Ultraviolet-A LED Based on Quantum-disks-in-AlGaN-nanowires – Optimization and Device Reliability

Bilal Janjua,^{1,3} Davide Priante,^{1,3} Aditya Prabaswara,¹ Lafi Alanazi,² Chao Zhao,¹ Abdullah A. Alhamoud,^{1,2} Mohd Sharizal Alias,¹ Abdulrahman M. Albadri,² Ahmed Y. Alyamani,² Tien Khee Ng,^{1‡} *Senior Member, IEEE*, Boon S. Ooi^{1*}, *Senior Member, IEEE*

¹King Abdullah University of Science and Technology (KAUST), Photonics Laboratory, Thuwal 23955-6900, Saudi Arabia.

²National Center for Nanotechnology, King Abdulaziz City for Science and Technology (KACST), Riyadh, 11442-6086, Saudi Arabia.

³These authors contributed equally to this work.

[‡]tienkhee.ng@kaust.edu.sa

^{*}boon.ooi@kaust.edu.sa

Abstract: Group-III nitride-based ultraviolet (UV) quantum-disks (Qdisks) nanowires (NWs) light-emitting diodes grown on silicon substrates offer a scalable, environment-friendly, compact, and low-cost solution for numerous applications in solid-state lighting, spectroscopy, and biomedical. However, the internal quantum efficiency, injection efficiency, and extraction efficiency need to be further improved. The focus of this paper encompasses investigations based on structural optimization, device simulation, and device reliability. To optimize a UV-A (320-400 nm) device structure we utilize the self-assembled quantum-disk-NWs with varying quantum-disks thickness to study carrier separation in active-region and implement an improved p-contact-layer to increase output power. By simulation, we found a 100× improvement in the direct recombination rate for samples with thicker Qdisks thickness of ~5 monolayers (MLs) compared to the sample with ~2 MLs-thick Qdisks. Moreover, the sample with graded top Mg-doped AlGaN layer in conjunction with thin Mg-doped GaN layer shows 10× improvement in the output power compared to the samples with thicker top Mg-doped GaN absorbing contact layer. A fitting with ABC model revealed the increase in non-radiative recombination centers in the active region after a soft stress-test. This work aims to shed light on the research efforts required for furthering the UV NWs LED research for practical applications.

Index terms: Light-emitting diodes (LEDs); quantum disks, nanowires, nitrides, ultraviolet.

1. Introduction

AlGaN-based ultraviolet (UV) to deep-UV (DUV) (3.5 – 6.2 eV) light emitting diodes (LEDs) technology is attractive for applications in environmental cleaning, medicine, printing, microscopy and lighting [1, 2]. Although progress has been remarkable for planar LEDs based on this material, such diodes still exhibit low external quantum efficiency (EQE) close to 20 % for UV-A and < 1 % for UV-C devices [3-5]. One of the main challenges is the absence of economical lattice-matched native substrate technology, which results in high dislocation density exceeding 10^9 cm^{-2} . This limits the internal quantum efficiency (IQE) to only 40 % [6]. In addition, high polarization fields of several $\text{MV} \cdot \text{cm}^{-1}$ in such planar polar materials further increase carrier separation and make the useful radiative recombination process less efficient [7]. Furthermore, for large bandgap materials, reduced dopant ionization efficiency requires thick UV absorbing top p-GaN contact layer which considerably reduces the light extraction efficiency of such devices [1].

Group-III-V nanowires (NWs) geometry, on the other hand, have shown reduced excitation-induced energy shift and short radiative lifetime in the range of picoseconds [8], signifying that the quantum confined stark effect (QCSE) has low influence on these nano-structures. However, it has been recently reported that the center of InGaN quantum disks (Qdisks) has large strain-induced polarization field of $0.9 \text{ MV} \cdot \text{cm}^{-1}$, while being almost zero at the edge. This means that the reduced QCSE observed in NWs devices is a consequence of the high localization of carriers at the edge of the Qdisks, where the strain is minimal [9]. Moreover, NWs, can be grown catalyst-free on economical foreign substrate without the formation of threading dislocation in the active region [10]. Though a large free surface may lead to unwanted surface trap states, IQE values of more than 50 % have been demonstrated in AlN and AlGaN-based NWs, due to the spontaneous formation of AlGaN core-shell structure acting as a self-passivation layer [11, 12]. Despite the recent interest in UV NWs devices, much less attention has been geared towards the design of the NWs structure.

Active region designs based on both double heterostructure (DHT) [13, 14] and quantum confined structures [15-19] have shown promising results. Quantum-confined structures have the advantage of allowing increased carrier densities which directly result in lower recombination lifetimes and increase in radiative recombination rates. In addition, smaller linewidths for the EL spectrum have been obtained using Qdisks in NW. Currently, there are only very few works focusing on systematic active region design based on AlGaIn/GaN Qdisks in NW structures [20].

