

Electroluminescence Characterization of Reverse Biased (In,Ga)N/(Al,Ga)N/GaN Systems Grown on SiC

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Abstract: Optical characterization of $In_xGa_{1-x}N/Al_yGa_{1-y}N/GaN$ grown on silicon carbide (SiC) is reported in this work. Room temperature electroluminescence (EL) was obtained for reverse bias (27V, 350 μ A) at 2.86eV (Blue luminescence) and at 1.98eV. At low temperature the yellow luminescence (YL) peaked at 2.16eV is also observed. The BL band is mostly given by electron-hole recombination in (In,Ga)N layer corresponding to the indium concentration equal to 15% (theoretically) and 13% (experimentally). The YL is more consistent with oxygen implication and can be explained by donor-acceptor (DAP) transition in which the complex $(V_{Ga} - O_N)^-$ is involved.

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1. Introduction

Gallium nitride (GaN) and its alloys with InN and AlN have gained an unprecedented attention due to their wide-ranging applications encompassing green, blue, violet and ultraviolet (UV) emitters and detectors, in photon energies inaccessible by others semiconductors, and high-power amplifiers. They have also received recently a great interest in the few years which is due to their potential applications in light emitting diodes (LEDs), LD sources for high-density optical storage systems, full-colour displays, medical applications and laser for aeronautic and space utilization. It has been recognized that the band gap range of the (In,Ga)N alloy is an almost perfect match for solar spectrum, which makes it a potential material for tandem solar cells [1-4]. However, the high density of defects, dislocations and high resistivity p-type cladding typical of heteroepitaxial material affect notably the electrical and optical properties of the host material and can seriously degrade and limit the performance and reliability of devices made based on these material semiconductors [5-6]. To overcome these problems, silicon carbide (SiC) and sapphire substrates have been utilized for the III-nitride heteroepitaxial growth [5-6]. On the other hand, the epitaxial lateral overgrowth (ELO) technique has further proven to be used effectively for obtaining low dislocation densities [7-10]. In this work, we investigated the temperature effect on the light emission obtained from (In,Ga)N/(Al,Ga)N/GaN diode. Our EL measurements were taken at the optic laboratory of physics department of Sherbrooke University in Canada.

2. Experimental

The CREE type C430-DH85 light emitting diode commercial was used in this work. A schematic diagram of studied diode is shown in Fig.1. A standard deviation of 6% was observed using 15 diodes. Spectral EL intensity was obtained using 1-meter Jarrel-Ash monochromator equipped with a cooled Hamamatsu R636-10

photomultiplier and photon counting system. To avoid the saturation of the photomultiplier response, which occurs when the signal is about 10^4 stronger than the dark noise, the neutral density filters were inserted at the appropriate EL values. For temperature dependence, the sample was placed at the cold finger of CTI Cryogenics closed system refrigerator.

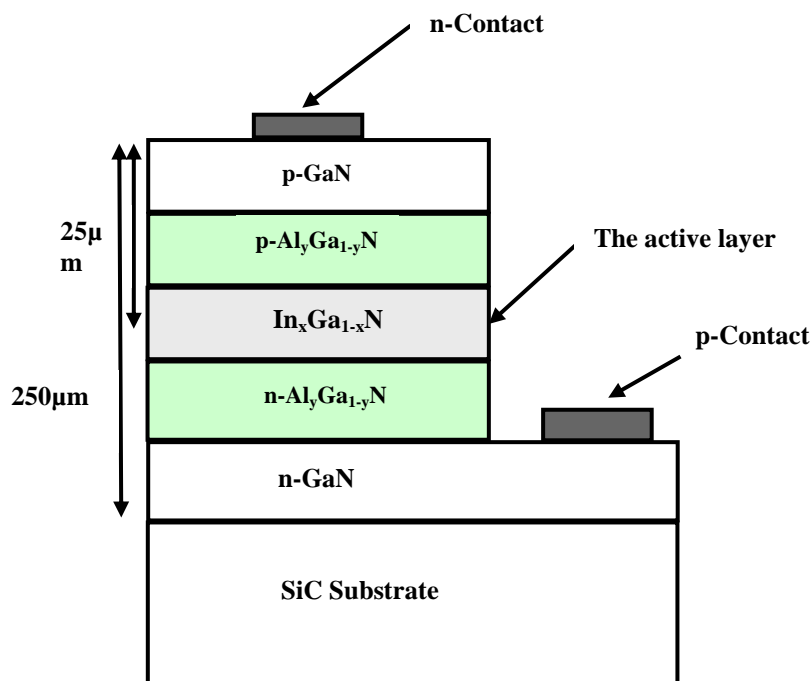


Figure 1: Schematic structure and composition of the InGaN LED under study according to the information supplied by the manufacturer.

