Growth and optical properties of direct band gap Ge/Ge_{0.87}Sn_{0.13} Core/Shell nanowire arrays

Citation for published version (APA):

Assali, S., Dijkstra, A., Li, A., Kölling, S., Verheijen, M. A., Gagliano, L., von den Driesch, N., Buca, D., Koenraad, P. M., Haverkort, J. E. M., & Bakkers, E. P. A. M. (2017). Growth and optical properties of direct band gap Ge/Ge_{0.87}Sn_{0.13} Core/Shell nanowire arrays. *Nano Letters*, 17(3), 1538-1544.
<https://doi.org/10.1021/acs.nanolett.6b04627>

Document license:
TAVERNE

DOI:
[10.1021/acs.nanolett.6b04627](https://doi.org/10.1021/acs.nanolett.6b04627)

Document status and date:
Published: 01/03/2017

Document Version:
Publisher's PDF, also known as Version of Record (includes final page, issue and volume numbers)

Please check the document version of this publication:

- A submitted manuscript is the version of the article upon submission and before peer-review. There can be important differences between the submitted version and the official published version of record. People interested in the research are advised to contact the author for the final version of the publication, or visit the DOI to the publisher's website.
- The final author version and the galley proof are versions of the publication after peer review.
- The final published version features the final layout of the paper including the volume, issue and page numbers.

[Link to publication](#)

General rights

Copyright and moral rights for the publications made accessible in the public portal are retained by the authors and/or other copyright owners and it is a condition of accessing publications that users recognise and abide by the legal requirements associated with these rights.

- Users may download and print one copy of any publication from the public portal for the purpose of private study or research.
- You may not further distribute the material or use it for any profit-making activity or commercial gain
- You may freely distribute the URL identifying the publication in the public portal.

If the publication is distributed under the terms of Article 25fa of the Dutch Copyright Act, indicated by the "Taverne" license above, please follow below link for the End User Agreement:

www.tue.nl/taverne

Take down policy

If you believe that this document breaches copyright please contact us at:

openaccess@tue.nl

providing details and we will investigate your claim.

Growth and Optical Properties of Direct Band Gap Ge/Ge_{0.87}Sn_{0.13} Core/Shell Nanowire Arrays

S. Assali,^{*,†,‡,§} A. Dijkstra,[†] A. Li,^{†,‡,§} S. Koelling,^{†,§} M. A. Verheijen,^{†,||} L. Gagliano,^{†,§} N. von den Driesch,^{†,§} D. Buca,[†] P. M. Koenraad,[†] J. E. M. Haverkort,[†] and E. P. A. M. Bakkers^{†,‡,§}

[†]Department of Applied Physics, Eindhoven University of Technology, 5600 MB Eindhoven, The Netherlands

[‡]Kavli Institute of Nanoscience, Delft University of Technology, 2600 GA Delft, The Netherlands

[§]Beijing Key Lab of Microstructure and Property of Advanced Materials, Beijing University of Technology, Pingleyuan 100, Beijing 100024, P. R. China

^{||}Philips Innovation Laboratories Eindhoven, High Tech Campus 11, 5656AE Eindhoven, The Netherlands

[⊥]Peter Gruenberg Institute 9 (PGI 9) and JARA-Fundamentals of Future Information Technologies, Forschungszentrum Juelich, 52428 Juelich, Germany

Supporting Information

ABSTRACT: Group IV semiconductor optoelectronic devices are now possible by using strain-free direct band gap GeSn alloys grown on a Ge/Si virtual substrate with Sn contents above 9%. Here, we demonstrate the growth of Ge/GeSn core/shell nanowire arrays with Sn incorporation up to 13% and without the formation of Sn clusters. The nanowire geometry promotes strain relaxation in the Ge_{0.87}Sn_{0.13} shell and limits the formation of structural defects. This results in room-temperature photoluminescence centered at 0.465 eV and enhanced absorption above 98%. Therefore, direct band gap GeSn grown in a nanowire geometry holds promise as a low-cost and high-efficiency material for photodetectors operating in the short-wave infrared and thermal imaging devices.

KEYWORDS: Semiconductor nanowire, germanium tin, direct band gap, absorption, photoluminescence

Integration of light-emitting group IV semiconductors on a silicon platform would enable on-chip data communications and infrared (IR) imaging devices compatible with the complementary metal–oxide–semiconductor (CMOS) technology.¹ Currently, the performance of Si- and Ge-based photonics is severely limited by the indirect nature of the band gap of those materials.² Among the several methods proposed to obtain high efficiency group IV photonics,^{3,4} a promising route relies on germanium–tin (GeSn) alloys, where a direct band gap below 0.55 eV (i.e., at wavelengths longer than 2.3 μm) is achieved in strain-free samples with Sn contents above 9%.^{5–8} Low-cost and high efficiency direct band gap GeSn alloys have the potential to strongly improve the performance of photodetectors operating in the short-wave IR (SWIR), with energies in the 0.4–0.9 eV range (i.e., in the 1.4–3 μm wavelength range).^{9,10} SWIR detectors allow imaging through

scattering media, thermal imaging, and molecular vibrations characterization for spectroscopy and biomedical applications. However, current SWIR detector technologies are based on InGaAs, InSb, and HgCdTe, and the high device cost prevents the broad use of SWIR technologies and the development of light sources in this wavelength range.^{1,5,11}

GeSn alloys are commonly grown on a Ge/Si (001)-virtual substrate (VS).^{7,8,11,12} Low temperature growth (below 400 °C) is required to minimize segregation and precipitation of Sn at the surface during growth due to the low equilibrium solubility of Sn in Ge ($\sim 1\%$).^{11,13,14} In addition, the large lattice mismatch between α -Sn and Ge (up to 15%) results in a high

Received: November 5, 2016

Revised: January 15, 2017

Published: February 6, 2017

Figure 1. Ge/Ge_{0.87}Sn_{0.13} core/shell nanowire arrays. (a) SEM picture of an array of Ge NWs (tilting angle 30°) showing the untapered middle-top section of the NWs and a wider bottom segment, which are related to the growth steps at 320 and 425 °C, respectively. (b,c) SEM pictures of Ge/Ge_{0.87}Sn_{0.13} core/shell NW arrays (tilting angle 30°) showing the uniformity of the nanoimprint pattern (b) and a magnified view of the NWs (c).