Other than device performance parameters such as optical power and quantum efficiencies, LEDs reliability constitutes an important factor for device applications. In fact, during operation, LEDs show a reduction in efficiency, and at elevated current, catastrophic failure. Different processes have been addressed as responsible for the optical power such as contact degradation, non-radiative recombination centers generation, tunneling, thermal emission etc [21-23]. However, the main origin is still under investigation. Few reports show temperature-related activation energies of the degradation process ranging from 0.13 to 0.27 eV correlated to the reduction in optical power and increase in junction temperature in planar AlGaIn-based LEDs [24-26]. Meneghini *et al.* analyzed the stability of GaN-based LEDs. They found that the prolonged injection current causes device degradation mainly due to non-radiative recombination centers in the active region while the high-temperature study induces ohmic contact degradation [27]. Notwithstanding the technology infancy of NWs compared to planar structures, it is important, for eventual application, to understand the mechanism behind such degradation.

Here we report a comparison of UV-A emitting structures with varying Qdisks thickness and a graded top Mg doped AlGaIn layer, all operating at room temperature. The devices with thicker Qdisks show a 100× higher direct recombination rates compared to the thin-Qdisks device. The LED with thicker top Mg-doped GaN layers showed lower turn-on voltage of around 6.1 V but at the trade of weaker collected signal. The device with UV-transparent graded p-AlGaIn layer obtained in unification with a thinner p-GaN layer showed 10× enhancement in the output power. A higher turn-on voltage of 9 V was due to the relatively higher Al composition in the Mg doped top AlGaIn layers. The degradation mechanism that limits the NWs LED reliability is studied using the LED rate equation based on the ABC model. At 16 A/cm², the device shows a good stability with no sign of degradation in optical power output. Lower non-radiative recombination center concentration was indirectly obtained from L-I curves, confirming the higher crystal quality of NWs compared to previously reported planar structures.

2. Experimental description

The UV NWs p-i-n LEDs structure was grown catalyst-free on n-type (100)-silicon substrate using plasma-assisted molecular beam epitaxy system (PAMBE) under N₂-rich conditions. Following HF treatment to remove the native oxide, the substrates were immediately loaded into the load-lock chamber. HF treatment results in hydrogen-terminated surface which repels oxide formation. Hence, further oxidation after HF cleaning is improbable. To remove any water components, the samples were outgassed in the load-lock at 200 °C using the IR filament. Outgassing at 650 °C was subsequently done in the buffer chamber to remove any organic-based contaminants. The substrate was then ramped up to 900 °C and kept there for 30 mins to ensure oxide-free surface. Up till the initiation of the growth, the wafer was kept away from the sources to avoid any unwanted nitridation of the surface.

The growth was initiated with a nucleation step at 702 °C to achieve high density, vertically aligned and uniform NWs. Following growth of ~150 nm of Si-doped GaN layer, the growth temperature was raised to 740 °C. A graded layer (n-GaN → n-Al_{0.27}Ga_{0.73}N) was introduced having a thickness of 20 nm. This ensured improved crystal quality of the 75 nm n-Al_{0.27}Ga_{0.73}N layer subsequently grown. The nominal Al composition was estimated by taking the ratio of Al with the total metal beam equivalent pressure (BEP) as measured by the beam flux monitor (BFM).

An active region with 15 GaN quantum-disks (Qdisks, ~2 monolayers (MLs) for sample 1 and ~5 MLs for samples 2 and 3) separated by Al_{0.27}Ga_{0.73}N quantum barriers (QBs), were then grown on the n-Al_{0.27}Ga_{0.73}N layer. For the QB, Al and Ga sources BEP were set to 1.5×10⁻⁸ and Ga 4×10⁻⁸ Torr while for GaN Qdisks another Ga source with BEP set to 3.5×10⁻⁸ was used. A ~10 nm magnesium (Mg) doped p-Al_{0.35}Ga_{0.65}N layer as electron blocking layer (EBL) was then grown using Al and Ga BEP of 2×10⁻⁸ and Ga 4×10⁻⁸ followed by a graded layer (p-AlGaIn → p-GaN) (10 nm for sample 1 and 2, 30 nm for sample 3) and p-GaN layer (40 nm for sample 1 and sample 2; 20 nm for sample 3) were grown. For optimized NWs morphology and density, nitrogen flow was maintained at 1 sccm with RF power fixed at 350 W.

The 0.5×0.5 mm² UV NW LEDs were fabricated using the standard UV contact lithography process. The samples were first cleaned by dipping in buffered HF solution for 10 s. Tilt-angle deposition using e-beam evaporator at 45° of Ni (5 nm)/Au (5 nm) was done directly on top of p-GaN layer, which formed an ohmic contact,

upon annealing at 550 °C in atmospheric ambient for 1 min. The thickness of Ni/Au was selected to provide good current spreading, as well as being sufficiently transparent to UV light. Ni (10 nm)/Au (500 nm) were then deposited as the top p-contact pad for probing. For n-contact, Si backside was etched by 200 nm using deep-reactive-ion-etching (DRIE) to expose a clean surface, followed by Ti (10 nm)/Au (150 nm) sputter deposition as n-contact pad.

