Role of quantum-confined stark effect on bias dependent photoluminescence of N-polar GaN/InGaN multi-quantum disk amber light emitting diodes

Malleswararao Tangi, Pawan Mishra, Bilal Janjua, Aditya Prabaswara, Chao Zhao, Davide Priante, Jung-Wook Min, Tien Khee Ng, and Boon S. Ooi

Photons Laboratory, Computer, Electrical, and Mathematical Sciences and Engineering (CEMSE) Division, King Abdullah University of Science and Technology (KAUST), Thuwal 23955-6900, Saudi Arabia

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We study the impact of quantum-confined stark effect (QCSE) on bias dependent micro-photoluminescence emission of the quantum disk (Q-disk) based nanowires light emitting diodes (NWs-LED) exhibiting the amber colored emission. The NWs are found to be nitrogen polar (N-polar) verified using KOH wet chemical etching and valence band spectrum analysis of high-resolution X-ray photoelectron spectroscopy. The crystal structure and quality of the NWs were investigated by high-angle annular dark field - scanning transmission electron microscopy. The LEDs were fabricated to acquire the bias dependent micro-photoluminescence spectra. We observe a redshift and a blueshift of the μPL peak in the forward and reverse bias conditions, respectively, with reference to zero bias, which is in contrast to the metal-polar InGaN well-based LEDs in the literature. Such opposite shifts of μPL peak emission observed for N-polar NWs-LEDs, in our study, are due to the change in the direction of the internal piezoelectric field. The quenching of PL intensity, under the reverse bias conditions, is ascribed to the reduction of electron-hole overlap. Furthermore, the blueshift of μPL emission with increasing excitation power reveals the suppression of QCSE resulting from the photo-generated carriers. Thereby, our study confirms the presence of QCSE for NWs-LEDs from both bias and power dependent μPL measurements. Thus, this study serves to understand the QCSE in N-polar InGaN Q-disk NWs-LEDs and other related wide-bandgap nitride nanowires, in general. © 2018 Author(s). All article content, except where otherwise noted, is licensed under a Creative Commons Attribution (CC BY) license (http://creativecommons.org/licenses/by/4.0/). https://doi.org/10.1063/1.5021290

INTRODUCTION

III-V semiconducting materials are promising candidates for electronic, piezo-electronic, and optoelectronics device applications due to their outstanding properties such as tunable direct bandgap, thermal and chemical stability, high carrier mobility, high power sustainability, and non-centrosymmetric crystal structure.1–6 Although much progress has been achieved in the growth of planar structures in the last few decades, the nanowires (NWs) technology has recently pursued as they exhibit high crystalline quality, low defect density, and large surface-to-volume ratio. The lithography-based selective area growth (SAG) and vapor liquid solid (VLS) growth schemes are time-consuming, complicated, lack of control over doping densities, and crystalline quality.7,8 To overcome such difficulties, the catalyst-free methods have been enormously employed to achieve high-density self-assembled group-III nitride NWs/Si at nitrogen-rich conditions.9–11 Though researchers achieved high-efficiency NW based light emitting devices,12,13 avoiding the strain and polarization induced internal electric fields, causing quantum-confined stark effect (QCSE) at the interface of quantum disk (Q-disk)/Q-dot and barrier, remains challenging.14,15 The internal piezoelectric fields having the magnitude of 1.2 MV/cm associated with the in-plane strain in InGaN/GaN quantum wells (QWs) which decreases the overlap of electron and hole wave functions.16,17 This spatial separation of wave-functions results from the QCSE which was first demonstrated for GaAs/AlGaAs QWs.18 However, the QCSE has been widely reported for III-V metal-polar quantum well (QW) based devices.17–22 This is least studied for N-polar nanowire-based light emitting diodes (LEDs) having the active region with the InGaN quantum disks (Q-disks) that embedded between high-quality p and n-type GaN NWs.14,23 The QCSE has deleterious effects on the device performance by reducing the rate of recombination, whereas it has less impact on cubic and non-polar wurtzite devices.24,25 In literature, QCSE was observed by employing power and bias dependent photoluminescence (PL) and electroluminescence (EL) studies on group III nitride-based devices.20–22 Several attempts were made to control the QCSE on the device performance by varying the well or dot thickness, Si doping of the QW, and cubic and non-polar epitaxy.15,25–27 Recently, Young et al. achieved complete screening of polarization-induced electric fields in c-plane (0001) InGaN/GaN QWs by employing the same n- and p-doping of 10 nm layers immediately adjacent to the QW.22 On the other hand, Wang et al. observed the suppression of QCSE for nanorod based device structures formed by a top-down approach.28 Later, the coexistence of QCSE along with localized states for InGaN/GaN NW heterostructure having the emission centered at around 2.5 eV has been demonstrated.14 Thus, in view of QCSE on NWs-LEDs, common consensus has not arrived yet.

