InGaAs/GaAsP strain balanced multi-quantum wires grown on misoriented GaAs substrates for high efficiency solar cells

D. Alonso-Álvarez, T. Thomas, M. Führer, N. P. Hylton, N. J. Ekins-Daukes, D. Lackner, S. P. Philipps, A. W. Bett, H. Sodabanlu, H. Fujii, K. Watanabe, M. Sugiyama, L. Nasi, and M. Campanini

Citation: *Applied Physics Letters* **105**, 083124 (2014); doi: 10.1063/1.4894424

View online: <http://dx.doi.org/10.1063/1.4894424>

View Table of Contents: <http://scitation.aip.org/content/aip/journal/apl/105/8?ver=pdfcov>

Published by the AIP Publishing

Articles you may be interested in

[Optical characteristics of GaAs nanowire solar cells](#)

J. Appl. Phys. **112**, 104311 (2012); 10.1063/1.4764927

[Structural atomic-scale analysis of GaAs/AlGaAs quantum wires and quantum dots grown by droplet epitaxy on a \(311\)A substrate](#)

Appl. Phys. Lett. **98**, 193112 (2011); 10.1063/1.3589965

[Self-catalyzed GaAs nanowire growth on Si-treated GaAs\(100\) substrates](#)

J. Appl. Phys. **109**, 094306 (2011); 10.1063/1.3579449

[Transport and strain relaxation in wurtzite InAs–GaAs core-shell heterowires](#)

Appl. Phys. Lett. **98**, 152103 (2011); 10.1063/1.3579251

[Stacked GaAs multi-quantum wires grown on vicinal GaAs\(110\) surfaces by molecular beam epitaxy](#)

Appl. Phys. Lett. **72**, 465 (1998); 10.1063/1.120787

A promotional banner for the 2014 Special Topics in APL Materials. The banner has an orange background with a white border. In the center, the text '2014 Special Topics' is written in a large, white, sans-serif font. Below this text, there are five circular icons, each representing a different material category: PEROVSKITES (red and black geometric shapes), 2D MATERIALS (blue and red hexagonal pattern), MESOPOROUS MATERIALS (green and yellow molecular structure), BIOMATERIALS/BIOELECTRONICS (yellow and black abstract pattern), and METAL-ORGANIC FRAMEWORK MATERIALS (brown and black abstract pattern). At the bottom left, the AIP logo is followed by the text 'APL Materials'. At the bottom right, a red banner with white text says 'Submit Today!'.

InGaAs/GaAsP strain balanced multi-quantum wires grown on misoriented GaAs substrates for high efficiency solar cells

D. Alonso-Álvarez,¹ T. Thomas,¹ M. Führer,¹ N. P. Hylton,¹ N. J. Ekins-Daukes,¹
 D. Lackner,² S. P. Philipps,² A. W. Bett,² H. Sodabanlu,³ H. Fujii,³ K. Watanabe,³
 M. Sugiyama,³ L. Nasi,⁴ and M. Campanini⁴

¹Department of Physics, Imperial College, London SW7 2BZ, United Kingdom

²Fraunhofer Institute for Solar Energy Systems ISE, Heidenhofstrasse 2, 79110 Freiburg, Germany

³Department of Electrical Engineering and Information Systems, The University of Tokyo, Tokyo, Japan

⁴CNR-IMEM Sezione di Parma, Parco Area delle Scienze 37/A, 43010 Fontanini-Parma, Italy

(Received 23 May 2014; accepted 20 August 2014; published online 29 August 2014)

Quantum wires (QWRs) form naturally when growing strain balanced InGaAs/GaAsP multi-quantum wells (MQW) on GaAs [100] 6° misoriented substrates under the usual growth conditions. The presence of wires instead of wells could have several unexpected consequences for the performance of the MQW solar cells, both positive and negative, that need to be assessed to achieve high conversion efficiencies. In this letter, we study QWR properties from the point of view of their performance as solar cells by means of transmission electron microscopy, time resolved photoluminescence and external quantum efficiency (EQE) using polarised light. We find that these QWRs have longer lifetimes than nominally identical QWs grown on exact [100] GaAs substrates, of up to 1 μ s, at any level of illumination. We attribute this effect to an asymmetric carrier escape from the nanostructures leading to a strong 1D-photo-charging, keeping electrons confined along the wire and holes in the barriers. In principle, these extended lifetimes could be exploited to enhance carrier collection and reduce dark current losses. Light absorption by these QWRs is 1.6 times weaker than QWs, as revealed by EQE measurements, which emphasises the need for more layers of nanostructures or the use light trapping techniques. Contrary to what we expected, QWR show very low absorption anisotropy, only 3.5%, which was the main drawback *a priori* of this nanostructure. We attribute this to a reduced lateral confinement inside the wires. These results encourage further study and optimization of QWRs for high efficiency solar cells. © 2014 AIP Publishing LLC.