Room temperature photoluminescence (RTPL) measurement was performed using a 266-nm pulse laser (Teem Photonics SNU-20F-100, 10 mW average power, ~1 mm in diameter). The PL signal was collected using a UV objective (Thorlabs LMU-15X-UVB, with anti-reflection coating (ARC) for the 240-360 nm range) and was focused into a monochromator (Andor Shamrock-750) using a 50:50 beam splitter (Thorlabs BSW19, with ARC for 250-450 nm range). Signal was measured using a cooled (-80 °C) UV/VIS silicon-based CCD camera (Andor iDus DU420A-OE) connected to the monochromator. Electroluminescent (EL) experiments were done using the PL setup with a camera attached to a beam-splitter mount. A current source (Keithley 2400) was used to inject direct current (DC) into the UV devices. The power was measured using a silicon photodetector with a power meter (Newport 818-UV and 1936-C, respectively). Scanning electron microscopy (FEI Nova Nano 630) was used to investigate morphology and area density of the NWs.

3. Results and discussion

Fig. 1. Structural characterization of the NWs. (a)-(c) Top-view SEM images show tightly packed, partially coalesced NWs. Insets show the statistical analysis of the NWs diameter. (d)-(f) Tilt-view SEM images show vertically aligned NWs with uniform height. Insets show the NWs length distribution charts.

Figs. 1(a)-(c) show the highly packed NWs top-view SEM image of the 3 samples with the diameter distribution charts. The diameters fall in the range of 40-70, 45-80, and 45-75 nm for sample 1, 2 and 3, respectively. Partial coalescence of the top p-GaN layers can also be observed, which prevents the top metal to coat the NWs sidewalls, and it helps in reducing carrier leakage during the device test. Figs. 1(d)-1(f) show the tilted-view SEM and the length distribution charts with values from 450-500, 460-500, and 420-490 nm for sample 1, 2 and 3. The areal density of the samples was also estimated to be $\sim 1.2 \times 10^{10}$, 9.2×10^9 , and $1 \times 10^{10} \text{ cm}^{-2}$, respectively. Due to the similar areal density obtained, we can emphasize the repeatability of the growth process. The NWs are expected to be n-polar as they were grown under nitrogen-rich conditions [28]. The NWs exhibit inverse tapered shape being thinner at the bottom ($\sim 30 \text{ nm}$) and reaching a diameter of $\sim 65 \text{ nm}$ at the top. This is due to the variation in growth temperature and slight lateral growth preference due to high Ga adatoms mobility. An encapsulation of active region

with few nanometers of AlGa_N having high Al content was also observed due to a limited Al-adatom diffusion-length.

Fig. 2. (a) Schematic of the NWs-based LED structures. (b) RTPL spectrum, with peak intensity at 328 nm for sample 1 and ~332 nm for sample 2 and 3 emitted from the quantum disk.

Fig. 2(a) shows the schematic structure of the three devices. Qdisk thickness in sample 2 and 3 is kept same at ~5 MLs and is double that of sample 1 (~2 MLs). In this design, thin Ga_N Qdisks are used to achieve shorter wavelength due to large quantum confinement. The design of sample 3 was further improved with a larger graded p-AlGa_N layer and reduced p-GaN thickness. This was done to improve light extraction properties of the UV device. The p-GaN thicknesses for sample 1 and 2 were kept constant while having similar grading scheme over 10 nm. Grading was done to reduce the energy barrier for electrons and holes entering the active region.

To investigate the emission properties of the three samples, PL measurements at room temperature were performed (Fig. 2(b)). Well defined emission peaks and PL intensity is an indication of improved carrier recombination mechanism in Qdisk structures with good AlGa_N crystalline quality. The PL emission from the three samples shows a blue shift due to the quantum confinement effects, compared to the expected ~364 nm emission from the bulk Ga_N. The slightly different emission peaks between the three samples are mainly due to the inhomogeneity across the wafer due to temperature gradient during the NWs growth. It was reported that strain distribution and polarization charges along the NWs highly affect the emission wavelength, showing variation from the Qdisks even within the same NW [29]. The three samples have also similar FWHM of 20 nm, 21.4 nm, and 22.1 nm. DHT based UV devices have shown to have linewidth up to 30 nm thus signifying the use of quantum confined structures in the active region [14]. The emission from the Qdisks has a smaller FWHM which correlates with homogeneous NWs and Qdisks with strong spatial localization (quantum confinement).

