Green LED development in polar and non-polar growth orientation

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ABSTRACT

The green spectral region provides a formidable challenge for energy efficient light emitting diodes. In metal organic vapor phase epitaxy we developed GaInN/GaN quantum well material suitable for 500 – 580 nm LEDs by rigorous defect reduction and thrive for alloy uniformity. We achieve best results in homoepitaxy on polar c-plane, and non-polar a-plane and m-plane bulk GaN. By the choice of crystal orientation, the dipole of piezoelectric polarization in the quantum wells can be optimized for highest diode efficiency. We report progress towards the goal of reduced efficiency droop at longer wavelengths.

Keywords: green LED, GaInN-GaN, MOVPE, homoepitaxy, non-polar, efficiency, green gap, threading dislocation

1. THE CHALLENGES OF THE GREEN GAP

Worldwide trade has triggered the unprecedented hope of lifestyle improvements for billions of people. Drastic resource and energy efficiency gains in all aspects of our daily life are urgently required to avoid an ecological collapse. Artificial lighting takes a preeminent role in the global savings potential by technological development. The human eye is most sensitive to light in the green spectral region, yet, state-of-the art artificial lighting is particularly inefficient in this range, aptly termed the "green gap".

Visible light emitting diodes (LEDs) can provide a spectrally narrow (~30 – 50 nm) band of light emission and so suppress the wasteful generation of a huge portion unwanted infrared emission in the case of incandescent lamps. The blue spectral region can most efficiently be generated in AlGaInN-based LEDs,[1,2] while the yellow and red spectral region currently is best addressed with AlGaInP LEDs.[3] Yet, in between for green, the best-most candidates are GaInN/GaN heterostructures of the former type even when their efficiency for photon emission still ranges not higher than some 20%.

For white light generation, the combination of various colors is required which can readily be obtained by the combination of a narrow-band blue LED and a rather broad band yellow phosphor photoexcited by the same blue LED.[4] Such approaches meanwhile can achieve efficiencies beyond 100 lm/W [5] and are the workhorse of all solid-state lighting effort. Yet this approach frequently fails in consumer acceptance due to the incomplete spectral emission pattern of such sources. Similar to compact fluorescence lamps, the absence of green and red spectral portions is poorly received by the selective customer resulting in only a poor penetration of such lamps in the residential market. Recent efforts have made big gains by adding stronger red-emitting phosphors to the lamps yet at a significant cost of lamp power efficiency. These losses are incurred as an inherent problem of the "single-color excitation plus broadband phosphor emitter" approach. The quantum mechanical nature of the photon requires such Stokes losses and amounts to a minimum of 20% for green light obtained by down-converting blue light in even a 100% effective phosphor. Realistic phosphor conversion efficiencies further lower the system performance.

The only feasible alternative is to implement LEDs that directly emit at the desired wavelength, e.g. green GaInN/GaN LEDs. The direct bandgap of Ga_{1-x}In_{x}N spans the entire visible spectral range and beyond from 370 nm (GaN) to 1800 nm (InN) and can be expressed in terms of \( E_g = 3.42 \text{ eV (1-x)} + 1.89 \text{ eV x} - b x (1 – x) \) for \( 0 \leq x \leq 0.2 \), where \( b = 3.2 \text{ eV for luminescence in strained layers} \) and \( b = 3.8 \text{ eV for the absorption bandgap extrapolated to the relaxed} \). Such findings have later been expanded to the entire composition range to InN and can be expressed as \( E_g = 3.42 \text{ (1-x)} + 0.77 x - 1.43 x(1-x) \) for the luminescence gap.[7] We find that the actual InN fraction is best determined...
using x-ray diffraction of the averaged multiple QW layers well under the condition of pseudomorphic growth. Composition values in the literature vary by up to a factor of two and should always be dealt with caution.