In this study, we prepare N-polar NW-LED device structure exhibiting an amber colored emission centered at around
620 nm. The polarity of GaN NWs is determined by high-resolution x-ray photoelectron spectroscopy (HRXPS) and KOH wet etching experiments. Since the applied bias has the same order of the internal piezoelectric field (10^6 V/cm) associated with the in-plane strain in InGaN/GaN interface, the bias-dependent μPL measurements were employed. Thereby, the observed bias-dependent μPL shifts were utilized to understand the fundamental insights of the QCSE on the emission properties of N-polar NW-LEDs.

EXPERIMENTAL DETAILS

InGaN multi-quantum disks (MQDs) NWs-LED structure was grown on Si(111) substrate using a Veeco GEN 930 plasma-assisted molecular beam epitaxy (PAMBE) at 700°C. The NWs device structure consists of ten InGaN Q-disks (≈3 nm), separated by a GaN-barrier (≈7 nm) as an active region which is embedded between p-GaN (≈450 nm) and n-GaN (≈600 nm) NWs. A combination of an ion pump and a cryopump is used to achieve a base pressure of 3 × 10^{-11} Torr and oxygen partial pressure of <10^{-12} Torr, as determined by a residual gas analyzer (RGA). The cleaning procedure of Si(111) substrate can be found elsewhere.29 During growth, the nitrogen plasma source is operated with the flow rate of 1.5 standard cubic centimeter per minute (sccm) and RF-power of 350 W and group III metals are supplied by standard Knudsen cell with beam equivalent pressure (BEP) values of (In) 10 × 10^{-8} and (Ga) 7 × 10^{-8} Torr. Based on the flux rates and XRD measurements, the estimated In composition of InGaN well is ≈37%. Field Emission Scanning Electron Microscope (FESEM), by Nova Nano FEI, was employed to investigate the surface morphologies of the NWs. The etching (40% KOH for 10 min) experiments and the high-resolution x-ray photoelectron spectroscopy (HRXPS) measurements were carried out to determine the polarity of grown NWs. The details of HRXPS can be found elsewhere.30 The crystallinity was investigated by high-angle annular dark-field scanning transmission electron microscopy (HAADF-STEM) operating a probe-corrected FEI Titan at an acceleration voltage of 300 kV. Electron energy loss spectroscopy (EELS) acquisition was performed by using Gatan’s GIF Quantum of Model 966. The temperature dependent photoluminescence (TDPL) measurements were performed using 405 nm laser as an excitation source. We used the grown device structures to fabricate NW LEDs with a mesa area of 500 × 500 μm^2, and the details can be found elsewhere.31,32 However, the schematic of the device is provided in a later section of this article. The light power-current-voltage (L-I-V) characteristics and electroluminescence (EL) spectra of the NW-LEDs were acquired under direct current (DC) injection using a microscope based EL system integrated with a Keithley 2400 source meter, a Newport power meter (model 2936-C), and an Ocean Optics QE65000 spectrometer. The light output power was measured from the top of the NW-LEDs, through an optical microscope objective, using a calibrated Si photodiode connected to the optical power meter. Using Horiba Aramis, bias and power dependent μPL measurements were performed at room temperature (RT) on LEDs with an excitation line of 473 nm.