To study the quantum confinement effects in thin QWs, one-dimensional band modeling was performed using the nextnano³ software. As the NWs were grown on Si, crystal orientation with n-polarity was adopted and the band offsets $\Delta E_c/\Delta E_v$ at heterointerfaces were adjusted to be close to 70/30 ratio [30]. Figs. 3(a)-(c) show the calculated band diagrams for the three AlGa_N/Ga_N samples under forward bias of 3.5 V. Figs. 3(d)-(f) depict the calculated Shockley-Read-Hall (SRH), Auger and direct recombination rates, respectively. Samples 2 and 3 show a 100× improvement in the direct recombination rate due to the higher average carrier density (thicker Qdisks). By simulation, we estimated the average electrons concentrations of $1.6 \times 10^{18} \text{ cm}^{-3}$, with holes concentrations of $2.9 \times 10^{18} \text{ cm}^{-3}$ in sample 2 and 3; as well as $0.42 \times 10^{18} \text{ cm}^{-3}$ and $0.1 \times 10^{18} \text{ cm}^{-3}$, respectively, in sample 1. At higher current values, SRH recombination is expected to decrease while direct and Auger will increase with the n^2 and n^3 relationship, where n is the carrier density. In the simulation the non-radiative recombination rates in sample 2 and 3 scale with the carrier densities according to the carrier recombination loss rate relation: $R = An + Bn^2 + Cn^3$, where A , B and C are the SRH, radiative and Auger recombination coefficients, respectively. However, for large bandgap materials, Auger recombination coefficient is expected to be relatively low [31]. It was reported that NWs have reduced defect-assisted Auger recombination [32] compared to the planar counterparts, thus contributing to high device performance. Furthermore, having larger number of Qdisks reduces the overall carrier density in each case, thus further reducing non-radiative Auger recombination rate. The calculated bandgap is 3.78 eV (328 nm), which correlates well with the PL peak emission wavelength. It is noted that the AlGa_N-based NWs show the spontaneous formation of the Al-rich AlGa_N shell at the sidewalls, preventing the carriers from recombining at these interfaces

and potentially leading to reduced non-radiative surface recombination [28, 33]. Therefore, the surface recombination factor at the NWs sidewalls of the Qdisks was not taken into account in the calculation of the recombination rates.

Fig.3. (a)-(c) Energy band diagrams under 3.5 V forward bias. (d)-(e) SRH, Auger and direct recombination rates in the Qdisks.

We then characterized the three NWs UV LED samples at different DC-biases for L-I-V characteristics as illustrated in Fig. 4(a)-(c). From the I-V plot, the turn-on voltage and respective series resistance for sample 1, 2 and 3 were measured to be around 6.1, 6.1, and 9 V, and 18.3, 14.5, and 77.8 Ω , respectively. For sample 1 and 2, I-V behavior is comparable as predicted due to the similar doped layers. The apparent higher series resistance and turn-on voltage for sample 3 can be attributed to the larger average bandgap for the Mg-doped graded layers and lower ionization of dopants. For all samples, the insulating SiN_x layer at the NWs/silicon interface also introduces a resistive layer resulting in larger turn-on voltage compared to their bandgaps. Other published works have reported onset voltages of up to 7.2 V for UV-A (320-400 nm) NW-based LEDs [14, 16, 28, 34, 35]. It is also noted that the Mg activation energy increases almost linearly with Al content, as well as the resistivity. Moreover, the deep acceptor states induced by Mg doping reduces the amount of carriers reaching the active region, hence increasing the turn on voltage. The large turn-on voltage of sample 3, can then be reduced by optimizing the growth of the p-dopant in the p-AlGaIn layer, therefore reducing the deep level recombination.

Fig. 4. Electrical characterization of the NW devices. (a)-(c) L-I-V characteristics of sample 1, 2 and 3 respectively, showing turn-on voltages (red dotted lines). (d)-(f) EL spectra of NW devices at different injection currents from 10-100 mA corresponding to sample 1, 2 and 3, respectively.

Having larger active region volume and therefore higher direct recombination rates, sample 2 and 3 exhibited higher optical power compared to sample 1 at similar injection current. This is evident in Fig. 4(a) that sample 1 requires 160 mA to achieve the same output power as that of sample 2 at 100 mA. Moreover, a significant improvement in output power was observed for sample 3 (p- $\text{Al}_{0.35}\text{Ga}_{0.65}\text{N} \rightarrow \text{p-GaN/p-GaN}$ as p-contact layer) compared to sample 2 (p-GaN as p-contact layer), both grown with nominally the same thickness of p-contact layer. This difference is due to the significant absorption in the UV regime for sample 2 having all p-GaN contact layer, thus reducing light extraction. Nevertheless, a thinner p-GaN in sample 3, with $\text{Al}_{0.35}\text{Ga}_{0.65}\text{N} \rightarrow \text{p-GaN}$ graded layer led to a larger turn-on voltage.

Figs. 4(d)-(f) show the room temperature band-edge electroluminescence (EL) under different DC biases for samples 1, 2 and 3. All three samples show peak emission at ~ 340 nm along with a hump beyond ~ 360 nm which can be ascribed to recombination in the sub-bandgap extended states in the p-GaN layer [36]. The slight differences observed in the emission spectra shape are mainly due to the inhomogeneity introduced by NWs initial nucleation, which is a thermodynamically-governed random process. This leads to variation in morphology of the NWs and in turn emission inhomogeneity. For the same reason, we obtained slightly different emission in the PL spectrum. Similar differences between PL and EL emissions have been observed by Sarwar et al. [29]. The corresponding linewidths of the samples are 15.4, 13.3, and 13.6 nm as obtained with a Gaussian function fitting. Further LEDs characterization reveals that the emission peaks are nearly independent of injection current and display a negligible blue shift when the injection current was increased from 10 to 100 mA (see Fig. (d)-(f)). An insignificant band filling effect is due to a reduced alloy broadening effect in AlGaIn under optimized growth kinetics [37, 38]. The blue-shift behavior, however, does occur in AlGaIn quantum well based planar devices in the presence of high polarization fields [39]. However, such effect is reduced in NWs due to lateral strain relaxation at the nucleation site [10, 40, 41].

