Deep Ultraviolet Light Emitting Diodes Based on Short Period Superlattices of AlN/AlGa(In)N

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We report a systematic study of the optical properties of superlattices of AlN/AlGa(In)N. The superlattices were grown on sapphire substrates using gas source molecular beam epitaxy with ammonia. Effective bandgaps between 4.5 eV (276 nm) and 5.3 eV (234 nm), as determined by optical reflectivity measurements, were 1.25–2.25 nm. The superlattices were grown reproducibly and with excellent structural and optical properties. When doped with Mg, room temperature grown reproducibly and with excellent structural and optical properties. SPSLs containing as many as 400 well-barrier pairs can be demonstrated an LED emitting at 262 nm. All samples were grown by gas source molecular beam epitaxy with ammonia on c-plane sapphire. Si, derived from silane, and Mg were used for n-type and p-type doping, respectively. Details of epitaxial growth and mesa-LED fabrication have been described elsewhere. The effective bandgaps of AlN/AlGa(In)N SLs are known to depend on the well/barrier thickness ratio, the well composition, and the SL period. Our SPSLS containing as many as 400 well-barrier pairs can be grown reproducibly and with excellent structural and optical properties. When doped with Mg, room temperature growth and mesa-LED fabrication have been described elsewhere. The effective bandgaps of AlN/AlGa(In)N SLs are known to depend on the well/barrier thickness ratio, the well composition, and the SL period. Our SPSLs containing as many as 400 well-barrier pairs can be grown reproducibly and with excellent structural and optical properties.

In this paper we systematically explore the effects of SPSL parameters on optical properties in the short wavelength limit. Based on the results of these investigations, we demonstrate an LED emitting at 262 nm. All samples were grown by gas source molecular beam epitaxy with ammonia on c-plane sapphire. Si, derived from silane, and Mg were used for n-type and p-type doping, respectively. Details of epitaxial growth and mesa-LED fabrication have been described elsewhere. The effective bandgaps of AlN/AlGa(In)N SPSLS were obtained from optical reflectance. Because of the high growth temperature, the InN content in our wells is quite low. Secondary ion mass spectrometry (SIMS) measurements show approximately 10¹⁷ cm⁻³ in atoms in the SL structure. The amount added is small enough not to reduce the bandgap or to alter the period of the superlattice but it results in improved luminescence efficiency. It has been argued recently that even small amounts of In in the lattice may have important effect on the interfacial electric field and therefore electrical properties of the SL.

UV devices were designed for emission through the transparent sapphire substrate. The cross-section of basic ~260 nm double heterostructure (DH) LED and its energy bandgap profile are shown in Fig. 1. Note an each SPSL functions as an artificial crystal having a well-defined effective bandgap. Thus a sandwich of such SPSLs looks similar to a DH composed of bulk semiconductors having different bandgaps. The carrier’s injection and optical confinement properties of SPSLs-based DHS are under active investigation. The LED consists of four main parts: (1) AlN or Al₀.₉₂Ga₀.₀₈N buffer layer; (2) n-type SPSL cladding layer of ~ 400 nm; (3) undoped SPSL active region of ~ 30 nm; (4) p-type SPSL cladding layer of ~ 210 nm. The buffer layer was incorporated in order to reduce dislocation density in the device SPSL. Dislocation density in the top part of the buffer layer was estimated from TEM measurements at ~ (6–8) × 10⁹ cm⁻². The cladding and active layers were composed of AlN/Al₀.₉₂Ga₀.₀₈N SPSLs having different effective bandgaps. The electron concentration in the n-type SPSL was ~ 10¹⁹ cm⁻³ with the resistivity ~ 0.04 Ω·cm, measured on a control SPSL grown on sapphire. The p-type Mg-doped SPSL had a total thickness of ~ 210 nm and the same composition as the n-type SPSL. The hole concentration of ~ 10¹⁹ cm⁻³ and resistivity of ~ 4 Ω·cm were obtained in Hall measurements on SPSL test structures. A 30-nm thick active region was introduced between the n- and p-type cladding layers to provide carrier confinement. Its bandgap was ~ 300 meV smaller than that of the cladding layers. The active region

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Fig. 1. Schematic cross-section and effective bandgap distribution across DHS LED. Buffer layer is composed of AlN or Al₀.₉₂Ga₀.₀₈N. The Ecl and Eac are the effective bandgaps of n- (p-) type cladding layer and active layer, respectively.
growth time. Detailed XRD simulations, to be published.