RESULTS AND DISCUSSION

To investigate the surface morphology of samples, FESEM is carried out. Figures 1(a) and 1(b) show the top and tilted views of the FESEM images obtained on the NW sample having ten pairs of InGaN MQDs as an active region. Figure 1(c) shows the density of nanowires, calculated from the plan view FESEM images, plotted as a function of nano-wire diameter. The density and an average diameter of nanowires are estimated to be ≈6 × 10^7 cm^{-2} and ≈100 nm, respectively. From the cross-sectional FESEM images, the length of the nanowires was determined to be ≈1.2 μm. We observe the tapering of the NWs which is due to the shadow effect and decrease in surface temperature that promotes lateral growth with increasing length of the NWs.33 To determine the polarity of the grown GaN NWs, we performed KOH etching test. The GaN NWs sample was kept in 40% KOH solution for 10 min. Figures 1(d) and 1(e) show the tilted view FESEM images of the pristine and KOH treated GaN NWs, respectively, displaying the flat and pyramidal tops of the NWs.34 The top pyramidal facets of GaN NWs in Fig. 1(e) confirm the N-polarity of the wires. Further, this has been verified by HRXPS valence band (VB) spectrum collected on GaN NWs. Figure 1(f) shows HRXPS valence band spectrum that acquired on GaN NWs. While collecting the valence band (VB) spectrum, we employ 0° as takeoff angle to detect the electrons emitted only from c-plane of the nanowires. The intensity ratio of the peaks P_{II} and P_{I} indicated in Fig. 1(f) is found to be 1.3 which confirms that the polarity of the NWs is nitrogen polar.35,36 These results are consistent with the reports by Hestroffer et al., where it has been demonstrated that the MBE grown GaN NWs on Si(111) substrates were prone to have N-polarity due to the unintentional formation of Si_N_x/Si before the NWs growth.37 Hence, KOH etching and HRXPS VB spectrum analyses evidence the formation of N-polar GaN NWs on Si(111).

HAADF-STEM studies were performed on the nanowire consisting of 10 pairs of InGaN Q-disk/GaN barrier to evaluate the crystalline structure, quality, and the thickness of disks and barriers. Figure 2(a) shows the low-magnification HAADF-STEM image acquired along the ⟨1120⟩ zone axis. This depicts that the NWs were vertically grown without coalescence and tapered with increasing diameter from the bottom (≈25 nm) to top (≈100 nm) while the length is ≈1.2 μm. Figure 2(b) shows the high-magnification image of the NW consisting of ten pairs of InGaN/GaN Q-disks which were sandwiched between bottom n-GaN and top p-GaN regions. Figure 2(c) is the high-resolution atomic lattice image of Q-disks and barriers in MQDs which is represented by a rectangle in Fig. 2(b). The barrier and the Q-disk were observed to be very clear with the sharp interface regions as shown in the high-resolution atomic lattice image, Fig. 2(c). The thickness of Q-disk and barrier was measured to be ≈3 and ≈7 nm, respectively. From this lattice image, any defects such as the stacking faults and misfit dislocations in the active region were not observed.38 Figure 2(d) presents the selected area electron diffraction (SAED) pattern for bright field image Fig. 2(b) taken along the ⟨1120⟩ zone axis which
infers that the single crystalline wurtzite GaN NWs and Q-disks were grown along the $h002$ direction. STEM-EELS imaging data sets were acquired to generate the elemental maps of Ga and In elements by employing Ga (1117 eV) and In (443 eV) EELS edges, respectively. The elements Ga and In across the active region were, respectively, represented by Red and Green colored pixels. Figure 2(e) depicts the elemental mapping acquired for five InGaN Q-disks showing the existence of both Ga and In atoms at the InGaN Q-disk and solely Ga at the counterparts. Figure 2(f) is the schematic illustration of a c-oriented nanowire having ten Q-disks. Thus, these results reveal that the absence of diffusion of In adatoms into the barrier layers consequently results in a sharp interface.

