Growth and photoluminescence studies of *a*-plane $AlN/Al_xGa_{1-x}N$ quantum wells

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Both *a*-plane and *c*-plane $AlN/Al_{0.65}Ga_{0.35}N$ quantum wells (QWs) have been grown by metal organic chemical vapor deposition and their photoluminescence (PL) emission properties were studied and compared. It was found that the low temperature PL characteristics of *a*-plane QWs are primarily governed by the quantum size effect, whereas those of *c*-plane QWs are significantly affected by the polarization fields. The PL decay time was found to be only weakly dependent on the well width L_w for *a*-plane QWs, whereas a strong dependence of the PL decay time on L_w was observed for *c*-plane QWs. Moreover, L_w dependence studies also revealed that structures with $L_w > 2$ nm and $L_w \approx 2$ nm provide highest emission efficiency in *a*-plane and *c*-plane $AlN/Al_{0.65}Ga_{0.35}N$ QWs, respectively. © *2007 American Institute of Physics*.
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Deep ultraviolet (UV) emitters and detectors operating in the 200–340 nm wavelength range are important devices for many applications, including water purification, biological and chemical agent detection, and medical research/health care.[1] Al-rich AlGaN alloys have a tunable emission wavelength down to 200 nm, which makes them very useful for these applications. As demonstrated by light emitting diodes, laser diodes, and electronic devices, many III-nitride based devices must take advantages of quantum well (QW) structures in order to achieve optimal device performance. To realize deep UV emission ($\lambda < 280$ nm), Al-rich AlGaN based QWs are required. Recently, several groups have been studying Al-rich AlGaN based emitters to obtain UV emission wavelength below 300 nm.[2–7] The growth of the QW structures on the basal plane of either sapphire or SiC resulted in the deposition of wurtzite material with (0001)-plane (*c*-plane) orientation.[8] Interfacial polarization discontinuities within heterostructures are associated with fixed sheet charges which produce strong internal electric fields. These "built-in" polarization-induced electric fields significantly reduce the electron-hole wave function overlap and radiative efficiency and hence limit the performance of optoelectronic devices that employ nitride QWs as active region. Our previous studies on AlN/AlGaN QWs grown on *c*-plane sapphire substrate showed that these AlN/AlGaN QW structures possess a built-in polarization field of ~4 MV/cm.[9]

Nitride crystal growth along nonpolar directions provides an effective mean for producing nitride-based quantum structures that exhibit reduced effects by the strong polarization-induced electric fields since the polar axis lies within the growth plane of the film. The growth of *a*-plane GaN/AlGaN QW structures on *r*-plane sapphire substrates has been previously demonstrated.[10,11] Optical characterization of these structures has shown that the effects due to the polarization-induced electric fields are greatly reduced in nonpolar QWs.

In this letter we report on the growth and photoluminescence (PL) properties of *a*-plane $AlN/Al_{0.65}Ga_{0.35}N$ QWs grown on *r*-plane sapphire substrates by metal organic chemical vapor deposition (MOCVD). We also investigated a range of well widths for *a*-plane and *c*-plane QWs, which allowed us to reveal the emission characteristics that are unique to QWs with nonpolar orientation. Deep UV PL emission spectroscopy was employed to probe the L_w dependence of the optical properties. Our results showed that in *a*-plane QWs, high emission efficiency can be attained at $L_w > 2$ nm. In contrast, there exists an optimal choice of L_w (=2 nm) for obtaining high emission efficiency in *c*-plane QWs. The PL decay time of *a*-plane QWs exhibits only a slight increase with increasing L_w, while that of *c*-plane QWs has a strong dependence on L_w.

$AlN/Al_xGa_{1-x}N$ ($x \sim 0.65$) QWs were grown on *r*-plane and *c*-plane sapphires by MOCVD. The growth temperature and pressure were 1120 °C and 50 Torr, respectively. Prior to the growth of $AlN/Al_{0.65}Ga_{0.35}N$ QW, a thin AlN buffer layer and a 1 μm undoped AlN epilayer were first grown on sapphire. This was then followed by the growth of $Al_{0.65}Ga_{0.35}N$ QW and a 10 nm AlN barrier. The barrier and well widths were determined by the growth rates of the AlN and $Al_xGa_{1-x}N$ epilayers. The deep UV PL spectroscopy system consists of a frequency quadrupled 100 fs Ti:sapphire laser with an average power of 3 mW and a repetition rate of 76 MHz at 196 nm, a 1.3 m monochromator with a detection capability ranging from 185 to 800 nm, and a streak camera detector with 2 ps time resolution.[12]

