Growth and characterization of green GaInN-based light emitting diodes on free-standing non-polar GaN templates

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Abstract
We demonstrate homoepitaxial growth of GaInN/GaN-based green (500–560 nm) light emitting diodes (LEDs) on a-plane and m-plane quasi-bulk GaN prepared by hydride vapor phase epitaxy (HVPE). We find that in order to achieve an emission peak wavelength beyond 500 nm, a minimum InN-fraction of ~14% is needed for both, a- and m-plane quantum wells (QWs), while ~8% are enough for c-plane-oriented QWs. Besides increasing the InN-fraction in these non-polar QWs, widening the QW also proves to effectively shift the emission to longer wavelengths without loosing efficiency with the benefit of maintaining a low InN-fraction.

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1. Introduction
The growth of GaInN/GaN quantum wells (QWs) as an active region in light emitting diodes (LEDs) on non-polar GaN crystal planes is sought after as a promising approach for stable emission color and possibly improved quantum efficiency due to the absence of the quantum-confined Stark effect (QCE). This approach would be particular beneficial to green LEDs containing QWs of high InN-fractions since a much reduced drift in the emission wavelength is expected when compared to similar structures of strained QWs grown on polar c-plane GaN.

Within the past two years, the development of high brightness LEDs and laser diodes (LDs) on the non-polar m-plane has shown rapid progress in the UV and the blue spectral regions. State-of-the-art devices deliver high light output power and efficiency comparable to those achieved in the conventional polar c-plane GaN [1–6]. A major factor in this success is the introduction of high quality non-polar GaN templates as a substrate for homoepitaxy with a low threading location density typically in the mid $10^6$ cm$^{-2}$ or lower.

As the source of one of the primary colors for solid-state lighting, the green LED is of particular high technical relevance, yet its performance still lags far behind that of UV and blue emitting devices. Here, non-polar growth offers the highest promise for a doubling or tripling of the light emission efficiency in LEDs. In this study, we investigate the epitaxial growth and characteristics of green and deep green LEDs on non-polar a- and m-plane GaN free-standing templates.

2. Experimental procedure
Free-standing non-polar a- and m-plane GaN templates sliced from a centimeter thick c-plane GaN grown by hydride vapor phase epitaxy (HVPE), are utilized as the platform for the homoepitaxial growth of LED structures in metal organic vapor phase epitaxy (MOVPE). Template thickness ranges from 425 to 515 μm while surface dimensions vary from $5 \times 10$ mm$^2$ to $5 \times 25$ mm$^2$. These templates are chemo-mechanically polished with a surface roughness of 0.5 nm root mean square (RMS) or less. Threading dislocation densities are in the mid $10^6$ cm$^{-2}$ or lower. The crystalline orientation is well maintained within 2.0° and 1.5° for a- and m-plane templates, respectively. Templates are n-type conductive with resistivity less than 0.5 Ω cm.

A MOVPE technique with metal-organic sources of Al, Ga and In, as well as NH$_3$ and Si and Mg dopants are used in forming epitaxial LED layers including an active region of GaInN/GaN-based heterostructures sandwiched between 1-μm-thick n-GaN...
and a p-side consisting of 20 nm of p-AlGaN and 0.2 μm of p-GaN. The active region consists of 5 pairs of undoped GaInN wells and nominally undoped GaN barriers. The well thickness is 3–6 nm while the barrier is 20–25 nm in thickness. The growth conditions of these multiple QW (MQW) structures are optimized for the suppression of V-defects [7]. The residual density of V-defects in the active region is consistently below 10^6 cm^-2. The surface roughness of the active region is 2–3 nm (RMS) with a z range of 10–20 nm. These roughness values are still relatively large compared to the 0.1–0.3 nm (RMS) obtained for c-plane GaN. In terms of the threading dislocation density, these homoepitaxial structures well replicate the quality of the bulk GaN template.

In terms of the threading dislocation density, these homoepitaxial structures well replicate the quality of the bulk GaN template. In Fig. 2, the 2θ-XRD scan of an n-plane GaInN/GaN MQW sample has been published in Ref. [8]. In XRD, the zero-order satellite peak is controlled by the average distance of the lattice planes

\[ d_{\text{MQW}} = \frac{d_w Z_w + d_b Z_b}{Z_w + Z_b} \]  

where \( d_{\text{MQW}} \) is the average d-spacing of the MQW, \( d_w \) and \( d_b \) are the d-spacing and thickness of the GaInN well, respectively, \( Z_w \) and \( Z_b \) are the d-spacing and thickness of the GaN barrier, respectively. \( d_{\text{MQW}} \) and \( d_w \) have been derived from XRD while \( Z_w \) and \( Z_b \) have been derived from high-resolution TEM. The equation is then solved for \( d_w \) describing the GaInN alloy. The so-derived value \( d_w \) is then

\[ \sigma_{xx} = C_{13} \epsilon_{xx} + C_{15} \epsilon_{yy} + C_{15} \epsilon_{zz} = 0 \]  

\[ \sigma_{yy} = C_{12} \epsilon_{xx} + C_{11} \epsilon_{yy} + C_{13} \epsilon_{zz} = 0 \]  

where \( \epsilon_{xx} \), \( \epsilon_{yy} \), and \( \epsilon_{zz} \) are strain values along c-, a-, and m-axis, respectively. \( C_{ij} \) (i, j = 1, 2, and 3) are the stiffness constants [10]. E.g., in the case of a-plane GaN, the in-plane strain values parallel to and perpendicular to the c-axis are assigned as \( \epsilon_{xx} \), \( \epsilon_{yy} \), and \( \epsilon_{zz} \), respectively, while the in-plane strain perpendicular to the c-axis in m-plane GaN is assigned as \( \epsilon_{yp} \). Since an MQW on both types of non-polar templates is coherently grown on the underlying GaN template, as shown in Fig. 3, the GaInN well is pseudomorphically compressed so that the in-plane d-spacing values match those of the underlying layer in both in-plane directions. Therefore, the in-plane strain values for each type of non-polar GaInN can now be obtained. The strain in the growth direction is then estimated from \( \epsilon_{yp} \).