To inquire the internal quantum efficiency (IQE) and localized states from compositional inhomogeneity of InGaN Q-disks, TDPL measurements were performed using 405 nm laser as an excitation line. Figure 3(a) shows the TDPL measurements from which the internal quantum efficiency (IQE) is estimated to be $\approx 58\%$. This is a ratio of integrated PL intensity at 300 K and 10 K ($I_{300}/I_{10}$), assuming that the non-radiative recombination centers are frozen at 10 K. The obtained IQE value is reasonably high as expected from the STEM results described in Fig. 2. The full width at half maxima (FWHM) of RT PL spectrum is relatively higher ($\approx 400$ meV) than that of the blue LEDs, which is usual for In-rich InGaN devices. Thermal quenching of PL mainly involves two processes which are associated with the delocalization of carriers and activation of the extended defect states. The first process requires lower activation energy than
the latter. By considering these two recombination channels, the integrated intensity \( I_T \) of TDPL at a temperature \( T \) described by Arrhenius equation can be expressed as

\[
I_T = \frac{I_0}{1 + Ae^{E_1/k_BT} + Be^{E_2/k_BT}},
\]

where \( I_0 \) is the integrated intensity of the PL emission at 0 K, \( k_B \) is the Boltzmann constant, \( A \) and \( B \) are fitting constants, and \( E_1 \) and \( E_2 \) are the activation energies of the processes related, respectively, with the delocalization of carriers and thermal quenching.\(^40-42\)

Figure 3(b) describes the Arrhenius fit, using Eq. (1), to the integrated intensities of TDPL possessing the best fit resulting in the two processes with activation energies \( E_1 \) and \( E_2 \) of 1.4 and 64 meV, respectively. The lowest activation energy (\( E_1 \)) depicts the weak localization effect. Moreover, as shown in the inset of Fig. 3(b), Varshni fit is carried out to the TDPL peak energies using the following equation:

\[
E_T^g = E_0^g - \frac{\alpha T^2}{T + \beta} - \frac{\sigma^2}{k_BT},
\]

where \( E_T^g \) and \( E_0^g \) are the PL emission energies at temperature \( T \) and 0 K, and \( \alpha \) and \( \beta \) are the Varshni parameters with the fitting values of \( 4 \times 10^{-3} \) eV/K and 550 K, respectively. \( \sigma \) indicates the degree of localization effects and the more significant value of \( \sigma \) corresponds to the stronger localization effect. In this case, fitted \( \sigma \) value is as low as 0.1 meV which is far less than the literature values\(^43-45\) indicating the absence of localized states in Q-disks. Consequently, there is no S shape behavior found in the inset of Fig. 3(b).\(^31,39\) This observation is in good agreement with the obtained extremely low activation energy \( E_1 \) from the Arrhenius fit due to the delocalization of the carriers. As a result, Arrhenius and Varshni analyses of the TDPL data reveal that the formed Q-disks are homogeneous in the composition which is in good agreement with the HAADF-STEM results shown in Figs. 2(b), 2(c), and 2(e). Thus, this confirms the observed PL peak shift in the later section (Fig. 5) solely related to the resultant internal fields.

To investigate the performance of the NW InGaN Q-disk LEDs, L-I-V and EL characterizations were carried out under dc bias conditions. Figure 4(a) illustrates the schematic diagram of fabricated NWs Q-disk LED device. Figure 4(b) describes the light output power-current-voltage (L-I-V) characteristics of an LED with a mesa area of 500 \( \times \) 500 \( \mu \)m\(^2\) measured under dc bias, exhibiting constantly increasing output power. The output power of the device at 15 mA injection current is 40 \( \mu \)W with external quantum efficiency (EQE) of \( \approx 0.2\% \). The inset shows microscope image of LED having the onset voltage of \( \approx 2.5 \) V in the forward bias (FB). The high onset voltage is due to the unintentional formation of the silicon nitride (\( \text{Si}_3\text{N}_4 \)) intermediate dielectric layer at the interface of Si and GaN NWs.\(^11\) Figure 4(c) shows the stack of normalized EL emission peaks of the device with increasing injected forward current. Any defect related electroluminescence of devices was not observed in the reverse bias (RB). Figure 4(d) presents the EL peak emission values and integrated intensity with increasing the forward current. The insets of Fig. 4(d) depict the variation in the color of LED emission with increasing the forward current, which is in accordance with the EL shifts in Fig. 4(c). The charge-coupled device (CCD) images represent the illuminated LEDs at respective injected currents. It has been observed that EL peak has a blueshift of \( \approx 25 \) nm with raising integrated intensity as increasing the forward current. The observed blueshift of the EL emission even at low injection currents is due to the two-dimensional character of the short-range Coulomb carrier screening between the interacting carriers within the InGaN Q-disks.\(^46-48\)
In order to understand the role of the internal piezoelectric field induced QCSE on PL peak position of InGaN Q-disks embedded in N-polar GaN NWs, one must determine the direction of the electric field which is related with the strain induced polarization. From the elastic theory of wurtzite materials, the relation between the strain and the piezoelectric polarization expressed in the following is used to determine the direction of polarization:

