Photoluminescence microscopy of InGaN quantum wells

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Submicron spatial resolution photoluminescence is used to assess radiative efficiency and spatial uniformity of GaN/InGaN heterojunctions. Room temperature photoluminescence of multiple InGaN quantum wells with GaN barriers fabricated by electron-cyclotron resonance assisted molecular beam epitaxy was measured as a function of position on a facet perpendicular to the layer structure. Our high resolution studies reveal that the radiative recombination for the InGaN quantum wells is 50–60 times more efficient than for the underlying GaN film. © *1997 American Institute of Physics.* [S0003-6951(97)00111-3]

Gallium nitride and their alloys have recently become a technologically exploitable material system with important optoelectronic applications. InGaN single quantum wells (SQWs) and multiple quantum wells (MQWs) are already in use as active regions for blue-green light-emitting diodes (LEDs) and blue laser diodes.[1,2] Previous reports of luminescence studies on InGaN quantum well structures have been limited to observation of quantum confinement effects on the photoluminescence (PL)[3] and cathodoluminescence (CL)[4] on InGaN heterostructures fabricated by metal-organic chemical vapor deposition (MOCVD), and PL on InGaN MQW structures fabricated by electron cyclotron resonance assisted molecular beam epitaxy (ECR-MBE).[5]

Submicron spatial resolution PL is presented as an alternative to CL which has been used to identify microscopic spatial variations in the luminescence of GaN films and crystallites.[6–8] Microscopic uv PL can be used to assess GaN heterostructures for radiative efficiency and uniformity, and we apply it to InGaN MQWs fabricated by ECR-MBE.

The sample consisted of five periods of 80-Å-thick $In_{0.09}Ga_{0.91}N$ QWs clad by 90-Å-thick GaN barrier layers, and is schematically illustrated in Fig. 1. Epitaxial growth of the MQW structure was carried out on *C*-plane sapphire substrates.[5] Elemental gallium and indium were evaporated from Knudsen effusion cells, and active nitrogen was produced by passing molecular nitrogen through an ASTeX compact electron cyclotron resonance microwave plasma source. After solvent de-greasing and thermal out-gassing, the substrates were introduced to the growth chamber where they were heated to 775 °C and exposed to a nitrogen plasma for surface nitridation. The substrate temperature was then lowered to 550 °C for the deposition of a 300-Å-thick GaN buffer. Finally, the temperature was raised to 750 °C for the growth of a 2-μm-thick GaN film. The $In_{0.09}Ga_{0.81}N$/GaNMQW structure was grown at a substrate temperature of 670 °C, and was capped with 200 Å of GaN. All of the layers were intentionally doped *n* type with silicon. The gallium beam equivalent pressure was kept constant while the nitrogen plasma power was kept at 100 W during the InGaN layer growth and was reduced to 80 W for the GaN barriers.

The PL was excited by the 335 nm line of an Ar^+ laser which was reflected into coincidence with the collection path by means of a uv long wavelength pass dichroic mirror (50% transmission point at 350 nm). The beam was focused onto the sample using a .5 numerical aperature reflecting objective creating an excitation spot of ~.6 μm full width at half maximum (FWHM) as measured with a scanning knife edge in the focal plane. The same objective collected the PL signal and the sample image was relayed to the slits of a .64 m spectrometer and dispersed with a 300 line/mm grating onto a liquid nitrogen cooled back-thinned charge-coupler device (CCD) camera. The system provided ~.6 μm spatial and 0.6 nm spectral resolution. The sample was cleaved/broken to expose a facet perpendicular to the growth direction. The InGaN heterostructure facet was scanned beneath the excitation and collection optics using a Melles Griot piezo actuated flexure stage under computer control.