In order to investigate the device degradation mechanism, sample 3, which has the highest output power, was subjected to electrical stress over 5 hours period. Fig. 5(a) shows the power-voltage plot as a function of aging time at $16 \text{ A}/\text{cm}^2$ (40 mA) injection current at room temperature. The relatively low injection current used ensures that,

during stability test, the device self-heating is limited with respect to room temperature, and hence the effects of thermal stress can be attributed principally to carrier flow [27]. The output power and operating voltage stay almost constant, signifying that the device can sustain such injection current without sign of degradation or catastrophic failure. We analyzed the dominant recombination mechanisms in the device before and after stress. From charge neutrality conditions we can express the LED rate equation as [42]:

$$\frac{dn}{dt} = \frac{J}{qd} - An - Bn^2 - Cn^3 - \frac{J_L}{ql_m} \quad (1)$$

where d is the active layer thickness, l_m is the minority carrier diffusion length, J is the current density, q is the unit charge, A , B and C are the Shockley-Read-Hall, radiative and Auger recombination rates, respectively, and J_L is the leakage current. Under steady state $\frac{dn}{dt} = 0$, and for the case of large band-gap materials such as AlGaIn, the Auger recombination can be ignored and does not change with stress time. Moreover, under high injection current, and if radiative recombination is more prominent than the non-radiative, eq. 1 can be rewritten as:

$$\frac{J}{qd} \sim Bn^2 = L \quad (2)$$

where L is the output power that is proportional to J having a slope ~ 1 in the log-log L-I plot. The optical power can then be expressed with a linear behavior as:

$$P = \alpha \cdot I \quad (3)$$

where α is a luminescence factor (W/A for $I > I_{th}$) [43].

On the contrary, if the non-radiative processes are more pronounced than radiative, L is proportional to J^2 since

$$\frac{J}{qd} \sim An \quad (4)$$

$$L = Bn^2 \sim \frac{B}{A^2} \left(\frac{J}{qd} \right)^2 \quad (5)$$

hence, the slope of the log-log L-I curve is ~ 2 . Therefore, the optical power follows a quadratic behavior as:

$$P = \gamma \cdot I^2 \quad (6)$$

where γ is a luminescence factor (W/A^2 for $I < I_{th}$).

Fig. 5(b) depicts the power-current plot after a 5-h soft stress test. As stated above, we can distinguish two regimes where the power follows a quadratic and linear behavior. We noticed a change in optical power curves between the samples tested before and after stress test. Using eq. 6, we extracted the γ factors for both devices. A decrease of approximately 30 % was observed. On the other hand, α factor remained almost constant. Considering Eq. (5), we can then conclude that (B/A^2) is the only factor that causes the optical power degradation, hence only the LED active region is degraded.

Furthermore, by extrapolating the slope of the log-log L-I curve at low and high current injection we studied the dominant recombination mechanisms as shown in Fig. 5(c). Before and after stress the slopes in the high injection regime ($J > 20 \text{ A/cm}^2$) are 0.93 (P_{HI} , in blue) and 0.95 (P_{HI} , in red), confirming that the non-radiative recombination rates saturate and the radiative recombination rates dominate (slopes ~ 1). On the other hand, at low injection current ($J < 8 \text{ A/cm}^2$) the slopes are 1.09 (P_{LO} , in black) and 2.23 (P_{LO} , in orange), respectively, inferring that non-radiative processes dominate after electrical stress (slope ~ 2), leading to device degradation related to non-radiative recombination centers [44].

Figure 5(c) also shows a comparison with a GaN-based LED before and after stress test performed by Meneghini *et al.* [27] Similar to our results, at high current the slopes show values of 0.95 and 1.05 for samples before (P_{HI} , in blue) and after stress (P_{HI} , in red), respectively, meaning that radiative recombination rate dominates

in this regime (slopes ~ 1). At low current, however, the values slightly differ, namely higher than AlGaIn NWs before stress (1.51 for GaN planar, 1.09 for AlGaIn/GaN NWs) and lower after stress (1.81 for GaN planar, 2.23 for AlGaIn/GaN NWs). This can be explained considering the lower concentration of defect states in NWs compared to planar structures that limits the non-radiative recombination processes.

It is noted that the generation of non-radiative centers is not the only cause of device degradation. As reported in early works, ohmic contact damage [21], enhanced tunneling [22], and Mg-related acceptors complexes related effect [23] contribute to device degradation. However, the contact damage was not noticeable on the NWs device, suggesting that the increase in defect density is the dominating mechanism of degradation.