$$P_c = \frac{2d_{31}}{C_{11} + C_{12} - \frac{2C_{13}^2}{C_{33}}} e_{\alpha} \left[\text{InGaN}/\text{GaN}\right],$$

(3)

where $P_c$ is the polarization along $c$ direction, $d_{31}$ is piezoelectric modulus, $C_{11}, C_{12}, C_{13},$ and $C_{33}$ are the elastic constants, and $e_{\alpha}$ is the in-plane compressive strain ($<0$). Hence, for InGaN (Q-disk)/GaN (barrier) system, strain induced polarization is positive ($P_c > 0$) pointing along [0001]. As a consequence, the direction of the piezoelectric field in the N-polar devices directs along [0001].

Thereby, for our device, the direction of internal electric field induced by the piezoelectric polarization points from the substrate to surface. Figure 5(a) is an equivalent circuit of NWs device where the LED symbol represents single NW LED, while the resistance in series to the NW results from both the individual NW device and unintentionally formed thin Si$_x$N$_y$ layer. Hence, the LED can be treated as an ensemble of several NW-LEDs situated in parallel. Thereby, for instance, when the applied bias is 3 V, each Q-disk experiences the field of $\approx 10^6$ V/cm which is as similar value as the strain-induced internal piezoelectric field for InGaN/GaN system. Therefore, bias dependent PL studies can be utilized to explore the additional important insights of the QCSE of N-polar devices. The external field was employed by sweeping the voltage from negative (−4 V) to forward (5 V) bias. The forward bias was applied solely up to 5 V slightly higher than the onset voltage (2.5 V) of the device to avoid the effect of LED emission on the $\mu$PL.
peak wavelength. As shown in Fig. 5(b), under forward bias, the LED shows the enhanced µPL intensity with a significant blueshift from PL spectrum acquired at zero bias. Further increase of forward bias (>3 V) resulted in no significant blueshift which is due to the dominant electroluminescent emission of LED. On the other hand, PL intensity drastically reduces with increasing reverse bias with a substantial redshift in the peak position with respect to the unbiased peak. Figure 5(c) shows the values of photocurrent (Y-axis) and redshift in PL peak emission (alternate Y-axis) with increasing the reverse bias. The PL intensity given in Fig. 5(b) decreases with reverse bias in contrast to the photocurrent in Fig. 5(c). Thereby, the quenching of PL intensity, under the reverse bias conditions, is solely due to the reduction of electron-hole overlap which lowers the IQE, consequently, the PL intensity. Thus, this is not a consequence of the sweep out of charge carriers from the InGaN Q-disks. The PL redshift in the reverse bias is due to the enhancement of QCSE which is an immediate consequence of the superposition of both internal piezo and external electric fields in the same direction. The variation of PL intensity and shifts are in accordance with the directions represented in Fig. 5(d) for N-polar wurtzite crystals. Figure 5(e) shows the schematic band diagram of single Q-disk with biased and unbiased conditions. Thus, forward bias reduces the resultant field in the Q-disk, while reverse bias increases it. As a consequence, bias dependent PL shifts for N-polar Q-disks in this study are observed to be reversed in comparison with the reported PL shifts of metal polar QWs, which can be ascribed to the reverse direction of the piezoelectric field for the N-polarity of NW-LEDs emitting around 620 nm. Hence, this study provides fundamental insights in determining the polarity of the devices.