The (10 K) PL spectra of the *c*- and *a*-plane $AlN/Al_{0.65}Ga_{0.35}N$ QWs are shown in Figs. 1(a) and 1(b), respectively. Independent of the crystal orientation, the QW PL emission peak shifts to lower energies with increasing L_w due to the weakening of the quantum confinement effect. In particular, the emission peak energies of the *a*-plane QWs steadily approach but do not redshift beyond the band edge transition of the $Al_{0.65}Ga_{0.35}N$ epilayer with increasing L_w. Conversely, the *c*-plane QW emission peak energy redshifts with L_w and becomes even lower than the band edge transition peak of the $Al_{0.65}Ga_{0.35}N$ epilayer at $L_w > 2$ nm. This strong dependence of the PL emission energy on L_w in *c*-plane QWs is due to the strong spontaneous and strain-induced piezoelectric fields of ~4 MV/cm,[9] in addition to

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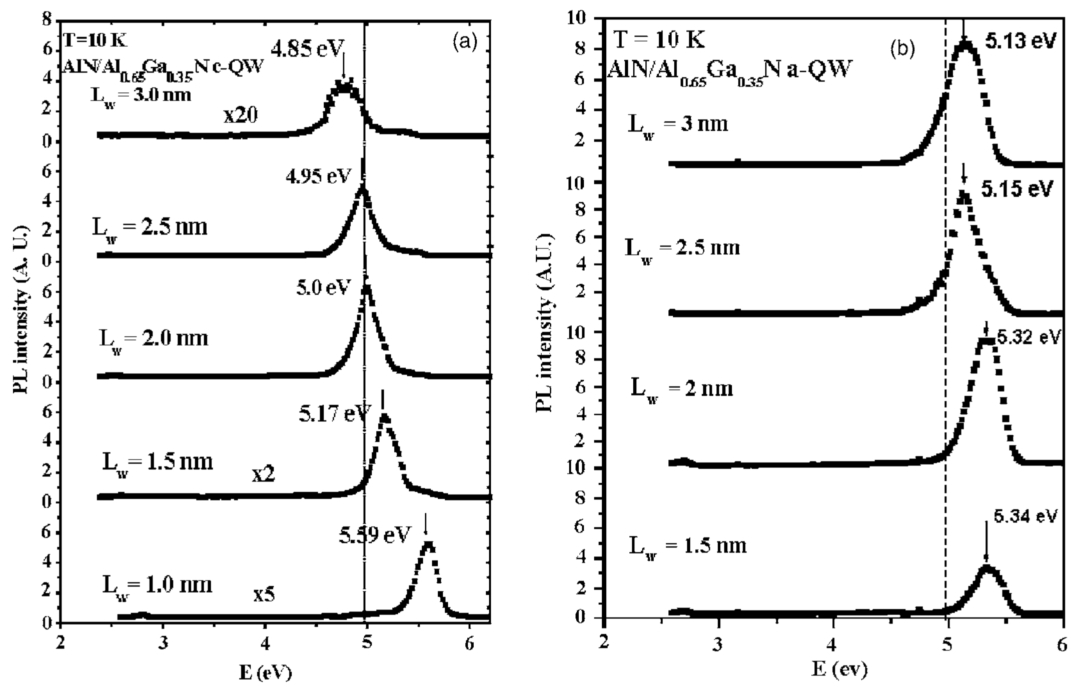

FIG. 1. Low temperature (10 K) PL spectra of (a) c-plane $AlN/Al_{0.65}Ga_{0.35}N$ QWs with well width L_w, varying from 1 to 3 nm and (b) a-plane $AlN/Al_{0.65}Ga_{0.35}N$ QWs with well width L_w, varying from 1.5 to 3 nm. All samples have a fixed barrier width of 10 nm. The vertical dashed lines represent the emission peak position of $Al_{0.65}Ga_{0.35}N$ epilayers.

the quantum size effect. These polarization fields are much weaker in a-plane QWs.[13]

The normalized PL emission intensity of the a- and c-plane QWs as a function of L_w is plotted in Fig. 2 (each orientation's intensity was normalized separately). We observed that high emission intensity can be obtained in a-plane QWs with $L_w > 2$ nm and in c-plane QWs with $L_w = 2$ nm. In c-plane QWs, the balance between the reduced radiative recombination efficiency in thick wells due to the polarization fields and in thin wells due to enhanced carrier leakage to the barrier region[14] results in an optimal choice of L_w for obtaining high emission efficiency. In contrast, since the polarization fields are much weaker in nonpolar QWs, high emission efficiency is obtained in QWs with $L_w > 2$ nm. This suggests that the a-plane AlN/AlGaN QW system potentially provides a much greater flexibility for the device structural design than its c-plane counterparts.