Fig. 5. Soft stress test of sample 3. (a) Plot of power and voltage versus degradation time at $16 \text{ A}/\text{cm}^2$ injection current, showing stable performance. (b) Plot of power (P)-current (I) curve after 5-h stress period. (c) Logarithmic plot of power versus current before and after 5-h stress test showing the fitted slopes at low and high current injection for NWs LED (AlGaIn/GaN NWs, i.e. the present work) and GaN-based planar LED (Meneghini et al.).

4. Conclusions

In summary, the different design of AlGaIn-GaN-based UV-A NWs LED grown on Si were compared. The vertically aligned NWs were grown using PAMBE with density, diameter and length in the range of $\sim 9.2 \times 10^9 - 1.2 \times 10^{10} \text{ cm}^{-2}$, $\sim 40-80 \text{ nm}$, and $\sim 420-500 \text{ nm}$ respectively. A $100\times$ improvement in the direct recombination rate was simulated for the samples with thicker nominal Qdisks thickness of $\sim 5 \text{ MLs}$ compared to the $\sim 2 \text{ MLs}$ -thick Qdisks sample, due to the larger average carrier density. Narrow EL linewidth of $\sim 15 \text{ nm}$ and a negligible shift in EL peak wavelength for all the samples were obtained in the presence of Qdisks, confirming the compositional homogeneity,

efficient quantum confinement, and negligible QCSE effect. A thicker graded p-AlGaIn layer and thinner p-GaN contact-layer gave the highest optical power thus emphasizing the need for a transparent conductive top p-layer for devices having higher light extraction efficiencies. However, such scheme may lead to larger turn-on voltage, thus a tradeoff exists. A stress test of a NWs-LED completes the study in understanding the contribution of non-radiative recombination defect center after extended device operations. Although UV-A planar devices already show EQE of ~20 %, the NWs on silicon platform provide a cheap alternative with a potential to surpass the commercially available EQE due to the low dislocation densities and the reduced polarization fields that is still present in the planar counterparts. We argue that due to the infancy of the NWs technology, more work is required on designing and processing the LEDs structure. For example, attention is required on the p-GaN top layer that hampers the light extraction, improvement in the injection efficiency, and passivation studies of the NWs surface. This work thus shed insights into potential advancement of the group-III-nitride NWs research towards practical and eventual commercial applications across the full UV spectrum.

Acknowledgement

We acknowledge the financial support from the King Abdulaziz City for Science and Technology (KACST), Grant No. KACST TIC R2-FP-008. This work was partially supported by the King Abdullah University of Science and Technology (KAUST) baseline funding, BAS/1/1614-01-01, and MBE equipment funding C/M-20000-12-001-77.

