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Revisiting the “In-clustering” question in InGaN through the use of aberration-corrected electron microscopy below the knock-on threshold

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The high intensity of light emitted in InxGa1−xN/GaN heterostructures has been generally attributed to the formation of indium-rich clusters in InxGa1−xN quantum wells (QWs). However, there is significant disagreement about the existence of such clusters in as-grown InxGa1−xN QWs. We employ atomically resolved CSTEM-corrected scanning transmission electron microscopy and electron energy loss spectroscopy at 120 kV—which we demonstrate to be below the knock-on displacement threshold—and show that indium clustering is not present in as-grown In0.22Ga0.78N QWs. This artifact-free, atomically resolved method can be employed for investigating compositional variations in other InxGa1−xN/GaN heterostructures. © 2013 AIP Publishing LLC. [http://dx.doi.org/10.1063/1.4807122]

Due to its unique optoelectronic properties, the semiconductor alloy InGaN finds increasing use in a wide range of applications, from light emitting diodes (LEDs),1 to laser diodes,2 photovoltaic solar cells,3 and solid-state bio-chemical sensors.4 Since the discovery of blue LEDs5—which have made phosphor-free white solid state lighting (SSL) a possibility6—in InxGa1−xN (0 ≤ x ≤ 1) LED heterostructures composed of either single or multiple InxGa1−xN quantum wells (QWs) sandwiched between thin GaN layers have sparked tremendous interest. In particular, wavelength-tunable InxGa1−xN LEDs have been projected to increase the efficiency of SSL technology and, therefore, make a significant impact on energy consumption.7,8 However, despite the commercial success in recent years, InGaN/GaN LEDs have not reached their projected potential because their experimental efficiencies remain below theoretical predictions.9 In order to lower the cost and energy consumption, development of LEDs with efficiencies above 70% is required.10 Therefore, it is imperative to address unresolved fundamental scientific questions pertaining to the mechanisms and efficiency of the light emission in these heterostructures.

InGaN/GaN QWs are typically grown on a GaN buffer layer, which is grown on top of a sapphire or SiC substrate; because of the large lattice mismatch between the buffer layer and the substrate, InGaN layers contain a threading dislocation density (TDD) of 10⁶–10⁹ cm⁻² or higher.11 In spite of such high TDDs, InGaN based LEDs emit intense light, whereas presence of even a fraction of this TDD in other semiconductor LEDs completely quenches the light emission. It has been proposed that because InGaN is an alloy with a large miscibility gap, it can undergo spinodal decomposition resulting in the formation of In-rich clusters within the QWs.10,11 Whereas small-scale statistical fluctuations are to be expected in any random ternary alloy, the In-clusters in InGaN were proposed to be large-scale (of 1–3 nm in size) and with significant compositional deviation from the nominal QW composition.12 These In-rich clusters have been postulated to be critical to the observed high emission efficiency of InGaN, because they could act as sinks for electron-hole pairs where they recombine to emit light away from any light-quenching centers such as threading dislocations. However, the existence of the In-rich clusters is a topic of considerable debate. High-resolution transmission electron microscopy (HRTEM)10 and scanning transmission electron microscopy (STEM) studies combined with low-loss electron energy loss spectroscopy (EELS)12 at accelerating voltages of 800 kV and 200 kV, respectively, suggested the existence of large-scale In-rich clusters adjacent to In-depleted regions in InGaN QWs. However, subsequent studies13,14 showed that no large-scale compositional fluctuations existed, beyond the small fluctuations expected in a random alloy such as InGaN. These authors suggested that exposure to a 200 kV or higher energy electron beam can cause irreversible irradiation damage in InGaN that could be misconstrued as spinodal decomposition. Thus, it was postulated that the reported In clustering was an artifact of TEM imaging and their existence in as-grown InGaN layers has since been questioned.