3. Experimental results

Fig. 2 shows EL spectrum from unirradiated sample at 307K, obtained for a reverse bias of 27V and a current of 350 μ A. For all samples, we have observed a large structure with a full width half at maximum about 304 meV ($FWHM \sim 304meV$) situated around 2.86eV which corresponds to the blue luminescence (BL). A fine structure ($FWHM \sim 5.7meV$) located at 1.98 eV is also observed. The BL band peaking at about 2.86 eV in GaN has attracted substantial attention of the semiconductor investigations for many years [11-18]. This is due to the controversial assignment of its origin and important role that the related defect plays in GaN and its alloys. It's often observed in undoped, Si-, C-, Mg- and Zn- doped GaN. In the literature, there is several, albeit contradictory, concerning the YL band origin. Generally, the YL band is located in the energy range of 2.2 – 2.25eV [19-27]. The structure located at 1.98eV is the same as that reported by Skromme et al. [22]. This structure is not reported by all works the same material as that studied in this paper. For example, this structure is not identified in Ref. [22] i.e., its origin remains unknown.

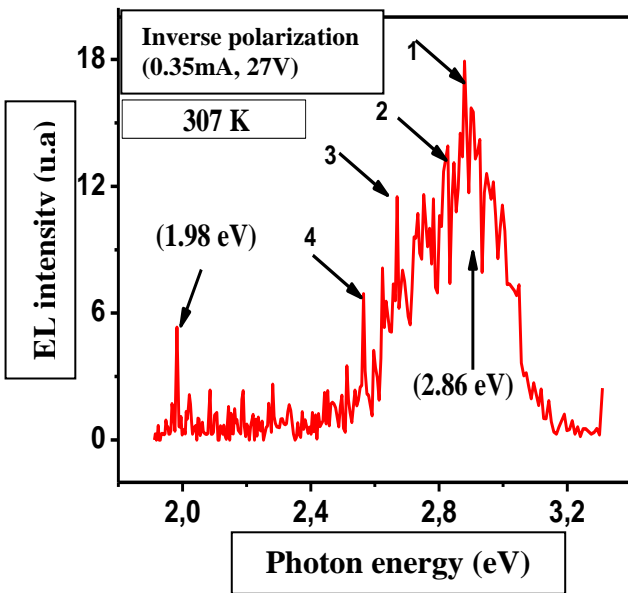


Figure 2: EL spectrum obtained from unirradiated diode as function of photon energy under breakdown conditions (350 μ A, 27 V) at 307 K.

The evolution of the EL spectra as function of the temperature shows an unusual behaviour. A low temperature dependence of the measured current is detected which is around of 350 μ A for a temperature range included between 26K and 307K. This effect shows that the thermal activation of free carriers contributing in light emission under breakdown conditions is low which can be attributed to the potential barrier increasing as a function of reverse voltage and thus the carriers must acquire a significant thermal energy to overcome this barrier.

Figure 3 shows the EL spectrum obtained at 26K for (27V, 350 μ A). The broadening and red shift (RS) of EL signal are observed. Indeed, the BL is located at $2.88 \pm 0.03meV$, $2.87 \pm 0.02meV$ and $2.87 \pm 0.03meV$ for temperatures equal to 26K, 250K and 307K respectively. The EL intensity as function of the temperature shows that the activation energy of this structure is about of $30.78 \pm 1.56meV$. A new fine yellow luminescence (YL) structure for which the intensity is comparable to that of BL band is detected. It's located at $2.16 \pm 0.04meV$ ($FWHM \sim 28.77 \pm 1.23meV$). The structure observed at 1.98eV disappeared completely at low temperature. The oscillations, noted 1, 2, 3 and 4, modulating the BL band at 307K persist even at 26K, which rejected their assignments to the phonons replicas. Thus, they are due to interference phenomena within layers constituting our samples.

