

Invited

OMVPE Growth of AlGaInN for UV Light Emitters

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1. Introduction

So far, much of the effort in III-nitride optoelectronic devices has focused on blue light-emitting devices. These devices typically employ AlGaIn and/or GaN as wide bandgap confinement layers while adding indium (In) into GaN as active layers to shift the emission into visible spectrum range ($\lambda > 400$ nm). There is a growing interest in shorter wavelength compact UV ($\lambda < 380$ nm) emitters for fluorescence-based chemical-sensing applications, flame detection, high-density optical storage, and high brightness UV white-light LEDs for efficient lighting using UV emitters together with phosphor coating. To date, there are relatively few reports on GaN-based UV emitters. GaN double heterostructure (DH) diodes with emission peaks centered around 375 [1] and 420 nm [2], respectively. In both cases, the widths of emission peaks are in excess of 250 meV (or 40 nm), suggesting the possibility of recombination through localized centers. Recently Nakamura [3] reported the introduction of In into GaN to enhance the luminescence efficiency; an external quantum efficiency of around 7.5% and a UV (370 nm) light output up to 5 mW was demonstrated from a GaN:In-based DH structure.

Several issues related to the growth of nitride-based UV light emitters will be discussed in this presentation. In contrast to the widely pursued, InGaIn-based, blue LEDs, UV LEDs mandates the use of an active layer of little or no indium. To provide comparable level of electrical and optical confinement, one has to employ the ternary AlGaIn with a higher fraction of Al and an increased thickness. Two issues which will be addressed in the talk, therefore, are the optical efficiency of GaN and GaN:In for UV emission, and the control of strain (due to tensile mismatch between AlGaIn and GaN) to avoid the generation of cracking.

2. Results and Discussion

We reported earlier [4] that the use of AlGaIn/GaN multiple quantum well (MQW) structures has led to the demonstration of UV LED operation at a wavelength of around 355 nm. The optical output power of these prototype devices, however, was about two orders of magnitude lower than the commercial high brightness blue LEDs. In this work the optical efficiency of the GaN active layer was investigated. The low-pressure condition employed for the earlier UV LED growth, a condition selected to facilitate mainly the incorporation of AlGaIn barriers, tends to yield very weak RT PL intensity. It is found that the emission efficiency of GaN can be improved by more than ten times, as measured by photoluminescence (PL) at room temperature (RT), when the growth pressure was increased. (Figure 1) The implication is a decrease of

point defects (non-radiative recombination centers) as the NH_3 partial pressure is increased. The origin of the non-radiative centers is under investigation, possibly it is related to reduced carbon incorporation as was speculated in the case of InGaIn quantum wells for blue LEDs.

While further optimization of the optical efficiency of GaN active layers is under way, we have also investigated the effect of incorporating indium (In) into GaN. A series of GaN:In test layers of 0.2 μm thick was grown on standard 1 μm GaN with indium fraction varied from 0.3 to around 4%, as determined by x-ray diffraction. The presence of Pendullung fringes in x-ray diffraction ($2\theta-\omega$) scan (top scan in Figure 3) indicates high structural quality and coherency. Furthermore, a pronounced and systematic increase of PL efficiency was observed as the In fraction is increased, at the expense of an increased PL emission wavelength. (Figure 2)

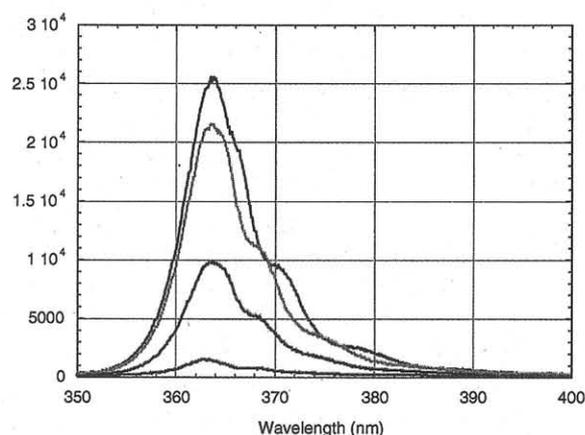


Figure 1 RT PL spectrum of 0.5 μm GaN test layers grown on standard 1 μm thick GaN. An increase of 13 times in integrated PL intensity was observed as growth pressure increases.