Quantum wells (QWs) of typically 3 nm Ga$_{1-x}$In$_x$N embedded in GaN barriers raise the transition energy and assure a strong wavefunction overlap of heavy holes and electrons for high dipole transition matrix elements and high radiative recombination rates. Even under strong piezoelectric polarization of the QWs, equivalent to electric field values of 1 MV/cm, we find that the recombination rate should not drop below half of its value without electric field.[8] In the case of significantly wider QWs, e.g. 10 nm, a relevant separation of electron and hole wave functions must be expected.

For 470 nm blue LEDs, we typically employ InN alloy fractions up to $x = 10\%$ and for 530 nm green and 555 nm deep green LEDs, we do not need more than $x = 20\%$. Yet, there are substantial differences between blue and green emitting LEDs. While for the shorter wavelength blue LEDs, we find a tremendous resilience towards high densities of structural defects such as threading dislocations and resulting alloy fluctuations in the QWs, we find that for the green emitting structures, it is of utmost relevance to reduce the density of structural defects, in particular threading dislocations and any alloy composition fluctuations in the active light emitting region.[9]

2. AVOIDING DEFECTS

Grown in heteroepitaxy on readily available c-plane sapphire, GaN-based structures always show high densities of threading dislocations ($10^9 – 10^{10}$ cm$^{-2}$) as initiated in the interface of both materials.[2] Various efforts of complex growth, processing and re-growth concepts have been developed over time to reduce those densities, yet it remains a fact, that threading dislocations will penetrate the active QW region at a density of at least $10^7$ cm$^{-2}$ and most-likely at significantly higher density. Upon their interaction with the GaInN layers of the QWs, threading dislocations frequently induce the generation of V-shaped defects (Figure 1) causing higher index facet growth at reduced growth rate (Figure 2) and resulting pit formation.[10]

![Figure 1](image1.png)

**Figure 1** Bright field TEM image of 549 nm green LED on sapphire showing the typical formation of V-defects.[10]

![Figure 2](image2.png)

**Figure 2** High resolution TEM images of (a) QWs on the sidewall of the V-defect in a highly defective sample; (b) Normal QW on the c-plane in a sample with low V-defect density. The same holds in the sample of (a) in the areas between the V-defects.[10]
3. HOMOEPITAXY ON A NATIVE SUBSTRATE

Naturally, the best possible approach to avoid the generation of such detrimental V-defects is the reduction of the density of threading dislocations all together. Here we make use of the capability of hydride vapor phase epitaxy to produce thick freestanding bulk GaN templates with a typical density of threading dislocations along the \(c\)-axis and growth axis in the mid \(10^6\) \(\text{cm}^{-2}\).[11]

We therefore developed epitaxial growth processes that, after appropriate surface preparation, allows the replication of the low dislocation density crystal quality into the MOVPE grown GaN layers, such as the n-GaN layers in the 420 nm blue LED shown in Figure 3.[12]

![Figure 3](image)

**Figure 3** Cross-sectional TEM (a) and high resolution TEM (b) images of a blue LED homoepitaxially grown on the \(c\)-plane surface of bulk GaN. The low threading dislocation density of the substrate is well maintained and replicated in the MOVPE grown layers. From TEM micrographs reaching over several \(\mu\text{m}\), a maximum defect density of \(1\times10^8\) \(\text{cm}^{-2}\) was derived. The high resolution image reveals very clear transitions from the lighter GaN barrier areas into the darker appearing GaInN well layers.[12]

The availability of such bulk-like GaN boules also allows the growth on one of the side planes of the wurtzite layers. By growing GaN along the \(c\)-axis into layers of several \(\text{mm}\) thickness, wafer slicing can then be performed along cutting planes that include the crystal growth axis. In this way, all threading dislocations are limited to the substrate and cannot propagate as such into the new overgrowth direction. We succeeded in such homoepitaxial MOVPE growth of GaN along various directions including the non-polar \(a\)-axis (Figure 3) and the non-polar \(m\)-axis (Figure 4) of wurtzite GaN.[13, 14]

![Figure 4](image)