[<http://dx.doi.org/10.1063/1.4894424>]

Strain balanced multi-quantum wells (MQW) are an established method for tailoring the absorption edge of III-V solar cells. Research over the last couple of years has addressed the performance as a function of the position of the QWs inside the device,^{1,2} their number,^{3,4} their composition to keep them strain-balanced,⁵ and the effect of background doping on carrier extraction.⁶ As a result of these efforts, InGaP/MQW/Ge quantum well devices have attained efficiencies in excess of 40% under concentration⁵ and over 30% under air-mass 0 solar spectrum (AM0).⁷

These achievements have required a mastery of the fabrication of strain balanced MQW structures on highly misoriented substrates (typically [100] 6° towards [111]B). However, growing on misoriented substrates brings an additional difficulty—as well as a possible degree of freedom—since it is well known that there is usually strong lateral thickness modulation in the QWs leading to the formation of quantum wires (QWRs). This effect has been observed in a number of material systems including lattice matched GaAs/AlGaAs and strained InGaAs/GaAsP QWs, both grown on GaAs substrates.^{8,9}

In principle, QWRs have an important undesirable property from the point of view of their exploitation as solar cells: They have a polarised, out-of-plane light emission, and absorption. This will effectively reduce their absorption coefficient given that part of the un-polarised sunlight will not be absorbed by the nanostructures. QWRs might have

also some advantages over QWs. For example, in the case of InGaAs-based nanostructures, they can tailor the absorption edge of the host material to lower energies than a QW of similar total indium content. This is important from the point of view of the structure fabrication since they will need less aggressive strain balancing and, hence, lower accumulated strain energy, responsible for the formation of dislocations.

We report high quality InGaAs/GaAsP multi-quantum wire stacks on misoriented GaAs substrates from the point of view of the performance of the solar cell. The aim is to determine if the presence of naturally occurring QWRs represent an advantage or a disadvantage in order to achieve high conversion efficiencies and, in the first case, how it can be exploited.

All samples studied were based on 40 QW/barrier pairs made of strain balanced Ga_{0.80}In_{0.20}As/GaAs_{0.85}P_{0.15}. This stack is sandwiched between 150 nm thick Ga_{0.49}In_{0.51}P barriers, to remove any influence in the carrier dynamics from the sample surface or the substrate. Nominal QW thickness is 8 nm and barrier thickness is 17 nm.

Sample labelled “QW” was grown on exact [100] GaAs substrate at the University of Tokyo, whereas sample “QWR” was grown on GaAs [100] 6° misoriented towards [111]B at the Fraunhofer ISE. Each sample is the best one of a series, where the growth conditions were optimized in order to achieve the highest photoluminescence (PL) intensity. The QWR sample includes a 1 nm thick GaAs interlayer

just before and after the GaAsP barrier, which reduces the In and P intermixing between QWR layer and the barrier and was found to improve the PL intensity. In addition to the heterostructure, the QWRs stack (excluding the external GaInP barriers) was also introduced into an n-i-p solar cell to enable the study of its electrical properties.

All samples were grown by metal-organic chemical vapour deposition (MOCVD). The QWRs samples were grown on an Aixtron 2800 G4-TM reactor at a pressure of 50 mbar. The precursors were trimethylgallium, trimethylindium, arsine, phosphine, dimethylzinc, and silane. The strain balanced nanostructures were deposited at 580 °C with a growth rate of 0.7 nm/s. The V/III ratio was 20 (50) for the GaInAs (GaAsP) growth. The QW samples were grown on an Aixtron 2000 reactor at a pressure of 100 mbar. The precursors were trimethylgallium, trimethylindium, tertiarybutylarsine, tertiarybutylphosphine, dimethylzinc, and silane. The strain balanced nanostructures were deposited at 610 °C with a growth rate of 9 nm/s. The V/III ratio was 12/30 for the GaInAs (GaAsP) growth.