Vegard’s Law. The SL period was determined by XRD from 1.25 to 2.25 nm. The average composition of the SPSL was found to lie between the trends shown for 2 ML (0.50 nm) and 3 ML (0.75 nm). We describe later how the intermediate well thickness was used for growing a LED.

The effective bandgap of the SPSL can thus be adjusted either by changing the well or barrier thickness. We observe two trends from the results in Fig. 2, which originate from varying the average AlN composition in the SPSL. First, the 0.50 nm thinner well series possess energy gaps 400 ± 30 meV higher than the 0.75 nm well set for the same period. For example, points a and b in Fig. 2 illustrate the effect of increasing the well thickness by one ML and leaving the barrier thickness constant while maintaining constant well thickness. A linear fit provides a shift ~100 ± 20 meV/ML (~7 nm/ML) for each well thickness series. We illustrate this in Fig. 2 through points b and c. We utilize here a simple linear approximation for guiding our device design. A more complete data set, with additional well thickness, is required to develop a better picture of the dependence of energy gap on SPSL parameters. It is clear from our results that control of the SPSL well and barrier thickness provides “coarse” and “fine” adjustment of the effective bandgap. Changing the well thickness with fixed period provides the coarse control of ~400 ± 30 meV. Keeping the well thickness constant and growing with different barrier widths provides the fine control of 100 ± 20 meV/ML. The two approaches to changing the energy gap may also be used in combination. For example, increasing the well by 1 ML and leaving the barrier thickness constant produces an intermediate change in bandgap energy of ~270 ± 20 meV, such as points a and c in Fig. 2. We design our LED cladding and active layers based on these results. The growth procedure of adding one ML to the well is straightforward, and avoids reducing the barrier to two MLs in thickness.

Mesa-etched LEDs were fabricated as previously described. The current-voltage (I-V) characteristic of a 160 μm mesa LED is plotted in Fig. 3. LED light emission is observed with forward dc current above 2 mA. The device turns on at ~6.0 V and has differential series resistance $R_{ms}$ of 110–120 Ω under forward bias from 8 V to 12 V. The $R_{ms}$ of mesa diodes is the sum of the contact ($R_c$), spreading ($R_s$), and vertical ($R_v$) resistances. The resistance of the etched
part of the mesa, corresponding to transport across the SPSL layers is \( R_s \approx \rho_{\perp}(h/A) \), where \( \rho_{\perp} \) is the perpendicular resistivity of the SPSL, \( h \) is the height of the mesa (\( \sim 300 \) nm, with the thickness of p-SL \( \sim 200 \) nm), and \( A \) is the contact area. The contact resistance of a 160-\( \mu \)m diameter diode, obtained from specific contact resistance, is \( R_c \approx 90 \) \( \Omega \). The estimated spreading resistance of this LED is \( R_s \approx 20 \) \( \Omega \) and comes from n-type buffer layer. Finally we obtain \( R_s \approx 5 \) \( \Omega \) resulting in \( \rho_{\perp} \approx 50 \) \( \Omega \)-cm for a p-type SPSL. Comparing this to the in-plane conductivity (\( \rho_{\parallel} \)) of the p-type SPSL obtained from Hall measurements, \( \rho_{\parallel} \approx 4 \) \( \Omega \)-cm, we obtain the conductivity anisotropy \( \rho_{\perp}/\rho_{\parallel} \approx 12 \). These simple considerations indicate relatively low \( \rho_{\perp} \), considering the high AlN fraction in our SPSLs, and underscores the importance of reducing the contact resistance to p-type materials. It is also important to lower \( R_c \) by optimization of the buffer layer thickness and its resistivity.