Moreover, to understand the recombination processes involved in µPL spectra of the device, the dependency of the excitation power on the PL peak emission and integrated intensity were studied. Figure 6 and the corresponding inset show the excitation power (P) dependence on the PL intensity (I). This is commonly studied using a power law: $I \propto P^\alpha$, where the exponent ($\alpha$) depicts the type of carrier recombination. The observed linear dependence ($\alpha \approx 1$), shown in the respective inset, indicates the weak localization effect. This is in agreement with the TDPL analysis of Fig. 3. Furthermore, screening of the internal electric field in InGaN/GaN Q-disks is explained by the photo-generated carriers which occupy degenerate Q-disk states thereby results in the overlap of the electron and hole wave functions. This results in a screening of QCSE; consequently, a blueshift of the
Figure 6. The power dependent µPL with an inset depicting the fit using a power law to the µPL integrated intensity.

characteristic PL signal with increasing excitation power is observed.\(^3\)

CONCLUSIONS

In summary, the NWs device structure having ten InGaN Q-disks as an active region was grown on Si(111) substrates using PAMBE. The polarity of GaN NWs is shown to be N-polar using KOH etching test and HRXPS valence band spectrum. These NWs-LED structures are of high quality and dislocation-free material which show the absence of localized defect states in InGaN QNWs also confirmed by the TDPL and power dependent PL measurements. Moreover, for N-polar NWs devices, the PL integrated intensity.

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\(^{1}\)S. Nakamura, Science 281, 956 (1998).

\(^{2}\)C. Tan, D. Borovac, W. Sun, and N. Tansu, Sci. Rep. 6, 19271 (2016).

\(^{3}\)N. J. Ku, J. H. Huang, C. H. Wang, H. C. Fang, and C. P. Liu, Nano Lett. 12, 562 (2012).

\(^{4}\)Y. Sun, K. Zhou, Q. Sun, J. Liu, M. Feng, Z. Li, Y. Zhou, L. Zhang, D. Li, S. Zhang, M. Ikeda, S. Liu, and H. Yang, Nat. Photonics 10, 595 (2016).

\(^{5}\)K. Tomioka, M. Yoshimura, and T. Fukui, Nature 488, 189 (2012).

\(^{6}\)M. Tangi, A. De, and S. M. Shivaaprasad, J. Appl. Phys. 123, 15701 (2018).

\(^{7}\)S. H. Oh, K. Van Benthem, S. I. Molina, A. Y. Borisevich, W. Luo, P. Werner, N. D. Zakharov, D. Kumar, S. T. Pantelides, and S. J. Pennycook, Nano Lett. 8, 1016 (2008).

\(^{8}\)K. Kishino and S. Ishizawa, Nanotechnology 26, 225602 (2015).

\(^{9}\)R. Calarco, R. J. Meijers, R. K. Debnath, T. Stoica, E. Sutter, and H. Lüth, Nano Lett. 7, 2248 (2007).

\(^{10}\)P. Kumar, M. Tuteja, M. Kesaria, U. V. Waghmare, and S. M. Shivaaprasad, Appl. Phys. Lett. 101, 131605 (2012).

\(^{11}\)K. Hestroffer, C. Leclere, V. Cantelli, C. Bougerol, H. Renevier, and B. Daudin, Appl. Phys. Lett. 100, 212107 (2012).

\(^{12}\)M. R. Philip, D. D. Choudhary, M. Djavid, K. Q. Le, J. Piao, and H. P. T. Nguyen, J. Sci.: Adv. Mater. Devices 2, 150 (2017).

\(^{13}\)R. Wang, X. Liu, I. Shih, and Z. Mi, Appl. Phys. Lett. 106, 261104 (2015).

\(^{14}\)J. Lahennemann, O. Brandt, C. Pfüller, T. Flissikowski, U. Jahn, E. Luna, M. Hanke, M. Knelangen, A. Trampert, and H. T. Grahn, Phys. Rev. B: Condens. Matter Mater. Phys. 84, 155303 (2011).

\(^{15}\)J. Renard, R. Songmuang, G. Tourbot, C. Bougerol, B. Daudin, and B. Gayral, Phys. Rev. B: Condens. Matter Mater. Phys. 80, 121305(R) (2009).