**Figure 4** Cross-sectional TEM micrographs of (a) an \(a\)-axis grown green LED and (b) an \(m\)-axis grown LED. Both are grown homoepitaxially in MOVPE on the side planes of \(c\)-axis grown bulk GaN. In larger field scans, virtually no threading dislocations can be observed penetrating the active QW regions, respectively. Furthermore, no in-plane defects can be observed in the surface morphology of the resulting layers (< \(10^5\) \(\text{cm}^{-1}\)).[13,14]
This is in contrast to the achieved quality in direct heteroepitaxial growth of $a$-plane GaN material on $r$-plane sapphire. Figure 5 shows a typical case of such heteroepitaxy in cross-sectional TEM. Dislocations densities of the order of $10^{10} \text{cm}^{-2}$ are commonplace.

![Figure 5 Cross-sectional TEM micrograph of an $a$-axis grown green MQW sample on $r$-plane sapphire. By this continuous growth along a single axis, a high density ($\sim 10^{10} \text{cm}^{-2}$) of threading dislocations is observed similar to conventional $c$-axis growth.[13]](image)

### 4. DEFECTS IN THE ACTIVE REGION

Epitaxy on even the most perfectly dislocation-free GaN may lead to the generation of misfit and threading dislocations during the growth of quantum wells. We find that the transition from MOVPE grown GaN to MOVPE grown GaInN layers can induce dislocation densities of the order of $10^9 \text{cm}^{-2}$ (Figure 6).[15] This is in spite of MOVPE homoepitaxially grown n-GaN layers perfectly replicating the low dislocation density of HVPE-grown GaN substrates with typical threading dislocation density of $10^7 \text{cm}^{-2}$. This propensity is strongly dependent on the achieved emission wavelength of the QWs. Virtually no new defects may arise in 420 nm material (see Figure 3), while structures of 530 nm may exhibit dislocation densities of the order of $\sim 4 \times 10^9 \text{cm}^{-2}$. In recent work we find indications that the generation of such defects coincides with the relaxation of biaxial strain induced by the QW growth.[15]

![Figure 6 Cross-sectional TEM micrograph of green LED on bulk GaN taken along $[\bar{1}1\bar{2}0]$ zone axis. Pairs (1,3,4) and individual branches (2) of inclined dislocations were observed to be generated within the first three QWs. All of them are tilted a certain angle off [0001], the growth direction.[15]](image)

Nevertheless, we have succeeded in suppressing the generation of V-defects at the intersection of threading dislocations with QWs to a very large extent in green LEDs up to at least 570 nm. Figure 7 shows the cases of a 545 nm (a) and 565 nm (b) deep green LEDs on sapphire. While a high density of threading dislocations reaches the QWs from the n-layers and additional dislocations are initiated in misfit defects in the QWs themselves, no generation of V-defects and associated disturbance of the QW growth can be observed. Atomic force microscopy of the last barrier of the active QW region shows substantially reduced surface roughness when V-defect generation can be suppressed (Figure 8).
Figure 7 Cross-sectional TEM micrograph of (a) 545 nm and (b) 560 nm deep green LEDs on sapphire. In both cases, threading dislocations densities in the QWs reach up to some $4 \times 10^{9}$ cm$^{-2}$ without generating any V-defects.

Figure 8 Surface morphology in atomic force microscopy of the last barrier in V-defect-free (a) and defective (b) GaInN/GaN MQW for green LED. The roughness in the better sample is (a) 0.56 nm (RMS) compared to (b) 4.54 nm (RMS) in the defective one.

Light output power in devices show substantial improvements as the number of V-defects can be reduced. Figure 9 shows the light output versus drive current in 530 nm LEDs in 1 mm diameter scratch diode arrangement.[16] Over the course of many of our LED fabrication runs, the light output power versus current density proves to be an excellent indicator of final LED performance after die fabrication in (350 µm)$^2$ devices.