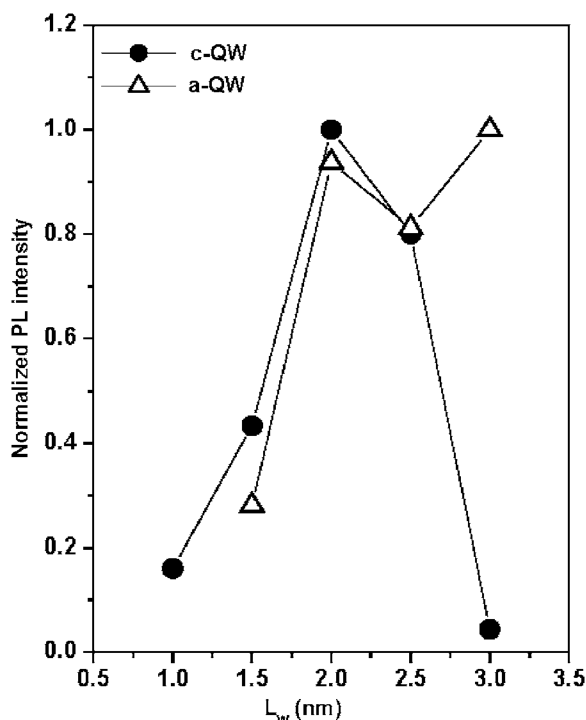

FIG. 2. Normalized low temperature PL intensity plotted as a function of well width L_w for both a- and c-plane $AlN/Al_{0.65}Ga_{0.35}N$ QWs.

To further investigate the differences between polar and nonpolar QWs, the PL decay characteristics of a- and c-plane QWs for two representative well widths ($L_w = 1.5$ and 3 nm) were measured and the results are shown in Fig. 3. The PL decay transients show nonexponential decay with a slower component at longer decay times. For c-plane QWs the PL decay time strongly depends on L_w. This can be explained by the presence of the strong polarization fields (~ 4 MV/cm) in the polar QWs. The electrostatic fields spatially separate the electron and hole wave functions, thereby reducing the oscillator strength for their radiative recombination. Conversely, for a-plane QWs the PL decay time exhibits only a weak dependence on L_w.

Figure 4 shows the Arrhenius plots of the PL intensity for (a) a-plane and (b) c-plane QWs with $L_w = 1.5$ nm. The solid lines in both plots are the least squares fit of the measured data to the following equation:

$$I_{emi}(T) = I_0[1 + Ce^{(-E_0/KT)}]^{-1}, \quad (1)$$

where $I_{emi}(T)$ and I_0 are, respectively, the PL intensities at finite temperature T and 0 K, while E_0 is the activation energy. At $T > 140$ K, the PL intensity is thermally activated with an activation energy of about 125 meV for a-plane QW and 110 meV for c-plane QW. Since the observed activation energies are much smaller than the band offsets of the band gap difference between the wells and barriers, the thermal quenching in these samples is not due to the thermal activation of electrons from QWs to barriers. The quenching of the

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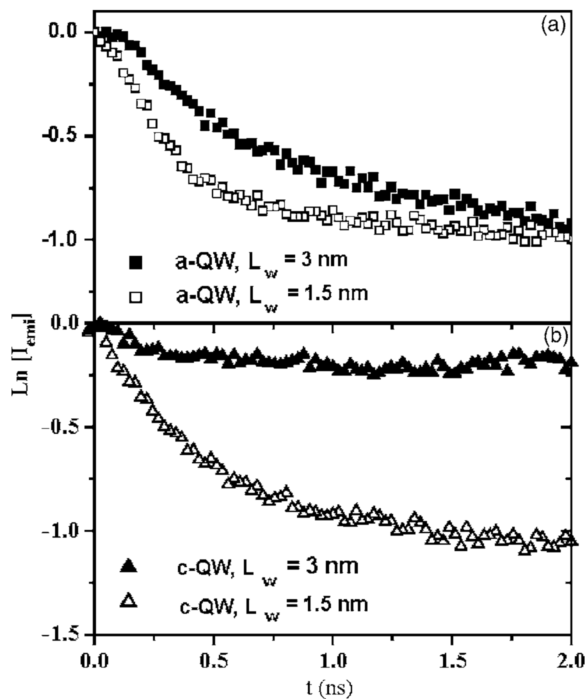