\(^{16}\)T. Takeuchi, C. Wetzel, S. Yamaguchi, H. Sakai, I. Akasaki, Y. Yankeo, S. Nakagawa, Y. Yamaoka, and N. Yamada, Appl. Phys. Lett. 73, 1691 (1998).

\(^{17}\)J. H. Ryoo, P. D. Yoder, J. Liu, Z. Lochner, H. S. Kim, S. Choi, H. J. Kim, and R. D. Dupuis, IEEE J. Sel. Top. Quantum Electron. 15, 1080 (2009).

\(^{18}\)D. A. B. Miller, D. S. Chemla, T. C. Damen, A. C. Gossard, W. Wiegmann, T. H. Wood, and C. A. Burrus, Appl. Phys. Lett. 53, 2173 (1984).

\(^{19}\)M. Leroux, N. Grandjean, M. Lautig, J. Massies, B. Gil, P. Lefebvre, and P. Bigenwald, Phys. Rev. B 58, R13371 (1998).

\(^{20}\)T. Takeuchi, S. Sota, M. Katsuragawa, M. Komori, H. Takeuchi, H. Amano, and I. Akasaki, Jpn. J. Appl. Phys. Part 2 36, L382 (1997).

\(^{21}\)H. Masui, J. Sonoda, N. Prafß, I. Koslow, S. Nakamura, and S. P. DenBaars, J. Phys. D: Appl. Phys. 41, 165105 (2008).

\(^{22}\)N. G. Young, R. M. Farrell, S. Oh, M. Cantore, F. Wu, S. Nakamura, S. P. DenBaars, C. Weisbuch, and J. S. Speck, Appl. Phys. Lett. 108, 061105 (2011).

\(^{23}\)J. Lahennemann, P. Cordfir, F. Feix, J. Camimura, T. Flissikowski, H. T. Grahn, L. Geelhaar, and O. Brandt, Nano Lett. 16, 917 (2016).

\(^{24}\)Y. H. Ra, R. Navamathavan, H. Il Yoo, and C. R. Lee, Nano Lett. 14, 1537 (2014).

\(^{25}\)S.-H. Park and S.-L. Chuang, Phys. Rev. B 59, 4725 (1999).

\(^{26}\)T. Deguchi, A. Shikanai, K. Torii, T. Sota, S. Chichibu, and S. Nakamura, Appl. Phys. Lett. 72, 3329 (1998).

\(^{27}\)J. Bai, T. Wang, and S. Sakai, J. Appl. Phys. 88, 4729 (2000).

\(^{28}\)C.-Y. Wang, L.-Y. Chen, C.-P. Chen, Y.-W. Cheng, M.-Y. Ke, M.-Y. Hsieh, H.-M. Wu, L.-H. Peng, and J. Huang, Opt. Express 16, 10549 (2008).

\(^{29}\)M. Tangi, P. Mishra, B. Janjua, T. K. Ng, D. H. Anjum, A. Prabaswara, Y. Yang, A. M. Albadri, A. Y. Alyamani, M. M. El-Dessouki, and B. S. Ooi, J. Appl. Phys. 120, 45701 (2016).

\(^{30}\)M. Tangi, P. Mishra, M.-Y. Li, M. K. Shafaka, D. H. Anjum, M. N. Hedhili, T. K. Ng. L.-J. Li, and B. S. Ooi, Appl. Phys. Lett. 111, 92104 (2017).

\(^{31}\)C. Zhao, T. K. Ng, R. T. Elafandy, A. Prabaswara, G. B. Consiglio, I. A. Ajia, I. S. Roqan, B. Janjua, C. Shen, J. Eid, A. Y. Alyamani, M. M. El-Dessouki, and B. S. Ooi, Nano Lett. 16, 4616 (2016).

\(^{32}\)J. Zhao, T. K. Ng, N. Wei, A. Prabaswara, M. S. Alias, B. Janjua, C. Shen, and B. S. Ooi, Nano Lett. 16, 1056 (2016).