For the Ga1-xInxN alloy, all stiffness constants \( C_{ij} \) are linear interpolations of the values of GaN and InN as follows:

\[ C_{ij}(x) = x C_{ij-\text{InN}} + (1 - x) C_{ij-\text{GaN}}. \]  

where \( C_{ij-\text{InN}} \) and \( C_{ij-\text{GaN}} \) are stiffness constants of InN and GaN, respectively. Here we use the experimental values of both \( C_{ij-\text{InN}} \) and \( C_{ij-\text{GaN}} \) as published in Ref. [11]. By solving Eqs. (3) and (4), the InN-fractions in the a- and m-plane GaN well layers are experimentally derived.

The photoluminescence (PL) peak emission wavelength under various InN-fractions and GaN/GaN QW thickness \( (Z_w) \) is shown in Fig. 4. We find, that in order to achieve green light emission, i.e. an emission peak wavelength beyond 500 nm, a minimum InN-fraction of ~14% is needed for 3-nm-thick a-plane QWs and ~14% for 4-nm-thick m-plane QWs, while ~8% are enough for a 3-nm-thick c-plane QW. Longer emission wavelengths can be achieved by either increasing the InN-fraction or by increasing the well width. In the case of the c-plane polar material, this leads to the well-known reduction in the luminescence efficiency, while in the non-polar systems, efficiency remains mostly unaffected. For instance, a wide a-plane QW of 5.4 nm can deliver a peak wavelength as high as 560 nm with an InN-fraction of 15% and a

![Fig. 1. Atomic force micrographs of the surface morphology in MQWs grown on a-plane GaN/r-plane sapphire (a), p-plane GaN bulk (b), and m-plane GaN bulk (c). White arrows indicate the in-plane c-axis. Scanning area is 5 × 5 μm².](image)

![Fig. 2. XRD scan of m-plane GaInN/GaN from the (1010) diffraction plane.](image)
A wide \( m \)-plane QW of 6.7 nm can emit green light ranging from 510 to 570 nm with an InN-fraction of only 10%. The wide spread of the emission wavelength in such a thick \( m \)-plane QW is likely due to partial relaxation of the GaInN as the QW width reaches the critical layer thickness.

### 3.2. Luminescence properties of non-polar GaInN/GaN MQW and LED

The PL spectra of these non-polar MQW samples exhibit a single emission peak with full-width at half-maximum (FHMW)
values ranging from 35 to 40 nm (Fig. 5). The PL intensity from the $a$-plane MQWs on bulk substrate is comparable to that of simultaneously grown samples on $c$-plane GaN/c-plane sapphire. This intensity is about half of what we achieve in $c$-plane green MQWs. Comparing the PL peak intensity of the $a$-plane MQW on bulk substrate to that on $r$-plane sapphire, we find a three-fold higher signal in the low-dislocation-density sample on bulk GaN.

The electroluminescence (EL) properties of the $a$-plane LEDs with dominant wavelength from 520 to 540 nm are shown in Fig. 6. We find that as the injection current is varied from 0.25 to 12.7 A/cm$^2$, the dominant wavelength shows a redshift of less than 10 nm (Fig. 6(a)). Since the forward bias voltage for these LEDs is still quiet high (> 10 V at 12.7 A/cm$^2$), the most-likely mechanism for this shift is thermal heating. The wavelength shift in the $c$-plane LED is typically larger than 15 nm over a similar injection current range. In this polar case, the dominant wavelength generally shifts due to the piezoelectric effect. This therefore indicates that these non-polar $a$-plane LEDs could serve as a green light source with stable dominant wavelength under a wide range of injection currents. The light output power and external quantum efficiency (EQE) as a function of drive current and as measured through the substrate are shown in Fig. 6(b) and (c). In the LED on $a$-plane GaN bulk they are 2.5 times and 2 times as high as in the LED on $r$-plane sapphire at, 12.7 A/cm$^2$, respectively. However, the absolute EQE values are still far smaller than those of $c$-plane green LEDs. The low EQE from these non-polar LEDs may be the result of a lower quality of the MQW and p-layers in this early stage of development of non-polar materials growth.

In our previous work, we found clear evidence that the PL and EL intensities are strongly affected by the density of threading dislocations [13]. Here, the sample of $a$-plane MQW/$a$-plane GaN on $r$-plane sapphire contains a high density of threading and misfit dislocations. Therefore, for the development of highly efficient green light emitters, it is important to develop low-dislocation-density $a$-plane bulk GaN. With further optimization in the growth conditions of these non-polar green MQWs, it therefore is highly likely that EL efficiencies will further improve and likely surpass that of $c$-plane material.

4. Conclusion

We have developed 500–560 nm green GaInN/GaN MQWs and LEDs on non-polar $a$- and $m$-plane GaN templates. In order to achieve green emission in the non-polar QWs at this wavelength range, a higher InN-fraction is required than in polar $c$-plane QWs. Similar to the growth of $c$-plane green LEDs, the performance of these non-polar devices is sensitive to the threading dislocation density in the underlying templates.
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References