a higher magnification image of the near-surface grain structure.

In Figure 9(a) the surface roughness is evident for this 350 nm thick film in agreement with the roughening trend from Figure 6. The film appears to grow as columnar grains exhibiting clearly delineated grain/domain boundaries, an observation that has been reported in other BGaN work.[16, 26] Looking closely at Figure 9(b), particularly the grain on the left side of the image, diagonal lines are clearly present which would typically be associated with stacking faults. In wurtzite GaN, the close-packed plane is the (0001) plane and no other equivalent planes exist, such that any basal plane stacking faults would appear as horizontal lines parallel to the interface for c-plane growth. However, for the cubic GaN system there are multiple equivalent close-packed planes on which stacking faults can occur.[27] In this case, the diagonal lines are inclined $\sim 70^\circ$ relative to the c-plane of the substrate. For a (111)-oriented cubic film, the close-packed (1 -1 1) plane is inclined 70.5° which is in agreement with the $\sim 70^\circ$ stacking faults shown in Figure 9(b).[28, 29] Moreover, closer examination of the grain on the right side of Figure 9(b) indicates stacking faults inclined $\sim 110^\circ$ relative to the underlying AlN. These stacking faults could occur in the case of a (111) twin crystal rotated by 180° which further explains the six-fold rotational symmetry for the cubic reflections in the x-ray pole figure shown in Figure 8. Finally, a selective area electron diffraction image (not shown) taken from the center of the BGaN film during TEM analysis also indicates a twinned cubic structure.

Altogether, the x-ray pole figures and TEM microstructure analysis clearly reveal the predominantly twinned cubic crystal structure of the BGaN film.

After thorough identification of the BGaN cubic phase as described above, a series of growths was performed to determine whether the cubic phase formation is inherent to the generally low-temperature growth conditions, or is instead due to the boron incorporation in the alloy. For these BGaN films the NH_3 and N_2 flows were fixed at 14 SLPM each, TMG flow = $83.3 \mu\text{mol/minute}$ for 60 minutes of growth at 800°C and 20 Torr. The TEB flow was varied from 0 to 0.42, 0.84, 1.7, 3.5, and $6.7 \mu\text{mol/min}$, resulting in 0, 0.5, 1, 2, 4, or 7.5% TEB in the gas phase, respectively. Pole figures (not shown) were performed using the same scan conditions as above to look for cubic-related diffraction peaks. The same cubic peaks as in Figure 8 were found in every pole figure for this series of growths, except in the case of the purely GaN film. Since a purely GaN film exhibited no cubic peaks, low temperature growth does not explain the cubic-phase formation, but instead appears to be caused primarily by the boron incorporation in the BGaN alloy. For the 0.5% TEB case, the (3 1 1) peak was weak at $\sim 2\times$ the background intensity which may indicate a transition to a dominant wurtzite phase for the very low TEB flows below 1%.

Figure 10 shows x-ray diffraction scans near the (0002) wurtzite reflection for these BGaN films with varied TEB flow. While the general trend in Figure 10 is broader and weaker peaks with increasing TEB flow, the 0% and 0.5% cases are nearly identical in both peak intensity and peak shape, with only a slight increase in Bragg angle due to the incorporation of boron. At 2% and higher TEB flow, the peak intensities are two orders of magnitude lower than for 0% or 0.5% TEB flow, and the peaks are very broad, which suggests poor crystallinity, significant mis-orientation, and/or non-uniform alloy composition. For the film with 1% TEB,

the peak character is a mix of the two distinct peak types discussed above, falling between the low and high boron composition peaks in both width and intensity. This might suggest that the 1% TEB condition represents a transition between the higher-quality wurtzite-dominated films and cubic-dominated films. Further evidence for this higher quality, wurtzite-dominated structure is found in the photoluminescence spectra, shown in Figure 11, where only the films with 0% and 0.5% TEB exhibit strong near-band-edge (NBE) emission at ~362 nm.