Structural characterization of both samples was performed by means of a TEM/STEM JEOL 2200FS system. Figure 1 shows the high angle annular dark field scanning electron microscope images (HAADF-STEM) of the two samples. In the QW sample (Figure 1(a)), we can see a homogeneous MQW structure with good quality, abrupt interfaces and absence of lateral composition modulation. The measured thicknesses of the layers agree with the nominal

ones within experimental uncertainty. The QWR sample (Figure 1(b)), on the contrary, shows a strong lateral thickness modulation which evolves with the number of layers, leading to well developed QWRs with triangular section from the 7th layer.

A detail of the QWRs around the middle of the stack and a schematic representation of them are shown in Figures 1(c) and 1(d), respectively. The QWRs are vertically aligned in columns tilted around 40° from the growth direction, although it depends on the position along the stack. The in-plane periodicity of the QWRs is ~121 nm, and the out-of-plane is ~28.8 nm. The total height of the QWR is ~13.6 nm, and the minimum thickness of the layer, at the boundary between one QWR and the next one, is ~4.6 nm. The wide top facet of the QWR forms an angle of ~5.5° with respect to the sample surface, very close to the 6° substrate misorientation, suggesting that it is a [100] plane. No obvious presence of the GaAs interlayer can be observed, pointing to a complete intermixing with the other layers.

It is worth highlighting that after the growth of each QWR layer, the surface flatness recovers, as deduced from the lower, flat interface at each layer. This is in contrast with the QWR structures grown previously on misoriented substrates, where the QWRs nucleate at the edges of multi atomic steps and have a more parallelepiped cross-section.^{8,9} It is, however, compatible with the findings of Xie *et al.*, where compressive strain tends to make the surface rougher and a tensile strain flattens it.¹⁰ In this context, although the QWR formation in the lower layers and its shape can be associated with the substrate misorientation, the subsequent formation and the arrangement in tilted columns is more likely associated with the strain fields created by the buried nanostructures.¹¹

We have analysed the optical properties of these samples by means of PL and time-resolved photoluminescence (TRPL). As excitation source, we use either a continuous wave (CW) laser (Ti:Sapphire crystal pumped by a neodymium yttrium vanadate (Nd:YVO4) laser, Spectra Physics) or a pulsed laser diode (PicoQuant), in both cases emitting at 780 nm. The duration of the laser pulse is 0.5 ns and its repetition rate is 10 MHz. PL is collected using a 0.5 m focal length spectrometer and an InP/InGaAsP photomultiplier tube (PMT) for the detection (H12694-45, Hamamatsu). For CW measurements, we use a mechanical light modulator and a lock-in amplifier, whereas for time resolved measurements, the signal of the PMT is sent to a correlation card (Becker & Hickl).

Figure 2(a) shows the CW-PL of both samples at room temperature. The QW sample emits at 1.252 eV with a full width at half maximum of 19 meV. QWRs emit at lower energies, 1.214 eV and the emission band is much broader, 45 meV. The width of this band is attributed to the size evolution of the QWRs along the stack, as revealed by TEM (Figure 1(b)). At low power (<2 W/cm²), QW sample has significantly lower integrated PL than the QWR one, but this difference disappears with increasing power. From 13 W/cm² onwards, the integrated PL of both QW and QWR samples is the same and increases linearly with power, suggesting that radiative recombination is the dominant recombination process (see inset in Figure 2(a)). Increasing the excitation power

FIG. 1. HAADF-STEM images of the (a) QW and (b) QWR samples. (c) Detail of three QWR layers around the middle of the stack. (d) Schematic representation of the QWRs, indicating the dimensions and important angles.

FIG. 2. (a) CW-PL emission of the QW and QWR samples at room temperature. The inset shows the evolution of the integrated PL intensity with power. (b) Decay times as a function of average excitation power density obtained from an exponential fit of the initial PL decay. Dashed (continuous) line is a fit to Eq. (3) for the QW (QWR) data. Inset shows a scheme of the QW/barrier band structure in the dark (dashed line) and the bending (continuous line) due to negative charge accumulation inside the nanostructure.

does not change the peak positions or their widths, confirming that band filling effects are unimportant in this power range and temperature.