FIG. 3. Low temperature (10 K) PL decay transients of two representative $AlN/Al_{0.65}Ga_{0.35}N$ (a) *a*-plane QWs and (b) *c*-plane QWs with L_w=1.5 and 3 nm.

luminescence with temperature can be explained by thermal emission of the carriers out of a confining potential (exciton localization potential), allowing their scattering with dislocations. An activation energy of around 120 meV has been previously measured for the PL intensity in $Al_{0.70}Ga_{0.3}N$ epilayers grown on *c*-plane sapphire,[15] which is close to the activation energies obtained in $Al_{0.65}Ga_{0.35}N$ QWs of the present study. The higher activation energy in *a*-plane QWs (125 meV) than in *c*-plane QWs (110 meV) may be attributed to an increased interface or alloy fluctuation, which is also reflected in a slight increase in the PL emission linewidth in *a*-plane QWs (Fig. 1). Furthermore, as pointed out previously,[16] "lateral" polarization fields may still present in *a*-plane QWs or quantum dots because the *c* axis lies in the *a* plane so that the well width fluctuations could induce polarization fields in the *a* plane, resulting in an inhomogeneous spectral broadening and an apparent enhanced localization energy. The larger drop in PL intensity in *a*-plane QWs than in *c*-plane QWs in the temperature region of 10–300 K may be related to the growth anisotropy along nonpolar directions so that that *a*-plane QWs possess more dislocations than *c*-plane QWs, as suggested by x-ray diffraction and atomic force microscopy measurements.

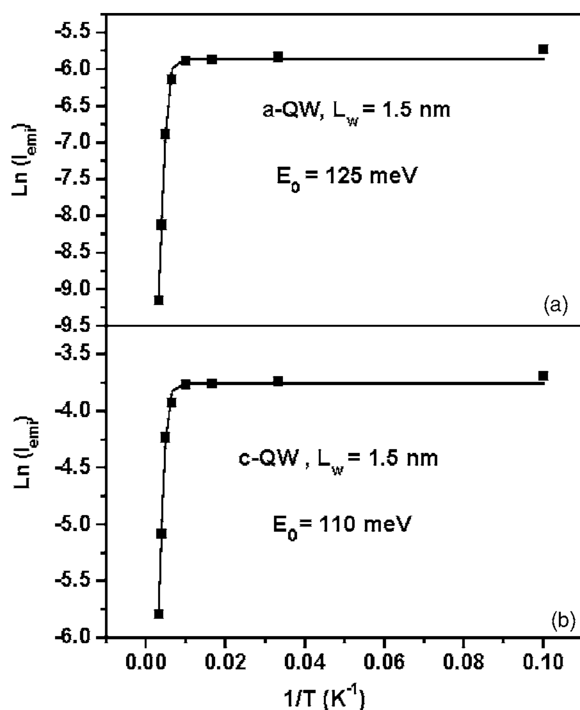

FIG. 4. Arrhenius plots of the PL intensity of (a) *a*-plane (b) *c*-plane QWs with L_w=1.5 nm. The solid lines in both plots are the least squares fit of the measured data to Eq. (1).

In summary, *a*- and *c*-plane $AlN/Al_{0.65}Ga_{0.35}N$ QWs have been grown by MOCVD and their PL emission characteristics were measured and compared. It was found that the low temperature PL characteristics of *a*-plane QWs are primarily governed by the quantum size effect, as in traditional III-V semiconductor QWs. In contrast, the emission characteristics of *c*-plane QWs are affected by strong polarization fields in addition to the quantum confinement effect. The PL decay time was found to be weakly dependent on the well width L_w for *a*-plane QWs. However, a strong dependence of the PL decay time on L_w was observed for *c*-plane QWs, which is caused by the variation of the polarization fields in QWs due to varying L_w. It was shown that to obtain highest quantum efficiency, *a*- and *c*-plane AlN/AlGaN QWs with high Al contents should be designed to have $L_w > 2$ nm and $L_w \approx 2$ nm, respectively.

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