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Abstract

We studied the degradation of MOCVD-grown InGaN LEDs on Si substrates under constant current stressing. Characterizations using Deep Level Transient Spectroscopy and Electron Energy Loss Spectroscopy on active areas showed that the stressing had generated defects that have trap states at 0.26 eV below the conduction band edge ($E_c - 0.26$ eV) and that correlated with the active area's lower nitrogen content as compared to unstressed samples. The combination of Current-Voltage, Electroluminescence, Cathodoluminescence, and device simulations indicate that an increase in the density of these defects is correlated with an increase in the non-radiative carrier recombination that causes degradation in light emission. Preventing formation of these defects will be critical for improving InGaN-on-Silicon LED reliability.

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1. Introduction

The light emitting diode (LED) is a versatile light source that is used in many applications, from general lighting to specialized lighting such as a backlight for LCD displays.

The adoption of GaN-based LEDs has been so far limited by the cost of producing the substrates. Much effort has been dedicated by various groups to grow GaN on various types of substrate. The first and still popular choice is sapphire, with silicon carbide (SiC) as the closest competitor [1], [2]. However, these two options are expensive. The alternative is to use silicon (Si), which is gaining popularity. Large Si wafers provide a less expensive option to both sapphire and SiC.

The use of silicon substrates has the added benefit of allowing integration of III-V based LEDs on silicon CMOS platforms. This integration would enable the fabrication of circuits with some unique capabilities that combine the benefits of both III-V optical devices and Si CMOS technology. This integration will open up an even wider range of applications, such as intelligent LED lighting and wireless optical systems [3], [4].

The gallium-nitride (GaN)-on-silicon (Si) platform is not yet mature. One of the contributing factors is the difficulty in achieving low defect densities. As a result, device performance and eventually reliability are limited by the effects of these defects [5], [6]. Understanding of the origin and nature of these defects and of their role in device degradation is critical for improvement of LED reliability.

In this paper, we report studies of InGaN-on-silicon LED degradation in constant current operation. We suggest that the physical origin for light intensity degradation is related to the formation of nitrogen vacancies (V_N) in the quantum wells.

It is well established that nitrogen vacancies can form during epitaxial growth. In this work, we find that they can also form during device operation and that their generation can contribute to light intensity degradation, by acting as non-radiative recombination centers.

2. Experimental Details

We have tested InGaN-on-silicon LED devices that were fabricated with films grown using Metal-Organic Chemical Vapour Deposition (MOCVD), with a threading dislocation (TD) density of $2.8 \times 10^9 \text{ cm}^{-2}$. Five quantum well InGaN/GaN epi stacks were grown on Si substrates as shown in figure 1. The epi-stacks were fabricated into $300 \times 300 \mu\text{m}^2$ LEDs, with effective emission areas of $75000 \mu\text{m}^2$. The LEDs had a peak intensity wavelength at 450 nm. They were stressed with a 20 mA (current density of 26.7 A/cm^2) constant current at room temperature.

Figure 1 Device structure with five quantum well InGaN/GaN epi stacks that were grown on Si substrates.

The electrical tests consisted of a series of stress and measure cycles. The total stressing time was set to 100 hours. Every hour, the stressing was stopped to

measure the electrical and optical characteristics. All the electrical characterizations were carried out using an Agilent 4156C Parametric Analyzer. The electrical-optical (EL) characterization was carried out using Ocean Optics USB4000.

Selected devices were characterized using Deep Level Transient Spectroscopy (DLTS) for electrically active deep level traps [7]. In this study, we used a customized DLTS system with the ability to vary the measurement temperature from 80 to 400 K. The sample's capacitance was measured with high precision using Boonton 7200 capacitance meter that is equipped with capacitor banks for capacitance offset. With this setup, we were able to obtain a minimum trap density of $\approx (N_D - N_A) = 10^4 \text{ cm}^{-3}$ [8].

Scanning Transmission Electron Microscopy with Electron Energy Loss Spectroscopy (sTEM-EELS) was used to determine the physical origin of the traps. Several cross section samples were prepared using a focused ion beam system (FEI NanoLab 600i). The sTEM-EELS characterization was carried out using a JEOL 2100F operating at 200 kV.

Cathodoluminescence characterization was carried out to map the defect distribution. The study was conducted using an electron source in a JEOL JXA-8530F, and the light emission was captured with a Hamamatsu detector (photomultiplier type). A band pass filter was used to select the wavelengths of interest. Casino [9] software was used to determine the suitable electron energy.