Although attempts have been made to minimize beam damage by investigating InGaN under low illumination current densities of 15–20 A cm⁻²,15 these studies have been conducted at accelerating voltages of ≥200 kV, which is presumably above the knock-on displacement damage threshold in these materials.16 Complementary techniques, such as atom probe tomography (APT), have also been employed to address the question of In clustering,14,17 but the inability of APT to attain atomic resolution in the lateral direction and lack of proper statistical sampling have rendered these studies inconclusive.18 In nearly the entire body of work...
pertaining to In clustering, TEM investigations were done by employing conventional (non-aberration corrected) TEMs. For these instruments, it is not possible to obtain atomic resolution at lower accelerating voltages and thus it has been difficult to conclusively rule out knock-on displacement damage as the origin of the observed large-scale In clusters. This debate on the existence of In clusters is further complicated by the fact that studies by different groups used samples grown under different conditions, and is known that variations in growth conditions and composition, \( x \), of the In\(_{x}\)Ga\(_{1-x}\)N alloy can influence its structural, electronic, and optical properties.\(^{19,20}\) In this work, we employ atomically resolved aberration (Cs-)corrected STEM operated at 80 kV and 120 kV for structural and electronic characterization of InGaN QWs samples and demonstrate that these operational conditions are below the knock-on displacement threshold. Critically, the samples we are investigating here are from the same region of the same InGaN QWs wafer used in Ref. 13, the study that first questioned the existence of In-rich clusters.

The InGaN/GaN samples were grown by metalorganic vapor phase epitaxy and were composed of 10 In\(_{0.22}\)Ga\(_{0.78}\)N 2.8 nm-thick QWs separated by 7.2 nm-thick GaN barriers. The thickness of the QWs and their nominal composition were calculated by x-ray reflectivity and x-ray diffraction.\(^{13}\) Room temperature cathodoluminescence of the wafer showed a high intensity peak centered at 508 nm consistent with the nominal In composition of 22%.\(^{21,22}\) We have independently verified the nominal composition by APT measurements (not described herein). Electron transparent TEM films were prepared from the parent wafer using a FEI Helios 600 dual beam focused ion beam (FIB) operated at 30 kV. A final cleaning step using a FIB voltage of 5 kV was also performed. The amorphized surface layer—an inevitable consequence of gallium ion (Ga\(^{2+}\)) implantation during FIB milling—was removed by a wet-chemical etch at room temperature using 14% weight/volume solution of potassium hydroxide (KOH) for 3–5 min. Finally, to remove any residue of the wet-etch, samples were cleaned in a low power 75% argon and 25% oxygen plasma for 20 min. HRTEM imaging of our samples was carried out in an image-corrected FEI Titan 80–300 TEM operated at 80 kV, while STEM and EELS studies were performed using a probe-corrected Hitachi HD-2700C operated at 120 kV.

The resulting TEM samples were <20 nm thick, as measured by EELS, and have minimal surface damage (Figure 1). If present, contrast in In-rich regions in InGaN cannot be observed using high-angle annular dark-field (HAADF) STEM due to the significant difference in the atomic numbers of In (\( Z = 49 \)) and Ga (\( Z = 31 \)).\(^{23}\) Additionally, Cs-correction of the electron probe makes atomically resolved STEM imaging possible, even at accelerating voltages lower than 200 kV. To confirm that the use of low-voltage STEM mitigates the effects of beam damage, we obtained STEM data as a function of the exposure time using an electron beam with the spot size of \( \sim 1 \) Å and a typical current density of \(<0.9\) A cm\(^{-2}\). These parameters are an order of magnitude smaller compared to previously reported HAADF-STEM and EELS\(^{10}\) (the beam spot size of 0.14 nm and \( \sim 1 \) nm, respectively) or HRTEM\(^{13}\) (current density of 35 A cm\(^{-2}\)) studies. Recently, a Cs-TEM study\(^{24}\) and a C\(_s\)-STEM study\(^{25}\) were performed using a dose on the same order as in our experiments; however, these studies were conducted at 300 kV and 200 kV, respectively, which are above the knock-on threshold for the InGaN alloy. Figure 2(a) presents a Cs-STEM HAADF micrograph acquired at 120 kV shortly after aligning the electron microscope (\( t = 0 \) min) in a different sample location. It can be seen that there are no large-scale contrast variations—as would be expected for large-scale In-rich clusters—in the InGaN QWs. By comparing Figure 2(a) with an image acquired after 16 min of continuous exposure (Figure 2(b)), no change in contrast with exposure time was observed. Together with a higher magnification, lattice resolved STEM image of one of the QWs (Figure 2(c)), these results confirm that there are no large-scale compositional fluctuations inside the QWs or significant electron beam damage after exposure of more than 16 min.