The PL transients (not shown) are very long compared to those typically found in type-I QW structures and show a complex evolution with excitation power, changing from a monoexponential decay at low powers to a slower one at higher powers, typical of bimolecular decay mechanisms. Figure 2(b) shows the decay time τ_D we obtain by fitting an exponential dependence to the initial portion of the TRPL curve, just after the absorption of the laser pulse and the thermalization of carriers. This fit is not intended to reproduce the full curve but just to reveal the dominant decay mechanism during the first few nanoseconds after the absorption of the laser pulse. In both samples, τ_D increases with increasing power density, reaches a maximum value and then decreases at the highest powers. The maximum values are $1 \mu\text{s}$ at 0.7 W/cm^2 for the QWRs and $0.1 \mu\text{s}$ at 4 W/cm^2 for the QWs. These numbers are much longer than those typically obtained in type-I QWs and QWR, usually in the order of nanoseconds or shorter.^{12,13}

It is often found that QW stacks show longer lifetimes than nominally identical single QWs. Although such behaviour has been associated with an increase in composition

fluctuation between layers and local disorder,¹⁴ we believe that the underlying mechanism is in fact related to an asymmetric escape rate of electrons and holes from the nanostructures.^{15,16} Such asymmetry would lead to nanostructures charged with one type of carriers and barriers charged with the other type. This charge separation will produce an electrostatic barrier between the QWs (QWRs) and the surroundings (see inset in Figure 2(b)), decreasing the escape rate of the type of carriers that escape more easily and, likewise, also the capture rate of the other type of carriers. In samples with a single layer of nanostructures or in the presence of a built-in electric field sweeping away the unconfined carriers, such an effect will be probably small, but in a good quality MQW the effect can be significant. Increasing the carrier density (higher excitation power) will increase the height of the barrier and reduce the radiative recombination time, allowing the latter to become a more efficient decay mechanism, in agreement with the observations of CW-PL (inset of Figure 2(a)). Assuming a capacitor-like dependence of the electrostatic barrier ΔE with the accumulated charge and this proportional to the excitation power P_{AV} , we can write the non-radiative time as

$$\tau_{NR} = \tau_0 \exp\left(\frac{E_a}{k_b T}\right) \exp\left(\frac{\Delta E}{k_b T}\right) = c_1 \exp(c_2 P_{AV}), \quad (1)$$

with E_a is the activation energy in the dark, τ_0 a characteristic time, k_b is the Boltzmann constant, and T is the temperature. $c_1 = \tau_0 \exp(E_a/k_b T)$ is the decay time in the dark—combining thermal escape, re-capture, non-radiative recombination, and radiative recombination in the barrier—and c_2 is a constant that depends on the exact carrier distribution in and around the nanostructures. On the other hand, the initial radiative recombination time can be written as a function of the radiative recombination coefficient B and the carrier density inside the nanostructure N_0 , as

$$\tau_R = (BN_0)^{-1} = (c_3 P_{AV})^{-1}. \quad (2)$$

Combining both, we have for the initial decay time τ_D

$$\tau_D^{-1} = \tau_{NR}^{-1} + \tau_R^{-1}. \quad (3)$$

Figure 2(b) shows a fitting of the experimental data to Eq. (3). The decay times in the dark, given by c_1 , are 3 and 6 ns for the QWs and the QWRs, respectively. c_2 is 9 times larger for the QWR sample, indicating a more efficient carrier accumulation process in these nanostructures, whereas c_3 is around 1.6 times larger for the QWs. This last number indicates that radiative recombination—and therefore light absorption—is more efficient in QWs than in QWRs, something we had anticipated as the possible drawback of QWRs.

From these PL experiments, we can derive several interesting conclusions. The first one is that photogenerated carriers survive for longer time in QWRs than in QWs at any carrier density. This can be seen as a potential for lower recombination losses, and therefore lower dark currents, in QWRs solar cells compared to QWs. The charge separation has a second interesting effect. In MOCVD solar cells, there is often a high level of background carbon impurities that leads to an undesirable light p-doping in the intrinsic region and reduces the performance of the devices. Adding an

n-type doping has been shown to mitigate this effect and improve the solar cell properties.⁶ In InGaAs/GaAsP QWRs and QWs, carriers that escape more easily are holes, due to their lower confinement than electrons. This means that when they are swept away by the electric field, the electrons left behind can partly compensate the unwanted background doping, leading to a similar improvement than with a compensating doping.