Figure 2 shows the room temperature PL collected both on the edge facet (PL_{edge}) and on the epitaxial surface ($PL_\perp$, perpendicular to the growth plane). GaN PL is completely absent in the $PL_\perp$ despite the presence of a 200 Å GaN cap layer. The small GaN peak visible at 368 nm in PL_{edge} is due to edge-on excitation of the GaN film. The spectra are similar and both are completely dominated by
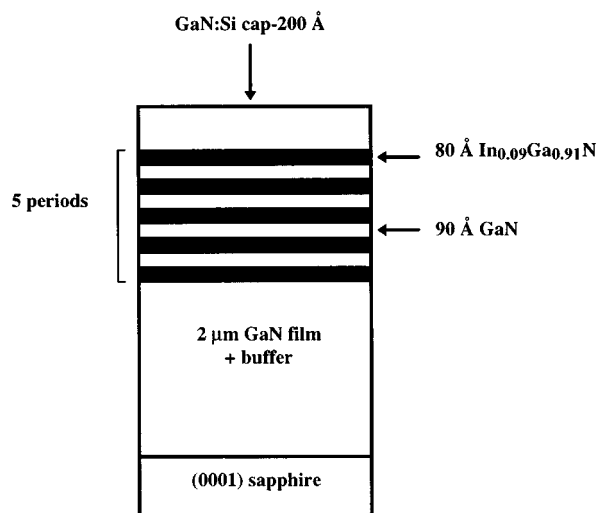

FIG. 1. Layer structure of InGaN quantum wells fabricated by electron cyclotron resonance assisted molecular beam epitaxy. Note that well and film regions are not drawn to scale.

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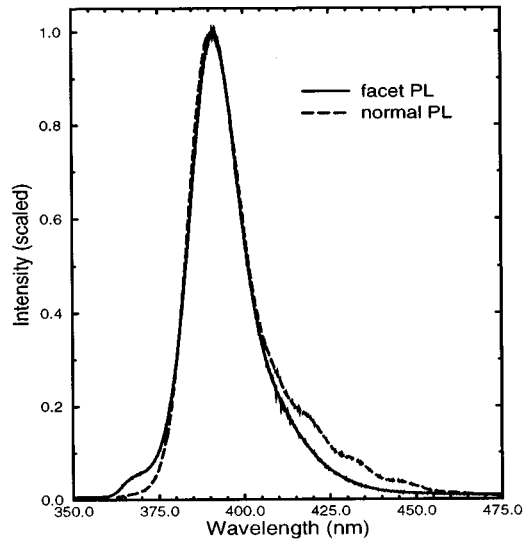[b]Address: Laser Diode Div., Polaroid Corp. (N1-1C), Norwood, MA 02062.

FIG. 2. Photoluminescence of InGaN/GaN multiple quantum well structure. The solid curve is excited and collected from the edge of the sample (PL_{edge}) and the dashed curve is excited and collected normal to the surface ($PL_\perp$). The $PL_\perp$ shows a cavity effect due to a standing wave from reflections off of the top surface and the interface between the GaN buffer and the sapphire substrate. PL was linearly dependent on pump beam power densities ranging from 20 to 1600 W/cm^2.

InGaN band-edge emission (FWHM<18 nm) indicating that the InGaN quantum wells provide a significant enhancement of the radiative recombination.

A series of PL_{edge} spectra taken on the edge facet of the layer structure as a function of position along the growth direction is displayed in Fig. 3. Each spectrum is labeled with a number which corresponds to the position indicated in the schematic of the device structure (inset, Fig. 3). The InGaN PL_{edge} intensity is greatest in spectrum 3 when the beam is centered on the quantum wells. The InGaN signal
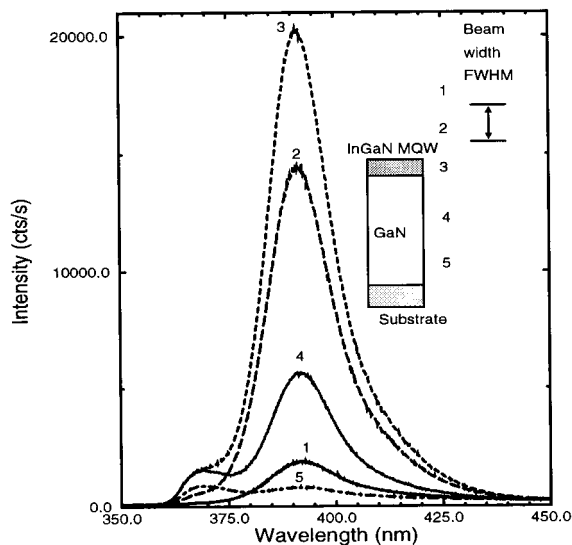