We note that the small variations in contrast apparent in Figure 2 can be attributed to residual surface amorphization that was not removed by the KOH wet-etch and is a consequence of FIB milling, as confirmed by a HRTEM focal series reconstruction shown in supplementary information.\(^{30}\) We repeated the same experiments on three different sets of samples prepared from the same wafer and using the same protocol described above. Several sets of time series acquired on each TEM sample gave consistent results. Furthermore, control experiments were performed by operating the C\(_s\)-corrected HRTEM at accelerating voltages of 200 kV and 300 kV and these showed significant electron beam damage to our samples, which was manifested as contrast variations in HRTEM images with prolonged exposure time. However, C\(_s\)-HRTEM images acquired at 80 kV and diffraction contrast imaging done at 120 kV showed no significant electron beam damage. Since the illumination current density is significantly lower in C\(_s\)-STEM mode, we can conclude that our approach of acquiring STEM data at 120 kV does not cause significant beam damage in the investigated samples.

In addition to HAADF STEM imaging, another method to detect In segregation is by measuring the low-loss region of the EELS spectrum.\(^{12}\) After the zero-loss peak (ZLP), the most intense and easily identifiable peak in the low-loss EELS spectrum corresponds to the bulk plasmon resonance. The energy of the plasmon peak (\( E_p \)) is composition-
FIG. 2. Time-series of the top three InGaN QWs separated by GaN barriers acquired by C₃-STEM at 120 kV at (a) t = 0 min and (b) t = 16 min. (c) Higher magnification image of the middle QW showing that there are no In-rich clusters present in the InGaN QWs. Together, these images confirm that there is no beam induced damage at 120 kV.

Dependent and is, therefore, characteristic of the alloy composition. However, the exact dependence of \( E_p \) of a ternary alloy on elemental composition, \( x \), is not well understood. Some studies have suggested a linear dependence of \( E_p \) on \( x \), while some have suggested a variation or “bowing” of \( E_p \) consistent with Vergard’s Law. Thus, for In\(_{x}\)Ga\(_{1-x}\)N, the plasmon energy can be described as

\[
E_{p}^{\text{InGaN}} = x E_{p}^{\text{InN}} + (1-x) E_{p}^{\text{GaN}} - b_{\text{plasmon}} x (1-x),
\]

where \( E_{p}^{\text{InGaN}} \), \( E_{p}^{\text{InN}} \), and \( E_{p}^{\text{GaN}} \) are the plasmon energies of the respective materials and \( b_{\text{plasmon}} \) is the bowing parameter. While \( E_{p}^{\text{InN}} = 15.7 \text{ eV} \) (Ref. 28) and \( E_{p}^{\text{GaN}} = 19.15 \text{ eV} \) are well documented for thin films, there is no agreement on the value of \( b_{\text{plasmon}} \). By using the nominal In composition \( x = 0.22 \) and experimentally obtained average \( E_{p}^{\text{InGaN}} = 19.125 \text{ eV} \) and \( E_{p}^{\text{InGaN}} = 18.810 \text{ eV} \) (reported later in this manuscript), we calculate \( b_{\text{plasmon}} = -2.55 \text{ eV} \) (see supplementary information) and this value can be used to calculate the alloy composition for each measured \( E_{p}^{\text{InGaN}} \) value. In this work, we rely on the change in \( E_p \) measured along a QW to ascertain the existence or absence of In-clusters. The measured variation in \( E_p \) is used to calculate the corresponding change is In composition, \( \Delta x \), by assuming both a linear as well as a bowing relationship between \( E_p \) and \( x \).