External quantum efficiency (EQE) measurements are useful to assess the combined effect of light absorption and carrier collection in a solar cell as a function of wavelength. Figure 3 shows the EQE (with bias light = 0.3 Suns) and the main electrical parameters deduced from current-voltage (IV) measurements at 1 sun of illumination under short circuit conditions. The device shows an extended absorption edge up to 1010 nm, making use of the extra photons available at the band between 950 and 1120 nm. There is a remarkable Stokes shift of 14 meV between the emission peak at 1021 nm (1.214 eV, Figure 2(a)) and the absorption, evidencing carrier relaxation into deep localised states arising from variation in the size of the QWR. EQE measurements at high reverse bias show no variation in the EQE signal at any wavelength, indicating good carrier extraction and transport at the experimental illumination conditions. The EQE at the band edge is relatively low, only 19%, meaning that most photons will cross the structure without being absorbed by the QWRs. The dimensionless absorption coefficient per QWR layer at the edge, $\alpha_{\text{Layer}} = -\log(1 - \text{EQE}_{\text{edge}})/N_{\text{Layers}}$, is 0.0055. This is around 40% less than the typical values observed for the InGaAs/GaAsP system (~ 0.0093 , excluding excitonic absorption¹⁷), which is in good agreement with the lower radiative recombination efficiency for QWRs compared to QWs discussed before. The light IV data show higher short circuit current than typical GaAs solar cells (normally below 30 mA/cm²), very good open circuit voltage, taking into account the position of the absorption edge, and a reasonable, although improvable, fill factor suggesting that dark current is not very large.

Evidence of QWR formation is assessed in the EQE experiments by using polarised monochromatic light (inset Figure 3),

FIG. 3. EQE of a solar cell containing the QWR structure. Dashed lines indicate the absorption edge of the bulk GaAs and the QWRs. Inset: Detail of the EQE in the QWR region for two perpendicular polarizations. The dots represent the polarization degree of the EQE.

thus revealing the anisotropy of light absorption. The polarization degree we find at the absorption edge (1010 nm) is of 3.5%, quite low compared with other QWR structures ($>80\%$),¹⁸ suggesting that the lateral confinement is relatively weak even at the ground state and the QWR character of these nanostructures is not strong. The polarization degree increases at longer wavelengths and goes down to zero at the shorter ones, consistent with more localised ground states in QWRs of different sizes and the presence of QW-like excited energy levels in the nanostructures.

These results are encouraging since they show that polarization losses are not detrimental, the performance of QWRs is comparable to that of QWs and carrier collection seems to be reasonable. Including more QWR layers and optimizing the growth conditions in order to increase the homogeneity and quality of the nanostructures could potentially lead to a performance improvement over QWs. However, impact on the transport properties of the enhanced 1D-photo-charging observed around the QWRs is yet to be fully analysed in order to have a complete picture of the QWRs performance in solar cells. In particular, this photo-charging of the nanostructures has been described for QWs in the context of solar cell performance^{16,19} but it seems to be more important in QWRs, in the light of our observations. The underlying mechanism by which the photo-charging in QWRs is enhanced is unclear at the moment. Among the possible explanations, there is a higher local electric field for the 1D-like charge distribution geometry along QWRs than for the 2D geometry of QWs, assuming the same amount of accumulated charge inside the nanostructures, or a strong valence band mixing in the QWRs²⁰ leading to higher escape of both heavy and light holes, and therefore higher charge accumulation. Tunnel transport between adjacent QWRs layers across the thinner barrier near the QWR apex and the inhomogeneous strain field around these nanostructures might also play a role in the process.

In this letter, we have studied the properties of strain-balanced InGaAs/GaAsP QWRs grown on 6° misoriented GaAs [100] substrates from the point of view of their performance as solar cells. PL measurements have shown exceedingly long lifetimes, up to 1 μ s, which could be exploited to reduce recombination on this type of devices while increasing their photocurrent. This long lifetime is attributed to an extraordinary photo-charging of the nanostructures, arising as a consequence of the asymmetric carrier escape from the QWRs, and to their longer radiative recombination time than QWs. EQE and IV measurements have confirmed the good performance of QWRs as means of extending the absorption edge of GaAs-based solar cells, comparable to that expected from QWs despite their estimated lower absorption. Polarization losses, possibly the most important drawback *a priori*, have revealed themselves as unimportant in these QWRs.

These encouraging results point to a number of interesting properties that, correctly optimized, could be exploited in high efficiency solar cells. Further work is underway to understand the fundamental mechanisms that lead to the enhanced photo-charging and to demonstrate the applicability of such barely known nanostructures in the field of photovoltaics.

This research was supported by a Marie Curie Intra European Fellowship within the 7th European Community Framework Programme (Grant No. PIEF-GA-2011-302088) and the European Commission and NEDO through the funding of the Project NGCPV EUROPE-JAPAN (EU Ref. No.: 283798). We also wish to thank Professor T. Kita (University of Kobe) for the fruitful discussion regarding this work.