While a post-growth compressive stress is inevitably occurred due to the thermal mismatch between sapphire and AlGaInN, control of grown-in strain requires the matching of the lattice parameters of the active heterostructures (QW or DH-based p-n junction) to the thick buffer layer (normally GaN of a thickness of 3 to 4 μm) during growth. Two approaches have been explored: (i) The use of a thick AlGaIn buffer [5] to better match the lattice constants of the upper AlGaIn/Ga(In)N active layers, and (ii) the use of novel AlGaInN quaternary to prepare active heterostructures which will be better lattice matched to the thick GaN buffers. The employment of AlGaInN alloy renders a two-

dimensional expansion on the lattice constant-bandgap plot and facilitates the possibility of growing wider bandgap barrier layers (for GaN) while maintaining the lattice compatibility. Figure 3 show (0002) $2\theta-\omega$ radial scans of several $\text{Al}_x\text{In}_{0.037}\text{Ga}_{0.967-x}\text{N}$ epilayers. By adding Al into GaInN films, one can achieve nearly perfect lattice matching to GaN (second scan from bottom). The effect of adding In into AlGaInN and the exploration of bandgap-lattice parameter relationship associated with quaternary AlGaInN will be presented.

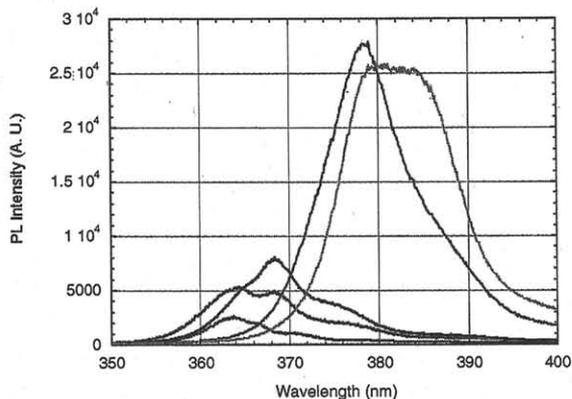


Figure 2 RT PL spectrum of 0.2 μm GaN:In test layers grown on standard 1 μm thick GaN. An increase of 11 times in integrated PL intensity was observed as the In fraction is increased to 3.7%. (PL emission is shifted accordingly from 364 to 382 nm.)

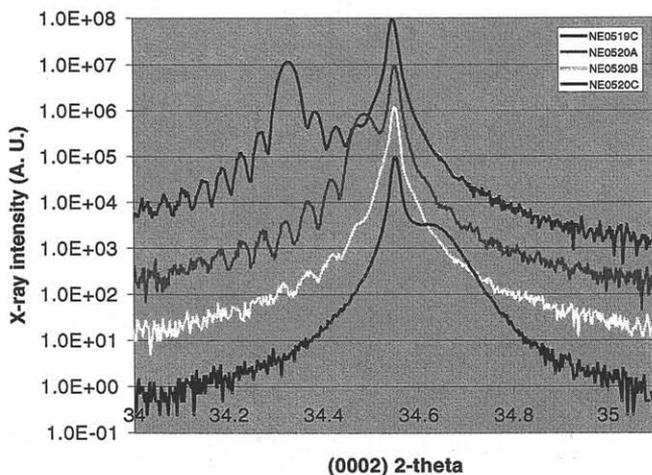


Figure 3 (0002) x-ray diffraction radial scans ($2\theta-\omega$) of AlGaInN layers with In fraction of around 3.5%. By increasing the Al fraction systematically, one can “tune” the lattice constant and achieve lattice-matching to GaN (third scan from top). Further increase of Al leads to films under tension.

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