\(^{33}\)K. Hestroffer and B. Daudin, J. Appl. Phys. 114, 244305 (2013).
34 N. Jamond, P. Chrétien, F. Houzé, L. Lu, L. Largeau, O. Maugain, L. Travers, J. C. Harmand, F. Glas, E. Lefevre, M. Thernycheva, and N. Gogneau, Nanotechnology 27, 325403 (2016).
35 T. D. Veal, P. D. C. King, P. H. Jefferson, L. F. J. Piper, C. F. McConville, H. Lu, W. J. Schaff, P. A. Anderson, S. M. Durbin, D. Muto, H. Naoi, and Y. Nanishi, Phys. Rev. B 76, 75313 (2007).
36 D. Skuridina, D. V. Dinh, B. Lacroix, P. Ruterana, M. Hoffmann, Z. Sitar, M. Pristovsek, M. Kneissl, and P. Vogt, J. Appl. Phys. 114, 173503 (2013).
37 K. Hestroffer, C. Leclere, C. Bougerol, H. Renevier, and B. Daudin, Phys. Rev. B 84, 245302 (2011).
38 M. Tangi, A. De, J. Ghatak, and S. M. Shivaprasad, J. Appl. Phys. 119, 205701 (2016).
39 S. Deshpande, J. Heo, A. Das, and P. Bhattacharya, Nat. Commun. 4, 1675 (2013).
40 M. Leroux, N. Grandjean, B. Beaumont, G. Nataf, F. Semond, J. Massies, and P. Gibart, J. Appl. Phys. 86, 3721 (1999).
41 E. Monroy, N. Gogneau, F. Enjalbert, F. Fossard, D. Jalabert, E. Bellet-Amalric, L. S. Dang, and B. Daudin, J. Appl. Phys. 94, 3121 (2003).
42 J. Granda1, J. Pereiro, A. Bengoechea-Encabo, S. Fernandez-Garrido, M. A. Sanchez-Garcia, E. Munoz, E. Calleja, E. Luna, and A. Trampert, Appl. Phys. Lett. 98, 61901 (2011).
43 M. Funato, Y. S. Kim, T. Hira, A. Kaneta, Y. Kawakami, T. Miyoshi, and S. I. Nagahama, Appl. Phys. Express 6, 111002 (2013).
44 T. Lu, Z. Ma, C. Du, Y. Fang, H. Wu, Y. Jiang, L. Wang, L. Dai, H. Jia, W. Liu, and H. Chen, Sci. Rep. 4, 6131 (2014).
45 Y. P. Varshni, Physica 34, 149 (1967).
46 F. Della Sala, A. Di Carlo, P. Lugli, F. Bernardino, V. Fiorentini, R. Scholz, and J.-M. Jancu, Appl. Phys. Lett. 74, 2002 (1999).
47 J.-Y. Bigot, M. T. Portella, R. W. Schoelein, J. E. Cunningham, and C. V. Shank, Phys. Rev. Lett. 67, 636 (1991).
48 C. Lu, L. Wang, J. Lu, R. Li, L. Liu, D. Li, N. Liu, and L. Li, J. Appl. Phys. 113, 013102 (2013).
49 A. L. Bavencove, G. Tourbot, J. Garcia, Y. Désières, P. Gilet, F. Levy, B. André, B. Gayral, B. Daudin, and L. S. Dang, Nanotechnology 22, 345705 (2011).
50 F. Limbach, C. Hausswald, J. Lähnemann, M. Wölz, O. Brandt, A. Trampert, M. Hanke, U. Jahn, R. Calarco, L. Geelhaar, and H. Riechert, Nanotechnology 23, 465301 (2012).
51 T. Schmidt, K. Lischka, and W. Zulehner, Phys. Rev. B 45, 8989 (1992).
52 M. Tangi, M. K. Shakfa, P. Mishra, M.-Y. Li, M.-H. Chiu, T. K. Ng, L. Li, and B. S. Ooi, Opt. Mater. Express 7, 3697 (2017).
53 H. Schöning, S. Halm, A. Forchel, G. Bacher, J. Off, and F. Scholz, Phys. Rev. Lett. 92, 106802 (2004).