Electro-optical characterization of the resulting 555 nm deep green LEDs on c-plane sapphire after mesa fabrication is shown in Figure 10 as a function of current density in (350 µm)$^2$ and (700 µm)$^2$ devices.[17] At this stage, no efforts have been made to enhance the light extraction efficiency by e.g. surface patterning or dielectric encapsulation. The dominant wavelength shows the familiar blue shift as the current density is increased. At 35 A/cm$^2$, the typical operation condition of (1 mm)$^2$ 350 mA devices (1W lamp), a dominant wavelength of 555 nm is reliably obtained (Fig. 10 a). The partial light output as measured through the substrate into a 1 cm orifice reaches 5 mW in the smaller device and 20 mW in the larger one (Fig. 10 b). It is well established, that upon standard measures to enhanced light extraction efficiency, a factor of three to five higher output power can be achieved. The external quantum efficiency (EQE) as measured under the same rudimentary extraction conditions is strongly dependent on current density. Figure 10 c) shows EQE under pulsed LED operation conditions (pulses of 50 µs duration at 1% duty cycle). Maximum values are reached at very low current densities, here up to 15%, while at 35 A/cm$^2$, EQE drops to some 11%. In individual dies, a low current maximum of EQE as high as 40% has been observed, possibly indicating some non-linearities in either aspect of the radiative recombination, light extraction, or detection scheme.
Figure 9 Partial light output versus current density in 1 mm diameter devices of 520 nm LEDs as measured through the substrate of the device. Samples without V-defects show higher output and output increases with numbers of QWs. Samples with V-defects show the lower output and output decreases with higher number of QWs.[16]

Figure 10 Partial light output versus current density in mesa-fabricated deep green LED devices of (350 µm)² and (700 µm)² size grown in c-plane epitaxy on sapphire. a) The dominant wavelength shows the well-known blue shift yet still reaches 555 nm at 1W lamp typical 35 A/cm². b) Under pulsed operation conditions, at partial light output power of the bare fabricated devices reaches 70 mW in the larger die and 45 mW in the smaller ones. c) The corresponding external quantum efficiency drops from values of some 15 % to some 3 %.[17]
5. PIEZOELECTRIC POLARIZATION

The uniaxial nature of the wurtzite group-III nitrides in combination with particularities of the strong and partially ionic bonds lead to strong piezoelectric polarization in those materials.[18,19] Combining layers of different polarization and/or biaxial strain condition can lead to discontinuities of polarization which can be expressed by planes of sheet charges of the order of $\pm 10^{14}$ cm$^{-2}$ at the interfaces.[20] For the typical case of pseudomorphically strained GaInN well layer embedded between relaxed layers of GaN barriers, grown along the positive $c$-axis of GaN, a negative sheet charge is formed at the interface from barrier to the GaInN well, paired by a positive sheet charge of identical density at the ending of the well, i.e. at the transition from the well to the GaN barrier (Figure 11).[21] This pair of charges forms an electrostatic dipole that strongly interacts with the injected non-equilibrium carriers. An immediately apparent consequence is induction of an effective bandoffset between both GaN barrier layers, given by the energy of the piezoelectric dipole.[22]

![Realistic electronic bandstructure of a GaInN/GaN quantum well as used in polar $c$-plane LEDs. The polarization dipole seen in the bandoffset across the well, apparently translates into a Stokes shift in the quantum well between absorption edge and peak emission (schematically shown).[21]](image-url)

This quantity also proves to be a very good approximation of the frequently observed huge Stokes shift between optical absorption bandgap and the energy of the luminescence peak.[21, 23] This discrepancy can amount to the large value of 300 meV in green LEDs, i.e. some relevant 15% of the photon energy.[24] A tentative explanation of the mechanism can be found in the consideration of an electrostatic charging of the local piezoelectric dipole in the instant of radiative recombination of the pair of screening electron and hole (figure 11). It can be argued that the mechanism of a well-defined Stokes shift in the recombination process of GaInN/GaN QW structures is a relevant advantage for the efficient light generation in group-III nitride heterostructures.

In the literature, commonly the case is made that the Stokes shift results from a large non-uniformity of the GaInN-alloy, essentially providing laterally separated zones of different bandgap energies. Such an interpretation can find support in the fact that phase separation in In-rich GaInN alloys is frequently observed in growth processes not optimized for lateral uniformity. We maintain that all of our above mentioned approaches to reduce the generation of structural defects include the efforts to reduce spatial non-uniformity of the alloy constituents as far as possible.