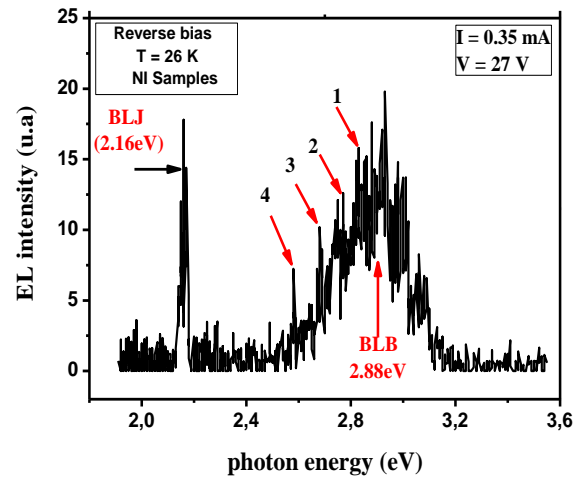


Figure 3: EL spectrum of unirradiated diode as function of photon energy under breakdown conditions (350 μ A, 27 V) at 26 K.

4. Discussion

The incorporation effect of InN and AlN semiconductors in GaN has been the subject of several studies [28-32]. Based on the theoretical study cited in [31], the band gap energy of $In_xGa_{1-x}N$ and $Al_yGa_{1-y}N$ ternary alloys at room temperature is calculated using the Luttinger-Kohn $4.4\vec{k} \cdot \vec{P}$ Hamiltonian. It is governed by the following formulas:

$$E_g(In_xGa_{1-x}N) = x E_g(InN) + (1-x) E_g(GaN) - 1.43 x (1-x) \quad (eV) \quad (1)$$

$$E_g(Al_yGa_{1-y}N) = y E_g(AlN) + (1-y) E_g(GaN) - 1.3 y (1-y) \quad (eV) \quad (2)$$

The experimental results reported by Davydov et al [32] show that this latter can be also expressed as follows:

$$E_g(In_xGa_{1-x}N) = 3.493 - 2.843x - 2.5x(1-x) \quad (3)$$

$E_g(InN)$, $E_g(GaN)$ and $E_g(AlN)$ are the band gap energies of InN , GaN and AlN at room temperature which are equal respectively to 0.77eV [33], 3.42eV [33-34] and 6.2eV [34].

For all indium and aluminium concentrations in $In_xGa_{1-x}N$ and $Al_yGa_{1-y}N$ respectively, the band gap energies of the latter are:

$$\begin{aligned} E_g(InN) &\leq E_g(In,Ga)N \leq E_g(GaN) \leq \\ E_g(Al,Ga)N &\leq E_g(AlN) \quad (4) \end{aligned}$$

The above analysis shows that the electrons provided by ionized donors in $n - GaN$ and $n - (Al,Ga)N$ or those injected when the diode is biased are guided to the potential well constituted by the $(In,Ga)N$ layer. In the same way, the holes originating of the acceptors ionization in $p - GaN$ and $p - (Al,Ga)N$ or those injected in the diode during its polarization leave the valence bands of the latter to accumulate in the same layer as that of the electrons. On the one hand, this analysis is in good agreement with that of Lefevbre et al [28] which have reported that growing the topmost $(Al,Ga)N$ barrier yields a stronger localization of carriers in the $(In,Ga)N$ system than if the barrier is GaN i.e., the optical emission in systems with an $(Al,Ga)N$ barrier remain principally due to localized electron-hole pairs, even at room temperature. On the other hand, it's more consistent with the information supplied by the manufacturer which indicates that the active layer is situated in $(In,Ga)N$ at $25\mu m$ below the surface top. According to the manufacturer, the active layer thickness is about $8 - 10\mu m$, and then, the quantification effect can be neglected in this situation. Then, we can assume that the BL result from the electron-hole recombination corresponding to In composition equal to 15% (using eq.1) and 13% (using eq.3). The In concentration obtained is approximately the same as that reported by Nakamura et al. [35] which show that the BL band ($450nm$) from $(In,Ga)N/(Al,Ga)N/GaN$ grown on sapphire is obtained for x around 20%. They reported also in others works that the In concentration is included between 10% and 25%, in particularly 16% for the blue emission ($417nm$) [36-37].

The red shift observed is more consistent with the band gap temperature dependence. Indeed, this displacement, which is approximately equal to $11.4 meV$, is lower than GaN band gap shift in the same temperature range ($60meV$). These results support our BL band attribution. Notice that, when the temperature increases, the emission blue shift, due to thermal activation of free carriers, is masked by the red displacement due to the band gap energy.

