

Direct Growth of Uniform In-rich InGaN on Si: A New Basic Technology.

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Abstract

Chemically uniform, compact and thick InGaN layers covering the full alloy composition range are grown on bare Si(111) substrates by plasma-assisted molecular beam epitaxy. Photoluminescence emission is observed up to room temperature from the ultra-violet to the near-infrared including the important 1.3 and 1.55 μm telecom wavelength bands. This opens the door for the direct integration of InGaN based devices with well-established Si technology.

1. Introduction

The growth of InGaN layers over the whole composition range directly on Si opens the door for novel device designs and applications [1], in particular for high In content $\text{In}_x\text{Ga}_{1-x}\text{N}$ ($x > 0.5$) when the emission and absorption of the material are extended to the near-infrared. The direct growth on Si was, however, so far considered an extremely challenging task due to the nucleation stage that leads to the formation of rough layers with strong composition fluctuations. To improve the InGaN crystal quality and uniformity of the growth, AlN and/or GaN buffer layers have been typically employed. Such buffer layers, however, electrically isolate the epitaxial InGaN from the underlying Si substrate, constraining the electrical functionality of the InGaN/Si heterojunction. For making InGaN devices cheaper due to reduction of the substrate cost and for the realization of structures such as InGaN/Si tandem solar cells [2] (where an Ohmic junction is naturally formed due to the band alignment of p-Si and n- $\text{In}_x\text{Ga}_{1-x}\text{N}$, for $x > 0.46$) direct epitaxial growth of highly uniform InGaN on Si is highly demanded.

2. Methods

In this study, all samples were grown by plasma-assisted molecular beam epitaxy (PA MBE). P-type doped Si substrates were degassed at 850 °C inside the growth chamber in order to remove the native oxide. Prior to the growth, the samples were exposed to active N for 5 min, at the same high temperature, to enable formation of a homogeneous 2–3 nm thick SiN_x layer. Then, bulk InGaN layers were grown at low temperatures, around 450 °C, with thicknesses between 350 and 500 nm. At such low temperatures, In desorption is negligible. Therefore,

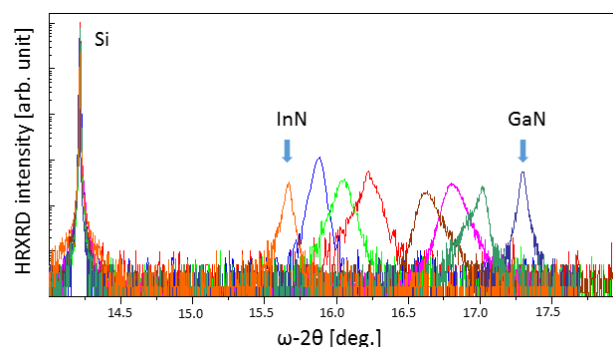


Fig. 1 HRXRD ω -2 θ spectra recorded around the symmetric InGaN(0002)/Si(111) Bragg reflections. The spectra of pure InN/Si and GaN/Si layers are shown for reference.

slightly N rich conditions were used to avoid metallic droplets accumulation. The In content is then easily adjusted by the Ga and In fluxes as all supplied metal incorporates into the layer.

3. Results

All samples were first analyzed by optical microscopy. No cracks were observed, which is a direct consequence of the low growth temperatures making the difference of the thermal expansion coefficients of InGaN and Si insignificant. The absence of regions with varying colors suggests that there are no prominent inhomogeneities on the macroscopic length scale. Scanning electron microscopy measurements revealed compact films, however with rough morphology, expected for the low growth temperatures.

High-resolution X-ray diffraction (HRXRD) measurements were done for each sample. ω -2 θ scans performed around the symmetric InGaN(0002) Bragg reflection are presented in Fig. 1. The In content was determined from reciprocal space mapping around the asymmetric InGaN(10-15) Bragg reflection. All analyzed samples exhibit a single InGaN diffraction peak which points to the absence of phase separation. This is somewhat surprising as for the present growth temperatures the miscibility gap of InN and GaN to form InGaN is predicted to span over almost the entire composition range from 5 to 95% [3]. However, at the same time such low growth temperatures likely kinetically suppresses phase separation, not consid-

ered in the thermodynamic models, which we believe is the main reason for uniform InGaN layers.

The uniform composition was further confirmed by high angle annular dark field images and punctual energy-dispersive X-ray spectroscopy measurements, both in Scanning Transmission Electron Microscopy (STEM) mode, revealing In fluctuations of only a few percent. Information about the InGaN/SiN_x/Si heterointerface, at the atomic scale, is provided by the high-resolution TEM images of samples with both medium ($x \sim 0.25$) and high ($x \sim 0.72$) In content (Fig. 2, top). The interfaces are sharp and exhibit an expected clear epitaxial relation InGaN(0001)||Si(111) which is not found to be affected by the 2 nm thick SiN_x interlayer.