Low-loss EELS data were acquired by scanning a focused 120 kV electron beam along several GaN and InGaN layers. The data have an energy dispersion of 0.1 eV/channel and an energy resolution of <600 meV, determined from the full width at half maximum of the ZLP. To check the validity and reproducibility of the low-loss EELS data, line scans were first acquired along two different GaN barrier regions in steps of 0.3 nm, and were analyzed individually to identify the position of \( E_p \). For each acquired spectrum, the tail of the ZLP was first removed using a Fourier-log routine, and care was taken to ensure that low-energy features were not lost in the process. A representative ZLP-subtracted EELS spectrum is shown in Figure S3 (supplementary information). \( E_p \) was extracted from the spectra and plotted as a function of the relative position along the line scan (Figure 3). This process was then repeated for two different InGaN QWs and also plotted in Figure 3. The EELS data were acquired in regions that are sufficiently away from each other because the build-up of carbon contamination during data acquisition over an extended period of time can lower the signal-to-noise ratio.

We measure \( E_p^{\text{InGaN}} \) to be 19.10 eV and 19.15 eV for the two GaN barriers identified in Figure 3. It is to be noted that \( E_p^{\text{InGaN}} \) remains constant for each GaN barrier, thereby confirming the expected compositional uniformity in GaN. The difference in \( E_p^{\text{GaN}} \) for the two different GaN barriers is within the error of measurement (±0.1 eV). These measured values are in agreement with the literature, again confirming the validity and reproducibility of our technique. By following the same procedure, \( E_p^{\text{InGaN}} \) was measured at each scan point for the two InGaN QWs identified in Figure 3. The average value of \( E_p^{\text{InGaN}} \) for QW 1, identified as InGaN 1 in Figure 3, is 18.84 eV and for QW 2, identified as InGaN 2 in Figure 3, is 18.78 eV. In composition in different QWs can be different even when grown under identical conditions; thus, the average \( E_p^{\text{InGaN}} \) is expected to be different in two different QWs. However, the variation of \( E_p^{\text{InGaN}} \) within individual QWs merits a more detailed discussion. It can be seen from Figure 3 that data points spanning a few-nm correspond to \( E_p^{\text{InGaN}} = (18.8–18.9) \text{ eV} \) in QW 1 and similarly \( E_p^{\text{InGaN}} = (18.75–18.85) \text{ eV} \) in QW 2. To investigate whether the observed ~0.1 eV changes in \( E_p^{\text{InGaN}} \) can be ascribed to...
large scale In clustering, we calculate the compositional change $\Delta x$ that would correspond to these variations (see supplementary information for more details). For QW1 (QW2), we find $\Delta x = 2.4\%$ or $5.2\%$ ($\Delta x = 3.0\%$ or $5.4\%$), depending upon the assumptions of linearity or bowing of $E_p$ with $x$, respectively. Such small variations are due to the statistical fluctuations of a random alloy and are compositionally too small to be attributed to the existence of large scale In-rich clusters. Furthermore, no additional secondary bulk plasmon peaks were observed in the EELS spectra collected along the InGaN QWs. Taken together, the EELS results are consistent with the STEM images from Figure 2, and imply that there is no large-scale In-clustering in either of the InGaN QWs.

In summary, we have employed Cs-corrected STEM imaging and EELS measurements below the knock-on threshold to determine the compositional variation in InGaN QWs. By combining careful sample preparation and low-voltage imaging, we have demonstrated the ability to characterize InGaN MQWs without introducing electron beam artifacts. The samples studied were the same as those investigated by Smeeton et al. and we confirm their observation that no gross-scale compositional fluctuations are present in those samples. However, it is possible that other samples—having different compositions or different processing parameters—may in fact show compositional fluctuations. While our results cannot imply that large-scale compositional fluctuations do not occur in any In$_x$Ga$_{1-x}$N sample, we clearly demonstrate that the samples under investigation do not contain large-scale compositional fluctuations, and, importantly, they are apparently not critical for strong light emission. Finally, we have also established a systematic methodology to investigate this question in other In$_x$Ga$_{1-x}$N heterostructures. It is anticipated that this approach can be used to clarify the role of structural inhomogeneities in light emission in these systems.

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30See supplementary material at http://dx.doi.org/10.1063/1.4807122 for focal series reconstruction and low-loss electron energy loss spectroscopy analysis.