References

- [1] A. Khan, K. Balakrishnan, and T. Katona, "Ultraviolet light-emitting diodes based on group three nitrides," *Nature photonics*, vol. 2, pp. 77-84, 2008.
- [2] H. Hirayama, "Advances of AlGaIn-based high-efficiency deep-UV LEDs," in *Asia Communications and Photonics Conference and Exhibition*, 2010, p. 79870G.
- [3] T. Takano, T. Mino, J. Sakai, N. Noguchi, K. Tsubaki, and H. Hirayama, "Deep-ultraviolet light-emitting diodes with external quantum efficiency higher than 20% at 275 nm achieved by improving light-extraction efficiency," *Applied Physics Express*, vol. 10, p. 031002, 2017.
- [4] S.-i. Inoue, N. Tamari, and M. Taniguchi, "150 mW deep-ultraviolet light-emitting diodes with large-area AlN nanophotonic light-extraction structure emitting at 265 nm," *Applied Physics Letters*, vol. 110, p. 141106, 2017.
- [5] M. Shatalov, W. Sun, A. Lunev, X. Hu, A. Dobrinsky, Y. Bilenko, *et al.*, "AlGaIn deep-ultraviolet light-emitting diodes with external quantum efficiency above 10%," *Applied Physics Express*, vol. 5, p. 082101, 2012.
- [6] M. Kneissl, T. Kolbe, C. Chua, V. Kueller, N. Lobo, J. Stellmach, *et al.*, "Advances in group III-nitride-based deep UV light-emitting diode technology," *Semiconductor Science and Technology*, vol. 26, p. 014036, 2010.
- [7] M. Feneberg and K. Thonke, "Polarization fields of III-nitrides grown in different crystal orientations," *Journal of Physics: Condensed Matter*, vol. 19, p. 403201, 2007.
- [8] J. B. Schlager, K. A. Bertness, P. T. Blanchard, L. H. Robins, A. Roshko, and N. A. Sanford, "Steady-state and time-resolved photoluminescence from relaxed and strained GaN nanowires grown by catalyst-free molecular-beam epitaxy," *Journal of applied physics*, vol. 103, p. 124309, 2008.
- [9] W.-C. Tsai, Y.-T. Chen, C.-H. Lin, W.-T. Hsu, Y.-S. Hsu, L.-C. Chen, *et al.*, "Suppressed piezoelectric polarization in single InGaIn/GaN heterostructure nanowires," *Physical Review B*, vol. 88, p. 155323, 2013.
- [10] Q. Wang, S. Zhao, A. Connie, I. Shih, Z. Mi, T. Gonzalez, *et al.*, "Optical properties of strain-free AlN nanowires grown by molecular beam epitaxy on Si substrates," *Applied Physics Letters*, vol. 104, p. 223107, 2014.
- [11] Q. Wang, H. Nguyen, K. Cui, and Z. Mi, "High efficiency ultraviolet emission from Al_xGa_{1-x}N core-shell nanowire heterostructures grown on Si (111) by molecular beam epitaxy," *Applied Physics Letters*, vol. 101, p. 043115, 2012.
- [12] S. Zhao, A. Connie, M. Dastjerdi, X. Kong, Q. Wang, M. Djavid, *et al.*, "Aluminum nitride nanowire light emitting diodes: Breaking the fundamental bottleneck of deep ultraviolet light sources," *Scientific reports*, vol. 5, p. 8332, 2015.
- [13] S. Zhao, M. Djavid, and Z. Mi, "Surface emitting, high efficiency near-vacuum ultraviolet light source with aluminum nitride nanowires monolithically grown on silicon," *Nano letters*, vol. 15, pp. 7006-7009, 2015.
- [14] Q. Wang, A. Connie, H. Nguyen, M. Kibria, S. Zhao, S. Sharif, *et al.*, "Highly efficient, spectrally pure 340 nm ultraviolet emission from Al_xGa_{1-x}N nanowire based light emitting diodes," *Nanotechnology*, vol. 24, p. 345201, 2013.
- [15] A. G. Sarwar, B. J. May, J. I. Deitz, T. J. Grassman, D. W. McComb, and R. C. Myers, "Tunnel junction enhanced nanowire ultraviolet light emitting diodes," *Applied Physics Letters*, vol. 107, p. 101103, 2015.
- [16] H. Sekiguchi, K. Kato, J. Tanaka, A. Kikuchi, and K. Kishino, "Ultraviolet GaN-based nanocolumn light-emitting diodes grown on n-(111) Si substrates by rf-plasma-assisted molecular beam epitaxy," *physica status solidi (a)*, vol. 205, pp. 1067-1069, 2008.
- [17] B. Janjua, H. Sun, C. Zhao, D. H. Anjum, D. Priante, A. A. Alhamoud, *et al.*, "Droop-free Al_xGa_{1-x}N/Al_yGa_{1-y}N quantum-disks-in-nanowires ultraviolet LED emitting at 337 nm on metal/silicon substrates," *Optics Express*, vol. 25, pp. 1381-1390, 2017.
- [18] B. Janjua, H. Sun, C. Zhao, D. H. Anjum, F. Wu, A. Alhamoud, *et al.*, "Self-planarized quantum-disks-in-nanowires ultraviolet-B emitters utilizing pendeo-epitaxy," *Nanoscale*, 2017.
- [19] D. Priante, B. Janjua, A. Prabaswara, R. C. Subedi, R. T. Elafandy, S. Lopatin, *et al.*, "Highly uniform ultraviolet-A quantum-confined AlGaIn nanowire LEDs on metal/silicon with a TaN interlayer," *Optical Materials Express*, vol. 7, pp. 4214-4224, 2017.
- [20] A. Sarwar, B. J. May, and R. C. Myers, "Effect of quantum well shape and width on deep ultraviolet emission in AlGaIn nanowire LEDs," *physica status solidi (a)*, 2015.