Compared to the 0% TEB film, the NBE peak for the 0.5% TEB film broadens from 7.4 nm to 12.1 nm and blue shifts as expected from the slightly larger band gap energy, while the deep-level “yellow luminescence” intensity increases dramatically. For example, with 0% TEB, the NBE intensity is ~12x larger than the deep-level luminescence, while with the addition of just 0.5% TEB, the deep-level emission intensity is ~3x higher than that of the NBE luminescence. This increasing deep-level luminescence suggests increased impurity incorporation or point-defect formation, even with small boron compositions less than 1%. This is consistent with SIMS analysis of our other BGaN films, which show both very high carbon concentrations greater than 10^{19} cm^{-3} and insulating electrical behavior, even at boron compositions less than 1.5%. While we cannot rule out increased carbon incorporation as a contributing factor in the material degradation, it seems unlikely to be a primary cause at the dopant-level concentrations most likely present herein. With increased TEB flow of 1%, the PL spectra becomes dominated by the deep-level luminescence with an intensity 47x larger than the NBE emission, while the NBE emission decreases by more than 20x compared to the 0.5% TEB case.

4. Conclusions

In summary, we have presented a comprehensive study of the growth of $B_xGa_{1-x}N$ by MOVPE wherein a wide range of growth conditions were used to achieve higher boron compositions. Low-temperature and low-pressure growth (750-900°C and 20 Torr) can lead to higher boron compositions up to 7.4% -- but at the cost of a severely deteriorated crystal structure. Structural analysis by TEM and x-ray pole figures reveal a twinned cubic-phase film structure containing a high density of stacking faults in the highest-boron-content BGaN films, with photoluminescence measurements of these defective films exhibiting little or no PL emission. For films with >1% TEB flow in the gas phase, the crystal structure appears to be dominated by the defective cubic phase, whereas for lower TEB flows the primary phase appears to be wurtzite, where the observed recovery of a strong PL signal and sharper, more-intense XRD peaks indicates improved crystal quality. However, only for a purely GaN film with no TEB flow does the cubic-phase contribution completely disappear from the x-ray pole figure. Thus, while the low-temperature growth conditions used herein can result in solid solutions of $B_xGa_{1-x}N$ with relatively high boron compositions, the resulting deterioration in the crystal structure is clearly insufficient for device applications where thick, single-phase, single-crystal films are required.

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Figure 1.

Figure 2.

Figure 3.

Figure 4.

Figure 5.

Figure 6.

Figure 7.

Figure 8.

Figure 9.

Figure 10.

Figure 11.

Figure Captions

Figure 1. X-ray diffraction scans about the (0002) GaN template peak showing distinct peaks or shoulders for BGaN layers grown with boron concentrations of approximately 0.3% and 1.1% grown at 500 Torr, while a reduced growth pressure of 75 Torr enabled a boron concentration of ~3%.

Figure 2. Boron composition in BGaN with varied TEB molar flow rates and ammonia flow rates. Inset: optical microscope images of BGaN surfaces for 1.75 and 7 $\mu\text{mol/min}$ with 7 SLPM NH_3 . (Temperature = 800°C, Pressure = 75 Torr, TMG flow = 83.3 $\mu\text{mol/min}$)

Figure 3. Boron concentration and optical microscope images for BGaN films grown with varied H_2/N_2 gas flows. (Temperature = 800°C, Pressure = 75 Torr, NH_3 flow = 14 SLPM, TMG flow = 83.3 $\mu\text{mol/min}$, TEB flow = 7 $\mu\text{mol/min}$)

Figure 4. BGaN (0002) omega scans for different H_2/N_2 flow conditions

Figure 5. Boron composition and microscope images for BGaN films with reduced growth temperature of 750°C and varied H_2/N_2 gas flows and varied growth pressure. (Temperature = 750°C, NH_3 flow = 14 SLPM, TMG flow = 83.3 $\mu\text{mol/min}$, TEB flow = 7 $\mu\text{mol/min}$)

Figure 6. 5x5 μm AFM images of BGaN films grown to (a) ~20 nm thickness and (b) ~700 nm thickness on AlN templates.

Figure 7. Nearly identical symmetric x-ray diffraction spectra about the (0002) reflection for 40 nm (red) and 700 nm (black) thick BGaN films. The peak at 36° is the (0002) AlN reflection from the underlying AlN template.

Figure 8. Pole figure for 700 nm thick BGaN film grown on AlN-on-sapphire template. The (3 1 1) and (3 -1 -1) zinc-blende BGaN reflections are clearly present at chi angles of ~30° (black arrows) and ~81° (red arrows), respectively.

Figure 9. TEM bright-field images of (a) 350 nm thick BGaN film and (b) close-up image of grain structure with stacking faults.

Figure 10. X-ray diffraction scans about the (0002) wurtzite reflection for BGaN films with varied TEB flow from 0% up to 7.5%.

Figure 11. Photoluminescence spectra of BGaN films with increasing TEB flow from 0% up to 7.5%.