The bottom of Fig. 2 shows the current-voltage (I-V) characteristics measured between front and back contacts for the samples with medium and high In content. The two InGaN/Si heterojunctions exhibit rectifying and ohmic behavior, respectively. It is to be noted that the In_xGa_{1-x}N layers with In content above 20% are highly n-type conductive due to defects, acting as donors. For samples with In content below 46%, there is a p-n junction at the Si/InGaN interface. However, layers with higher In content exhibit Ohmic behavior of the interface, as predicted [2], experimentally proving the feasibility of InGaN/Si tandem solar cells.

Both low (15 K) and room temperature photoluminescence (PL) measurements were done. The normalized PL spectra of the InGaN layers taken at 15 K are presented in Fig. 3. The integrated intensity was not varying more than a factor of two for samples covering the In content range from 20 to 88%. For these samples clear PL emission was also observed at room temperature where the integrated intensities were changing insignificantly, revealing high and comparable optical quality independent on the In content within the mentioned range. Lowering the In content below 20% resulted in stronger PL emission (not shown here), however, additional peaks associated with yellow band emitting defects appeared as expected for the low growth temperatures. The PL emission covers the full range from the ultra-violet to the near-infrared, including the important 1.3 and 1.55 μ m telecom wavelength bands, vastly enlarging the application fields of the InGaN materials system.

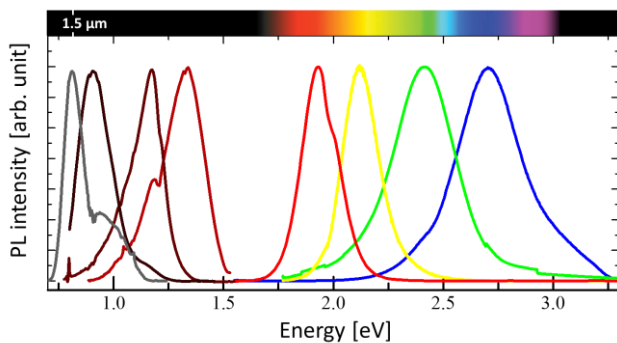


Fig. 3 Normalized PL spectra of the InGaN/Si samples taken at 15K.

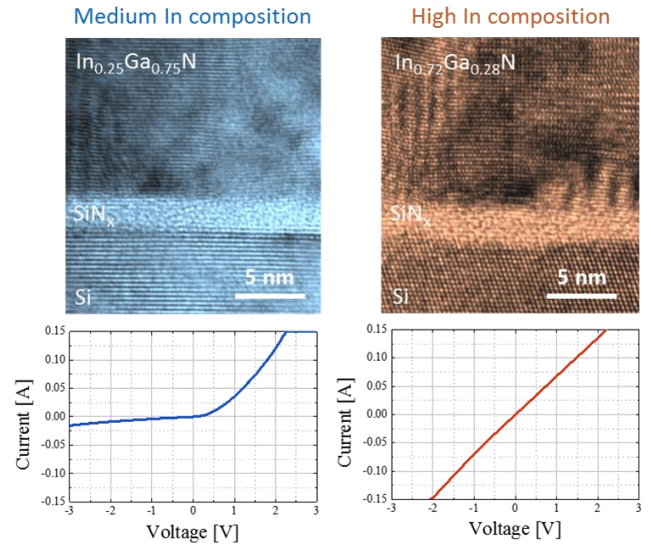


Fig. 2 HRTEM images of the InGaN layers with medium and high In content (top) taken along the [01-10] and [11-20] zone axis, respectively. I-V-curves (bottom) of the InGaN/Si samples measured between front and back contacts.

4. Conclusion

In this work we have demonstrated the growth of chemically uniform, compact, thick and crack-free InGaN layers on bare Si(111) (i.e., no buffer layers), covering the full alloy composition range. The high compositional uniformity is achieved by initial Si surface nitridation and the employment of low growth temperatures. However, this is accomplished at the cost of a nonplanar, undulated surface. High-resolution X-ray diffraction and energy-dispersive X-ray spectroscopy studies proved the chemical uniformity of the InGaN layers for the whole composition range. Transmission electron microscopy confirmed the InGaN(0001)||Si(111) epitaxial relationship. Moreover, the In_xGa_{1-x}N/Si heterointerfaces exhibited Ohmic behavior for $x > 0.46$. Both low (15 K) and room temperature PL measurements revealed emission covering a wide range from the ultra-violet to the near-infrared including the important 1.3 and 1.5 μ m telecom wavelength bands. Altogether, these results prove high quality and functionality of the InGaN layers grown on bare Si substrates, and open the road to develop novel InGaN based opto-electronic devices and their integration with Si technology, which, in addition, will significantly decrease the price of the final devices.

References

- [1] P. Aseev *et al.*, Appl. Phys. Express **6** (2013) 115503.
- [2] L. Hsu and W. Walukiewicz, J. Appl. Phys. **104** (2008) 24507.
- [3] I. Ho and G. Stringfellow, Appl. Phys. Lett. **69** (1996) 2701.