Based on the results of Fig. 2, we produced two LEDs operating at different wavelength. The first uses cladding layers described by point \( a \) in Fig. 2. In the active region the well width is increased by 1 ML and the barrier thickness is unchanged, corresponding to point \( c \) in Fig. 2. The EL spectrum of this LED is shown in Fig. 4. For the 4 mA dc drive current shown, two peaks are observed at wavelengths of 262 and 320 nm. We focus here on the short wavelength peak at 262 nm and will return to the 320 nm feature later. We include the peak at 262 nm (labeled "X") in Fig. 2. Agreement is excellent between the measured emission wavelength and what was expected based on the 1 ML (\( a \rightarrow c \)) increase in well thickness. A second LED uses cladding layers with properties illustrated by \( \alpha \) in Fig. 2. The average well thickness of the n- and p-type cladding materials, as discussed earlier, is intermediate to 2 and 3 ML. For the active layer we increase the well thickness by 1 ML and leave the barrier thickness intact. This is illustrated by points \( \alpha \) and \( \gamma \) in Fig. 2, assuming here that the energy differences \( \alpha \rightarrow \beta \rightarrow \gamma \) in this analysis are the same as what we used for analogous points \( a \), \( b \), and \( c \) with integer well thickness. The EL emission spectrum for this LED is also shown in Fig. 4. At the same forward dc current of 4 mA, two peaks at 282 nm and 320 nm are observed. The 282 nm peak is included in Fig. 2 as symbol \( Y \). We see from this that the design approach accurately predicts the short wavelength LED emission energy.

Now we turn our attention to the broad 320 nm peak seen in both EL spectra in Fig. 4. At low forward current its intensity is almost equal to that of the bandedge emission. However, the bandedge emission grows faster with current, exceeding the intensity of the 320 nm peak by factors of 2 and 3 at dc currents of 10 mA\(^{10}\) and 12 mA,\(^{6}\) respectively. A similar feature has been observed in 285 nm LEDs\(^{10}\) and 292 nm LEDs\(^{9}\) based on random alloy layers and attributed to a free-carrier to deep acceptor transition in Mg-doped p-Al\(_0.5\)Ga\(_0.5\)N. Such a transition could arise in our SPSL-based LED because of the formation of 1 ML thick barrier/well interfaces in which the AlN concentration may range from 30% to 90% of AlN, as mentioned above. Even though the interface regions are only 1 ML thick, the total volume fraction of these regions throughout the SPSL can be considerable since the period is so small. We note however that the position and width of the 320 nm peak in Fig. 4 do not scale with the effective bandgap of SPSL, making it difficult to assign our feature to a free-carrier to deep acceptor transition. Our CL studies, reported elsewhere,\(^{10}\) show that the intensity of the long wavelength peak scales with Si doping in the SPSLs. Similar sub-bandgap features are observed in 290 nm LEDs grown on silicon substrates.\(^{11}\) Clearly, additional experiments are required in order to better identify the origin of the 320 nm peak present in UV LEDs.

To summarize, we use the SPSL approach to engineer bandgaps as large as 5.3 eV and demonstrate DHS LEDs with emission wavelength as short as 262 nm. We systematically vary the properties of AlN/Al\(_{0.2-\alpha}\)Ga\(_{0.8+\alpha}\)(In)N SPSLs to show that absorption edge and emission wavelength are controllable by varying the well and barrier properties. Optical bandgaps could be adjusted reproducibly between \( \sim 4.5 \) and 5.3 eV by monolayer variations in the well and barrier thickness. Thus the DHS with bandgap offset of \( \sim 0.8 \) eV can be obtained. Emission is observed from the active region in excellent agreement with the design optical bandgap. With cladding layers composed of 0.50 nm wells and having \(~2\) nm period, SPSL-based “homojunction” LEDs with deep UV emission near 240 nm appear feasible.

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**Fig. 3.** Room-temperature \( I-V \) of a 160\( \mu \)m mesa LED.

**Fig. 4.** EL spectra of 262 nm and 280 nm LEDs operating at low dc current of \(~4\) mA.
Here “homojunction” means that the effective bandgaps of all SPSLs are the same.

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