FIG. 3. Photoluminescence spectra (PL_{edge}) of InGaN/GaN multiple quantum well structure as a function of position across the layer structure. The numbers in the inset figure represent the position of the corresponding spectra.
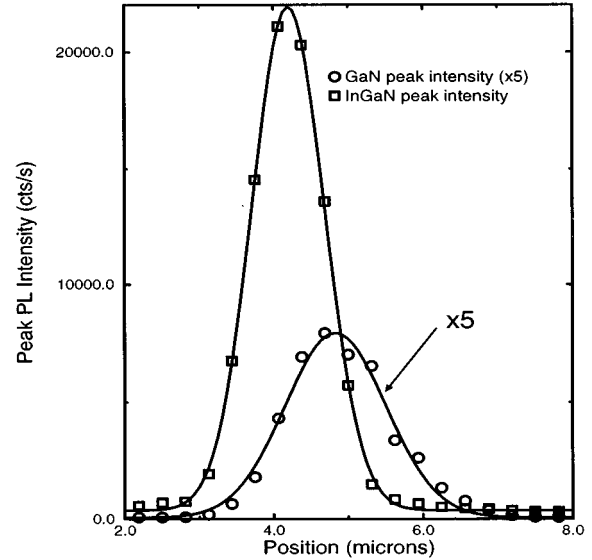

FIG. 4. Peak photoluminescence of InGaN (squares) and GaN (circles) as a function of position across the layer structure. Gaussian fits are a guide to the eye.

peaks at a value 20 times larger than the GaN PL_{edge} signal. The GaN PL_{edge} signal (FWHM=5 nm) has reached half of its peak value at the position corresponding to the InGaN PL_{edge} peak. This is consistent with half of the pump beam exciting the GaN film while the narrow InGaN region is at the peak of the pump beam cross section. Spectrum 4 is closer to the substrate and shows that the InGaN signal diminishes and the GaN PL_{edge} increases as the GaN film on the substrate side of the QWs is centered under the pump beam.

Figure 4 shows the peak amplitudes of the InGaN and GaN PL_{edge} plotted as a function of position along the growth direction as determined by fitting each spectrum with two Gaussians. The FWHM of the spatial dependence of the InGaN MQW PL_{edge} intensity is ~1.2 μm, slightly larger than the laser beam spot size measured by a scanning knife-edge. Multiple PL_{edge} line scans were performed as a function of vertical position along the optical axis, insuring that the sample plane was located in the pump beam waist minimum. Finite penetration of the pump beam into the sample will result in a measured FWHM of the spatial dependence of the InGaN PL_{edge} which is slightly larger than the pump beam. For a penetration depth of a few hundred nanometers, one would expect a measured FWHM for the InGaN PL_{edge} of less than a micron. The InGaN PL_{edge} FWHM can then be attributed to the pump beam width and possibly a small contribution from diffusion of photoexcited carriers from the GaN film. We set an upper limit of 400 nm for the diffusion length of minority carriers based on the spatial extent of the InGaN MQW PL_{edge}.

The measured PL_{edge} from the InGaN MQWs is 15 times stronger at peak than from the 2-μm-thick GaN film beneath. Assuming that all of the electron-hole pairs generated in the barrier diffuse to the QWs, accounting for the difference in cross-sectional areas of the materials under illumination, and correcting for the measured relative collection and detection efficiency, we estimate that the InGaN MQWs are approxi-

mately 50–60 times more efficient for radiative recombination than the GaN film. We attribute the increase in radiative efficiency of the InGaN MQWs versus the GaN film as being a result of the confinement of electron-hole pairs in the InGaN quantum wells.

In conclusion, we have performed photoluminescence microscopy in the ultraviolet spectrum on InGaN MQWs. We are able to spatially resolve luminescence from the GaN film and the MQW structure. We observe a very strong enhancement of the radiative recombination for the MQWs relative to the underlying GaN film. Finally, we anticipate the further application of this technique in the analysis of GaN alloy heterostructures.

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