3. Results

The light intensity was observed to gradually degrade during stressing, as shown in figure 2a. Annealing effects [10], [11] were observed in the first two hours of stressing as shown in the inset in figure 2a. The tested devices were considered to have failed when their intensity had dropped by 30% relative to their initial intensity.

Figure 2b shows the spectrum changes with stress. The peak intensity wavelength is observed at 450 nm, which corresponds to the band edge emission. The full width half maximum (FWHM) values are mostly constant around 16 nm throughout stressing. For increased clarity, the spectrum beyond 500 nm is given as inset in figure 2b in a semi-log plot. The observed light emission larger than 500 nm is most likely contributed by defect states [12]. Our results showed that no extra peaks have developed in those higher wavelength emissions. Thus, defects that formed did not seem to contribute to radiative recombination with longer wavelengths.

Figure 2. Typical light intensity degradation during stressing at 20 mA at room temperature: (a) integrated intensity decreases over time for 9 devices in the stressing sample, inset shows the initial increase in intensity due to annealing effect. The light intensity is observed to gradually degrade during stressing. (b) Selected spectrum from device 6. The light intensity decreases across the emission spectrum over time. Inset shows zoom into 500-750 nm in semi log scale. It shows that no extra peak developed which indicates stress generated defects contribute to non-radiative recombination.

Figure 3 shows DLTS analysis comparing fresh and the stressed samples with -1 V reverse bias, 0 V bias to fill the trap states and 100 ms pulse width. Figure 3a shows the DLTS results for 10 capacitance measurements per second rate, and shows two peaks for the stressed sample that are associated with defect states, while only one peak is observed in a fresh sample. They are around 150 K and 360 K, respectively.

The Arrhenius plot shows that the new developed defect has trap energy state at 0.26 eV below the conduction band edge ($E_c - 0.26 \text{ eV}$), as shown in figure 3b. The defect has a small capture cross section, $\sim 10^{-15} \text{ cm}^2$, which is a characteristic of point defects. Defects with such energy levels have been associated with nitrogen-vacancies (V_N) [13]–[17].

A defect with trap energy state at 0.7 eV below the conduction band ($E_c - 0.7 \text{ eV}$) and capture cross section $4 \times 10^{-15} \text{ cm}^2$ was also observed. Unlike the previous defect, this defect state has the same density between fresh and stressed samples. Hence, it can be concluded that it is not a result of the device stressing. This trap property is similar to traps reported in GaN on SiC [18] as well as GaN [19] on Si, but not yet well understood [8].

Figure 3. DLTS probing around the active area indicated a new broad peak formed in the stressed sample. (a) Comparison between fresh and stressed samples. (b) An Arrhenius plot shows the new peak has a trap energy state at $E_C - 0.26 \text{ eV}$ with an approximately 10^{-15} cm^2 capture cross section (not shown).

Cross-sectional EELS shows a lower nitrogen concentration in the first quantum well of the stressed samples, as seen in figure 4. The red circles and the blue squares in figure 4b represent the nitrogen concentration for fresh and stressed samples, respectively. A dislocation core with an edge component has been used as a reference point. In this region, we observed that the nitrogen concentration in the stressed sample is lower than in the fresh sample.

The nitrogen loss observed in the stressed sample combined with the DLTS results supports our suggestion that nitrogen vacancies V_N had formed during the test. Formation of nitrogen vacancies during device operation has not been previously reported.

Figure 4. EELS characterization. (a) STEM cross-sectional image of the LED's epi-stack in the active area, with a threading dislocation (TD) with an edge component. (b) Nitrogen quantification along the 1st quantum well, QW I. A lower nitrogen concentration is seen in the stressed sample.

Figure 5 shows IV curves for different stressing

time. A trend can be seen in which the IV curves shift to the right, except from 0-20 H, where they shift left. The right shift may indicate changes in the metal-semiconductor contact, while the early left shift may be attributed to the annealing effect [5], [6]. The right shift is similar to observations by Hamond *et al.* 2014, and they associated this effect with n-contact degradation [13].

Figure 5b focuses on IV curves at low voltages with the low injection currents plotted on a log scale for clarity. The low injection current is dominated by carrier tunnelling transport to the active region, which is associated with high doping and high defect densities [12], [21], [22]. This increase in the tunnelling component of the IV characteristic was observed for the first 20 H of stressing. Beyond that, we did not observe a consistent trend. For example, between the 20-40 H, this contribution decreases, and then increases again between 80-100 H.