Trying to establish the case of the relevance of piezoelectric polarization control in group-III nitrides, this group of authors have found open ears by virtue of the argument, that in wide QWs in the presence of a large piezoelectric dipole, the electron and hole wavefunctions may have only a minimal spatial overlap. As a consequence, the dipole transition matrix element should be reduced, rendering associated recombination processes extremely inefficient. This could potentially be considered a limiting factor for the observed reduced efficiency of green and deep green GaInN/GaN LEDs when compared to their blue counterparts. In order to reach the green spectral region, one would either have to increase the InN-fraction in the QW alloy and by these means increase the piezoelectric dipole, or widen the QW which...
also increases the piezoelectric dipole. By the same token, and approximated by the same quantity, the dipole transition matrix element would be reduced and could so explain the conflicting challenge of the green gap.

6. NON-POLAR GROWTH

To either circumvent the problem of the electron-hole separation within the electric dipole of a QW or at least to reduce its magnitude and impact, epitaxial growth of QW structures along crystallographic axes of GaN with either reduced or absent piezoelectric polarization is highly desirable.[25] Figure 12 gives a schematic of the polar c-plane, and non-polar m-plane and a-plane of the wurtzite structure.[26] Shown also are the respective bandedge line-ups in QWs grown on such planes resulting in either fast or slow recombination processes.

![Figure 12](image)

**Figure 12** Schematic of the bandstructure of GaInN/GaN quantum wells grown along polar and non-polar crystal axes (a, c, e) and the associated hexagonal crystal planes (b, d, f). In polar c-plane growth (b), radiative recombination may be reduced due a lateral offset of the charges (a). Higher efficiency, particularly in the green can be expected from non-polar m-plane (d) and a-plane (f) growth.[26]

The absence of a piezoelectric dipole in QW structures grown along non-polar axes has significant ramifications also for the achievable emission wavelengths in LEDs. Figure 13 shows the photoluminescence spectra of a c-axis grown polar and m-axis grown non-polar multiple QW LEDs. Both structures have been grown simultaneously under identical offered growth conditions.[26] For this, a template of c-plane GaN on sapphire has been placed next to an m-plane GaN template on bulk GaN. While the polar structure reveals a peak emission wavelength as long as 558 nm, the non-polar one reaches a mere 488 nm. This corresponds to a large discrepancy of 320 meV. From our earlier analysis of the Stokes shift in MQW samples leading to the interpretation of the piezoelectric dipole,[27] we extrapolate a dipole energy of 250 meV for samples of the present case. Apparently, the observed discrepancy of emission wavelengths in polar and non-polar structures can well be accounted for by the expected quantity of the piezoelectric dipole energy in polar structures of such type.

It should be noted that due to the different thermal conductivity of both sample structures, different In incorporation rates into both structures are conceivable. The same experiment performed with two polar sample structures, namely c-plane GaN templates on c-plane bulk GaN and on c-plane sapphire result in peak wavelength variations by not more than 20 nm.[28] Such a component can be a contribution in the remaining discrepancy of some 70 meV.