We have shown that, at $26K$, the structure located at $1.98eV$ is completely disappeared and a new structure appears at $2.16eV$. If we believe that these two structures have the same origin, so the fine structure picked at $1.98eV$ is moved to the blue ($183meV$) when the temperature varies from $307K$ to $26K$. This displacement is particularly important compared to that reported in the literature concerning the YL ($\sim 50meV$) and the GaN band gap ($\sim 60meV$). Indeed, variation of the shape and position of the YL with temperature has been the topic of several publications [16,38,39]. The position of the YL doesn't change within an accuracy of $20 - 40meV$ when

the temperature varied from about $10K$ to $300K$ [40], or $380K$ [16,38] and also in the range from $300K$ to almost $600K$ [39]. So, we can attribute this displacement to the change of the nature transition responsible of this emission.

In the literature, YL is observed in all GaN samples grown on different substrates and obtained by different growth techniques, i.e., it is independent of growth precursors. Its origin remains controversial.

Generally, the gallium vacancy (V_{Ga}) is the dominant native defect, particularly, in n-type GaN because its formation energy is low in this case. However, it could easily diffuse even at moderate temperatures of growth or thermal annealing and would readily form complexes with others defects. It should be underscored that formation of complexes is driven by the electrostatic forces. The impurities that are most likely to form stable complexes with V_{Ga} are donors. It is well-known that oxygen is the dominant contaminant of III-nitrides. The oxygen in the nitride site (O_N) in GaN corresponds to a shallow donor with activation energy of $33.2meV$ [41]. We have shown that the $1.98eV$ quenching takes place in $101 - 307K$ temperatures range with activation energy of about $30.78 \pm 1.56meV$. This result suggests that the residual defect species may be O_N and shows that the oxygen is implied in this emission. Several studies support this attribution. The study undertaken by Ploog et al. [42] showed that the residual concentration of oxygen is about $10^{16}cm^{-3}$. Other studies envisaged the concentrations of about $10^{18}cm^{-3}$ and $10^{20}cm^{-3}$ [43-44]. The correlation between the YL band intensity and the oxygen concentration has been noted also by Slack et al. [45]. Indeed, the concentration of both O_N and V_{Ga} and consequently that of $(V_{Ga} - O_N)$ increases as a function of oxygen doping. This is in good agreement with the attribution of this emission to the $(V_{Ga} - O_N)$ complex.

According to Neugebauer and Van de Walle calculations [13], $(V_{Ga} - O_N)$ act as a double acceptors in GaN and their $(-2-)$ energy levels (at $1.1 eV$ and $0.9 eV$ respectively) are close to the $(2-3-)$ transition level of the isolated (V_{Ga}) (at $1.1eV$). The formation of the $(V_{Ga} - O_N)$ complex is even more favourable than formation of isolated (V_{Ga}) . Similar results have been obtained by Mattila and Nieminen [46] who compared the formation energies of the isolated V_{Ga} and $(V_{Ga} - O_N)$ complex for different positions of Fermi level. The $(V_{Ga} - O_N)$ complex has a bending energy of $1.8eV$ which reject its direct involvement in this emission.

The results described above support the implication of $(V_{Ga} - O_N)^-$ or even $(V_{Ga} - O_N)^{2-}$ rather than $(V_{Ga} - O_N)$ in the YL. On the one hand, this is confirmed by the results reported by Elsner et al [47] which show that the dislocations in GaN induce locally a strain field which captures $(V_{Ga} - O_N)$ complex and can be at the YL origin. On the other hand, the first assignment is supported by Ogino and Aoki [21] study which show by PLE measurements that the YL band is activated from excitation energy equal to $2.6eV$. This latter is close to

2.52eV necessary to empty $(V_{Ga} - O_N)^-$ situated at 0.9eV.

Conclusion

Optical properties of $In_xGa_{1-x}N/Al_yGa_{1-y}N/GaN$ grown on SiC were investigated by temperature dependent EL experiments. The experimental data reveal that:

- BL is due to electron-hole recombination in $(In,Ga)N$ with indium concentration of about 15% (theoretically) and 13% (experimentally).
- YL is more consistent with Oxygen implication and it can be attributed to $(V_{Ga} - O_N)^-$ complex.

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