Figure 5. IV characteristic evolution as a device is stressed. (a) Plotted in linear scale, (b) zoom version on the region enclosed by the dashed rectangle in (a).

Defect-assisted carrier recombination has been reported to increase in the yellow-red part of the spectrum [23]. However, scanning electron microscope cathodo-luminescence (SEM-CL) analyses did not show that an increase in the carrier recombination causes radiative recombination in the yellow-red (550-680 nm) part of the spectrum (figure 6). This result supports the conclusion that defects generated during stressing increase the non-radiative

recombination component of the carrier transport mechanism. Those defects, as characterized using EELS, are nitrogen vacancies.

Figure 6. SEM-CL intensity map from stressed and fresh samples collected with a yellow-red filter (550-680 nm) and 7 keV electron energy: (a) fresh sample, (b) stressed sample. The filter is useful to map the defect that acts as a radiative recombination center. The stressed sample showed a lower density of defects that contribute to radiative recombination.

Finally, TCAD simulations indicate that the defects that formed in the active area correlate with light emission degradation as shown in figure 7.

The simulations also show that metal-semiconductor interface degradation does not result in light intensity degradation in constant current stressing. This can be rationalized using the following arguments. The contact degradation is manifested as an increase in the contact resistance. As the testing condition was a constant current stressing, biasing instruments would simply increase the bias voltage to compensate for the increase in the resistance. Since the current was kept constant, the number of carriers available for recombination would also be the same. Hence there would be no decrease in the light emission. Therefore, even though we observed possible metal interface degradation from the IV characteristic, this does not seem to contribute to the light emission degradation in our tests.

There are multiple possible sources for the increased nitrogen vacancy concentration in the stressed devices. They can diffuse from the p-GaN layer or new vacancies can form in the quantum well during device operation.

It is less likely that nitrogen vacancies in the quantum well diffused from the p-GaN layer. Formation of nitrogen vacancies in the p-GaN layer is possible as this requires a relatively low energy of formation (~ 1 eV [24], [25], with some reports of less than 0.5 eV [26], [27]). However, it is relatively difficult for N vacancies to diffuse from the p-GaN layer to the quantum well as N vacancies are characterized by a very low diffusivity, i.e. 1.65×10^{-85} cm²/s at 300 K [28], [29].

While nitrogen vacancy formation in defect-free InGaN (the quantum well) normally requires a high energy of formation (~ 2 eV) [24], [30], the presence of defects can reduce the formation energy. One possible candidate is that Mg in the quantum well reduces the V_N formation energy. Mg might diffuse into the quantum well along threading dislocations. Mg diffusion along threading dislocations is thought to be relatively easy, with an activation energy of 0.28

eV [31], [32]. However, the Mg concentration is too low for our current instruments to detect (STEM EDX and STEM EELS). Alternative instruments that might help to detect the presence of Mg along dislocation include NanoSim and atomic probe tomography.

Figure 7. Silvaco-TCAD simulation suggests that an increase in the density of $E_c-0.26$ eV defects in the quantum well causes the decrease in light emission.

The $E_c - 0.26$ eV traps observed using DLTS likely correspond to the formation of nitrogen vacancies in the quantum well. This trap energy state is relatively shallow and provides a higher recombination rate (possibly due to Shockley-Read-Hall recombination) as compared to radiative band edge recombination [33]. The presence of the pre-existing ($E_c - 0.7$ eV) traps provide an alternative path for fast recombination.

4. Conclusion

In conclusion, we found that InGaN-on-Silicon LED degradation is associated with the formation of defects in the active area and that these defects cause an increase in non-radiative carrier recombination that decreases the light emission. DLTS measurements indicate a trap energy of $E_c-0.26$ eV and small capture cross section $\sim 10^{-15}$ cm². These characteristics allow identification of the traps as associated with nitrogen vacancies.

We propose that nitrogen vacancy formation is assisted by the presence with an impurity such as Mg. STEM-EELS measurements indicated that the nitrogen vacancy concentration was 50% higher near threading dislocations in stressed samples. We postulate impurities such as Mg diffuses along TDs from p-GaN towards n-GaN. These results suggest that reducing threading dislocation densities and minimizing the effects of impurities will lead to improved device reliability.

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