In our interpretation, the shorter emission wavelength in such non-polar structures therefore should be a direct consequence of the absence of the piezoelectric dipole controlling the radiative emission process in the c-axis polar structures.[26] This observation provides supportive evidence of our related statement that the optoelectronic properties of the group-III nitride structures are predominantly controlled by the piezoelectric properties of the material.[29]
Under high optical excitation density, non-polar \( m \)-plane and polar \( c \)-plane LEDs also show substantially different behavior. Figure 14 shows photoluminescence spectra under 337 nm pulsed excitation. While the polar \( c \)-plane material shows the well-known blue shift of its maximum with increasing excitation level (Fig. 14 a), the non-polar sample maintains its peak wavelength at 488 nm fixed up to the highest levels of excitation employed (Fig. 14 b). This blue shift in the polar structure is commonly ascribed to the quantum confined Stark effect that is gradually being screened by high densities of photo-generated carrier pairs. Upon detailed inspection, however, it becomes apparent, that the longer wavelength peak at 560 nm does not disappear, but it merely loses in relative intensity and the spectrum becomes dominated by the peak near 510 nm at higher excitation. This peak also will be the one that eventually shows stimulated emission under sufficient high excitation. This is in agreement with other detailed reports in the literature.[30]
An alternate explanation for the strongly differing peak wavelengths between polar and non-polar structures has been given by proposing a substantially lower In incorporation rate on the non-polar $m$-growth surface when compared to regular polar $c$-plane growth.\[31\] To verify this possibility, we analyzed the non-polar structures in reciprocal space mapping by x-ray diffraction to derive the quantum well region's average alloy composition. By help of additional cross-sectional TEM, actual thicknesses of wells and barriers were determined to separate the region's alloy average into an actual alloy value for the well. While for $c$-plane grown structures commercial software is generally available to derive the alloy compositions, no such is available for the various other growth axes of the wurtzite system.

The reciprocal space map around the $(21-30)$ and $(20-22)$ lattice diffractions for various $m$-plane structures is shown in Figure 15.\[32\] Under the reasonable assumptions of pseudomorphic growth within the growth plane and absence of shear stress, the growth axis may be assumed to remain strain free.

**Figure 15** Reciprocal space map of $m$-plane GaInN/GaN QW structures revealing pseudomorphic growth along the $a$-axis and $c$-axis. The QW-averaged alloy composition is determined from the separation of the 0th-order alloy peak from that of GaN.\[32\]

By help of this analysis, the alloy composition on the wells was established as summarized in Figure 16 versus the achieved photoluminescence peak wavelength. Unlike the $c$-plane polar material, in this non-polar case, a well width beyond 3 nm can be used to increase the emission wavelength by a reduced quantum size effect. In the polar case, this would have lead to reduced wavefunction overlap of electron and hole and rapidly decreased the radiative recombination rate. Just like in the case of polar $c$-axis growth, the growth of thicker wells, however, is complicated by the GaInN alloy's tendency to segregate In into InN-rich regions effectively spoiling the well's homogeneity. Substantial effort in development of growth processes therefore is necessary to achieve well widths as large as 6.7 nm as shown here. In such structures, with as low an InN fraction as 10\%, luminescence peak emission wavelengths as long as 500 nm and 580 nm have been achieved. For comparison, published data by Tsuda *et al.*\[33\] is included requiring some 17\% of InN to achieve 470 nm blue emission in a 3 nm well.
The same analysis was performed for \(a\)-plane grown structures. Figure 17 shows the reciprocal space mapping for a range of \(a\)-axis grown QWs and LEDs. The alloy composition averaged over QW and barriers is determined from the separation of the 0th-order alloy peak from that of GaN. By help of additional cross-sectional TEM, the well widths and barrier widths are determined to conclude on the actual composition of the well. In this, the absence of shear strain is reasonably assumed resulting in a strain free growth axis, the \(a\)-axis.

The luminescence peak wavelength as a function of so derived well alloy composition is shown in Figure 18. The well thickness is listed as a parameter. Available literature data is included in the graph. A good proportionality between InN-fraction and peak wavelength and a general agreement with the literature data is seen. Interestingly, also in this case, deep green emission as long as 550 nm can be achieved with a mere 15% of InN in a 5.4 nm thick QW.
Figure 18 InN-fractions \( x \) of \( a \)-plane \( \text{Ga}_{1-x}\text{In}_x\text{N}/\text{GaN} \) MQW structures emitting at the indicated photoluminescence wavelength. The well width is a parameter. Emission wavelengths as long as 550 nm can be achieved with 15\% of InN using a 5.4 nm well. Included also is data from the literature, for which well widths are not known. Overall a linear trend can be seen.\[32\]