- ¹J. Nelson, I. Ballard, K. W. J. Barnham, J. P. Connolly, J. S. Roberts, and M. Pate, *J. Appl. Phys.* **86**, 5898 (1999).
- ²M. Sugiyama, Y. Wang, S. Choi, Y. Wen, and Y. Nakano, in *2010 35th IEEE Photovolt. Spec. Conf.* (IEEE, 2010), pp. 000376–000379.
- ³D. B. Bushnell, T. N. D. Tibbits, K. W. J. Barnham, J. P. Connolly, M. Mazzer, N. J. Ekins-Daukes, J. S. Roberts, G. Hill, and R. Airey, *J. Appl. Phys.* **97**, 124908 (2005).
- ⁴M. C. Lynch, I. M. Ballard, D. B. Bushnell, J. P. Connolly, D. C. Johnson, T. N. D. Tibbits, K. W. J. Barnham, N. J. Ekins-Daukes, J. S. Roberts, G. Hill, R. Airey, and M. Mazzer, *J. Mater. Sci.* **40**, 1445 (2005).
- ⁵J. G. J. Adams, B. C. Browne, I. M. Ballard, J. P. Connolly, N. L. A. Chan, A. Ioannides, W. Elder, P. N. Stavrinou, K. W. J. Barnham, and N. J. Ekins-Daukes, *Prog. Photovoltaics Res. Appl.* **19**, 865 (2011).
- ⁶H. Sodabanlu, S. Ma, K. Watanabe, M. Sugiyama, and Y. Nakano, *Jpn. J. Appl. Phys., Part 1* **51**, 10ND15 (2012).
- ⁷N. J. Ekins-Daukes, K.-H. Lee, L. Hirst, A. Chan, M. Führer, J. Adams, B. Browne, K. W. J. Barnham, P. Stavrinou, J. Connolly, J. S. Roberts, B. Stevens, R. Airey, and K. Kennedy, *J. Phys. D: Appl. Phys.* **46**, 264007 (2013).
- ⁸S. Hara, J. Ishizaki, J. Motohisa, T. Fukui, and H. Hasegawa, *J. Cryst. Growth* **145**, 692 (1994).
- ⁹C. Giannini, L. Tapfer, Y. Zhuang, L. De Caro, T. Marschner, and W. Stolz, *Phys. Rev. B* **55**, 5276 (1997).
- ¹⁰Y. Xie, G. Gilmer, C. Roland, P. Silverman, S. Buratto, J. Cheng, E. Fitzgerald, A. Kortan, S. Schuppler, M. Marcus, and P. Citrin, *Phys. Rev. Lett.* **73**, 3006 (1994).
- ¹¹S. I. Molina, T. Ben, D. L. Sales, J. Pizarro, P. L. Galindo, M. Varela, S. J. Pennycook, D. Fuster, Y. Gonzalez, and L. González, *Nanotechnology* **17**, 5652 (2006).
- ¹²Y.-C. Chen, P. Wang, J. J. Coleman, D. P. Bour, K. K. Lee, and R. G. Waters, *IEEE J. Quantum Electron.* **27**, 1451 (1991).
- ¹³X.-L. Wang and V. Voliotis, *J. Appl. Phys.* **99**, 121301 (2006).
- ¹⁴M. S. Minsky, S. B. Fleischer, A. C. Abare, J. E. Bowers, E. L. Hu, S. Keller, and S. P. Denbaars, *Appl. Phys. Lett.* **72**, 1066 (1998).
- ¹⁵J. Nelson, M. Paxman, K. W. J. Barnham, J. S. Roberts, and C. Button, *IEEE J. Quantum Electron.* **29**, 1460 (1993).
- ¹⁶S. C. McFarlane, J. Barnes, K. W. J. Barnham, E. S. M. Tsui, C. Button, and J. S. Roberts, *J. Appl. Phys.* **86**, 5109 (1999).
- ¹⁷H. Fujii, K. Watanabe, M. Sugiyama, and Y. Nakano, *Jpn. J. Appl. Phys., Part 1* **51**, 10ND04 (2012).
- ¹⁸P. Dua, S. L. Cooper, A. C. Chen, and K. Y. Cheng, *Appl. Phys. Lett.* **69**, 2261 (1996).
- ¹⁹A. Alemu, J. A. H. Coaquira, and A. Freundlich, *J. Appl. Phys.* **99**, 084506 (2006).
- ²⁰F. Vouilloz, D. Oberli, M.-A. Dupertuis, A. Gustafsson, F. Reinhardt, and E. Kapon, *Phys. Rev. Lett.* **78**, 1580 (1997).