- [21] D. L. Barton, M. Osinski, P. Perlin, P. G. Eliseev, and J. Lee, "Single-quantum well InGaN green light emitting diode degradation under high electrical stress," *Microelectronics Reliability*, vol. 39, pp. 1219-1227, 1999.
- [22] A. Polyakov, N. Smirnov, A. Govorkov, J. Kim, B. Luo, R. Mehandru, *et al.*, "Enhanced tunneling in GaN/InGaN multi-quantum-well heterojunction diodes after short-term injection annealing," *Journal of applied physics*, vol. 91, pp. 5203-5207, 2002.
- [23] M. Pavesi, M. Manfredi, G. Salviati, N. Armani, F. Rossi, G. Meneghesso, *et al.*, "Optical evidence of an electrothermal degradation of InGaN-based light-emitting diodes during electrical stress," *Applied physics letters*, vol. 84, pp. 3403-3405, 2004.
- [24] J. Glaab, C. Ploch, R. Kelz, C. Stölmacker, M. Lapeyrate, N. L. Ploch, *et al.*, "Degradation of (InAlGa) N-based UV-B light emitting diodes stressed by current and temperature," *Journal of Applied Physics*, vol. 118, p. 094504, 2015.
- [25] D. Monti, M. Meneghini, C. De Santi, G. Meneghesso, E. Zanoni, J. Glaab, *et al.*, "Defect-Related Degradation of AlGaIn-Based UV-B LEDs," *IEEE Transactions on Electron Devices*, vol. 64, pp. 200-205, 2017.
- [26] Z. Gong, M. Gaevski, V. Adivarahan, W. Sun, M. Shatalov, and M. Asif Khan, "Optical power degradation mechanisms in AlGaIn-based 280 nm deep ultraviolet light-emitting diodes on sapphire," *Applied physics letters*, vol. 88, p. 121106, 2006.
- [27] M. Meneghini, L.-R. Trevisanello, G. Meneghesso, and E. Zanoni, "A review on the reliability of GaN-based LEDs," *IEEE Transactions on Device and Materials Reliability*, vol. 8, pp. 323-331, 2008.
- [28] K. Li, X. Liu, Q. Wang, S. Zhao, and Z. Mi, "Ultralow-threshold electrically injected AlGaIn nanowire ultraviolet lasers on Si operating at low temperature," *Nature nanotechnology*, vol. 10, pp. 140-144, 2015.
- [29] A. G. Sarwar, B. J. May, M. F. Chisholm, G. J. Düscher, and R. C. Myers, "Ultrathin GaN quantum disk nanowire LEDs with sub-250 nm electroluminescence," *Nanoscale*, vol. 8, pp. 8024-8032, 2016.
- [30] J. Piprek, *Nitride semiconductor devices: principles and simulation*: John Wiley & Sons, 2007.
- [31] D. Gallagher, "Optoelectronic Devices: Advanced Simulation and Analysis," *Springer*, vol. 1, pp. 7-10, 2005.
- [32] W. Guo, M. Zhang, P. Bhattacharya, and J. Heo, "Auger recombination in III-nitride nanowires and its effect on nanowire light-emitting diode characteristics," *Nano letters*, vol. 11, pp. 1434-1438, 2011.
- [33] S. Zhao, A. Connie, M. Dastjerdi, X. Kong, Q. Wang, M. Djavid, *et al.*, "Aluminum nitride nanowire light emitting diodes: Breaking the fundamental bottleneck of deep ultraviolet light sources," *Scientific reports*, vol. 5, 2015.
- [34] A. Sarwar, S. D. Carnevale, F. Yang, T. F. Kent, J. J. Jamison, D. W. McComb, *et al.*, "Semiconductor Nanowire Light-Emitting Diodes Grown on Metal: A Direction Toward Large-Scale Fabrication of Nanowire Devices," *Small*, vol. 11, pp. 5402-5408, 2015.
- [35] B. H. Le, S. Zhao, X. Liu, S. Y. Woo, G. A. Botton, and Z. Mi, "Controlled Coalescence of AlGaIn Nanowire Arrays: An Architecture for Nearly Dislocation-Free Planar Ultraviolet Photonic Device Applications," *Advanced Materials*, vol. 28, pp. 8446-8454, 2016.
- [36] D.-S. Shin, D.-P. Han, J.-Y. Oh, and J.-I. Shim, "Study of droop phenomena in InGaIn-based blue and green light-emitting diodes by temperature-dependent electroluminescence," *Applied Physics Letters*, vol. 100, p. 153506, 2012.
- [37] V. Fellmann, P. Jaffrennou, D. Sam-Giao, B. Gayral, K. Lorenz, E. Alves, *et al.*, "Ternary AlGaIn Alloys with High Al Content and Enhanced Compositional Homogeneity Grown by Plasma-Assisted Molecular Beam Epitaxy," *Japanese Journal of Applied Physics*, vol. 50, p. 031001, 2011.
- [38] G. Steude, B. K. Meyer, A. Göldner, A. Hoffmann, F. Bertram, J. Christen, *et al.*, "Optical investigations of AlGaIn on GaN epitaxial films," *Applied Physics Letters*, vol. 74, pp. 2456-2458, 1999.
- [39] H. Yoshida, Y. Yamashita, M. Kuwabara, and H. Kan, "A 342-nm ultraviolet AlGaIn multiple-quantum-well laser diode," *nature photonics*, vol. 2, pp. 551-554, 2008.
- [40] W. Guo, M. Zhang, A. Banerjee, and P. Bhattacharya, "Catalyst-free InGaIn/GaN nanowire light emitting diodes grown on (001) silicon by molecular beam epitaxy," *Nano letters*, vol. 10, pp. 3355-3359, 2010.
- [41] M. Knelangen, V. Consonni, A. Trampert, and H. Riechert, "In situ analysis of strain relaxation during catalyst-free nucleation and growth of GaIn nanowires," *Nanotechnology*, vol. 21, p. 245705, 2010.
- [42] P. N. Grillot, M. R. Krames, H. Zhao, and S. H. Teoh, "Sixty thousand hour light output reliability of AlGaInP light emitting diodes," *IEEE Transactions on device and materials reliability*, vol. 6, pp. 564-574, 2006.
- [43] M. Fukuda, "Reliability and degradation of semiconductor lasers and LEDs," *Artech House*, pp. 325-333, 1991.
- [44] T. Egawa, T. Jimbo, and M. Umeno, "Characteristics of InGaIn/AlGaIn light-emitting diodes on sapphire substrates," *Journal of applied physics*, vol. 82, pp. 5816-5821, 1997.