7. DEVICE PERFORMANCE

LEDs emitting in green spectral region have been achieved, both in \( m \)-axis and \( a \)-axis growth on the bulk GaN template. Figure 19 summarizes device performance of \( a \)-plane LEDs in a direct comparison of an \( a \)-axis LED grown on an \( a \)-GaN template on \( r \)-plane sapphire and one on an \( a \)-GaN template on \( a \)-plane bulk GaN. The spectral performance is shown in Figure 19 a) reaching 3 \( \mu \)W/nm in the bulk-based sample. The dominant wavelength reaches up to 5350 nm in the bulk based sample and interestingly increases slightly with current density, which is opposite to the performance of \( c \)-axis polar material (Fig 19 b). The light output power reaches some 160 \( \mu \)W in the same sample, three times as high as the sapphire based one (Fig 19 c). The external quantum efficiency in these bare dies, as measured through the substrate side, are still quite low and actually do drop with current density (Fig. 19 d). A contribution in this drop is certainly die heating due to the low electrical conductivity of the substrate used here as an n-layer and possibly also due to early development stages of p-type doping in \( a \)-axis growth.

Figure 19 Electro-optical performance of \( a \)-axis grown LEDs on \( a \)-plane bulk GaN in comparison with \( r \)-plane sapphire. Device area is 1 mm diameter. Spectral performance (a), light output power (c) and external quantum efficiency as measured through the substrate side (d) show the superiority of the bulk-based sample. The dominant emission wavelength shows slightly higher values for the sapphire-based sample. Interestingly, the wavelength increases slightly with current.\[13\]
Electro-optical performance of a blue-green m-plane based LED is summarized in Figure 20. The polar c-plane structure shows the well-known blue sift with increasing current density (Fig. 20 a+c), while the non-polar m-plane one shows fixed emission wavelength, here ~490 nm (Fig. 20 b+c). Light output power as measured on bare fabricated but unencapsulated (350 µm)² dies through the substrate shows comparable achievement levels of several mW at 35 A/cm².

![Figure 20](image)

**Figure 20** Electro-optical performance of m-plane LEDs on m-plane bulk GaN in comparison with c-plane LEDs on c-plane bulk GaN. Fabricated dies are (350 µm)² in size. Spectral performance of the polar (a) and non-polar (b) structures show the stable emission wavelength in the non-polar m-plane one. c) Dominant wavelengths versus current density. d) Light output power as measured through the substrate reaches values of several mW at 1-W lamp typical current densities of 35 A/cm².[26]

## 8. CONCLUSIONS

Defect reduction and thrive for utmost alloy uniformity prove to be successful guidelines for the achievement of GaInN/GaN LEDs spanning the green spectral region. Reduction of V-defect decoration of existing threading dislocations, reduction of total number for threading dislocations by homoepitaxy on bulk GaN, replication of the bulk GaN quality in the epitaxial layers are the necessary steps towards this goal. Upon the availability of thick GaN bulk pieces, slices of arbitrary crystal orientation can be prepared[11] and utilized as growth templates for homoepitaxy. By tilting the epitaxial growth direction away from the original substrate growth direction, avoidance of all original threading dislocations becomes possible. By careful optimization of growth conditions, it is possible to avoid the generation of new defects including line defects. We demonstrate achievement of blue-green and green emitting LEDs homoepitaxially grown along non-polar a-axis and m-axis of GaN. We argue that the utilization of non-polar growth directions eliminates the help of the piezoelectric dipole in the Stokes shift of the emission wavelength seen in polar growth orientation. Achievement of similar emission wavelength therefore requires higher InN fractions of the QWs or a reduction of the quantum size effect by the use of wider wells. In absence of this polarization dipole, emission wavelength proves widely unaffected by variation of LED drive current. Achievement of those milestones demonstrate our progress towards closing the green gap of subpar light output efficiency in the green and deep green spectral region.

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30 e.g., poster presentations by Hewlett Packard Research Labs, Yokohama, Japan, presented at The Second International Conference on Nitride Semiconductors - ICNS 97, 27-31 October 1997, Tokushima, Japan.