Figure 2. Radial Sn incorporation in the GeSn shell. (a,b) EDX compositional map (a) and APT image (b) showing the Ge/Ge_{0.87}Sn_{0.13} core/shell structure. A uniform Sn distribution is observed in the GeSn shell along the NW growth axis, while no axial growth of GeSn is present. (c,d) Cross-sectional EDX compositional maps showing a ~120 nm thick GeSn shell, with enhanced Sn incorporation on the {112} side-facets compared to {110}. The integrated tangential composition profile in the yellow rectangle provides an average Sn content of 13 ± 1%, while the radial line-profile (white dashed arrow) is shown in (f). (e) APT measurements showing the Ge core and the inner portion of the GeSn shell. The line-profile (dashed arrow) is shown in (f). (f) Plot of the Sn content as a function of the distance along the radial direction for EDX and APT measurements. After a 10 nm transition region from the Ge core into the GeSn shell, the Sn content gradually increases toward the outer portion of the shell. The blue dashed line is the linear fit of the EDX profile.

amount of strain between the GeSn layer and the Ge substrate.¹¹ Strain in GeSn layers is partially released by the formation of defects at the Ge–GeSn interface (upper limit of $5 \times 10^6 \text{ cm}^{-2}$),⁸ which can trap charge carriers and reduce their mobility. In addition, compressive strain in the GeSn layer increases the amount of Sn required to observe an indirect to direct band gap transition in GeSn alloys.^{6,8,15} Thus, in order to preserve the direct band gap at room-temperature, strain in the GeSn layer has to be relaxed, which requires a minimum layer thickness of 400 nm.⁸ The impact of those limitations in the growth of GeSn bulk heterostructures for opto-electrical devices optimization can be minimized by choosing a nanowire (NW) geometry.¹⁶ Recently, the vapor–liquid–solid (VLS) NW growth of GeSn axial and radial heterostructures was shown.^{17,18} However, the reduced Sn incorporation below 10%, combined with segregation of Sn and formation of defects limit the control of the NW optical properties. Here, we show the growth of Ge/GeSn core/shell heterostructured NW arrays with high Sn content up to $13 \pm 1\%$ in the shell. The unique NW feature of a radial core/shell heterojunction allows strain relaxation and limits the formation of structural defects. In addition, Sn precipitation in the GeSn shell can be avoided by the low growth temperature (300 °C) of these NW heterostructures. The high crystal quality of the NWs leads to a room-temperature direct band gap emission centered at 0.465 eV and enhanced absorption above 98%, which are crucial for the development of SWIR photodetectors. By combining temperature-dependent absorption and photoluminescence measurements we provide new insights in the understanding of the optical emission from the direct band gap nature and the impurities in GeSn alloys.

The NW arrays are grown in a chemical vapor deposition (CVD) reactor with the VLS-growth method using nano-imprint-patterned gold islands as a catalyst^{19,20} and germane (GeH_4), tin chloride (SnCl_4), and hydrogen chloride (HCl) as precursor gases (see Supporting Information S1 for more details). A two-temperature step growth is employed: first, arrays of untapered Ge NWs with diameters of 50 nm are grown at 320 °C, as shown in the scanning electron micrograph (SEM) in Figure 1a; next, the sample is cooled down to 300 °C under GeH_4 supply, and SnCl_4 and HCl are introduced for the GeSn shell growth. With optimized growth parameters two samples with Sn contents of $13 \pm 1\%$ and $10.5 \pm 0.5\%$ in the outer portion of the GeSn shell have been fabricated. Both samples have been characterized, and the experimental data of Ge/ $\text{Ge}_{0.87}\text{Sn}_{0.13}$ (Ge/ $\text{Ge}_{0.895}\text{Sn}_{0.105}$) are presented in the main text (Supporting Information S2). The SEM images in Figure 1b,c show an array of Ge/GeSn NWs with uniform length ($\sim 3.2 \mu\text{m}$) and morphology. The NWs have a tapered top section but are not bent, indicating that there is no significant strain in these structures. A more complex faceting of the shell is visible in the bottom section of the wires, which relates to the small increase in diameter toward the base of the Ge core wires.

The composition of the Ge/GeSn core/shell heterostructure is studied using energy-dispersive X-ray (EDX) compositional maps in scanning transmission electron microscopy (STEM), and correlated with atom probe tomography (APT).^{21,22} A uniform Sn distribution is observed in the GeSn shell along the NW growth axis (see Figure 2a), while the axial growth of GeSn is suppressed, possibly related to a large change in the Au–Ge droplet surface energy upon Sn supply (see Figure S3 in Supporting Information). Interestingly, the EDX and APT data in Figure 2a,b indicate an enhanced Sn incorporation at

specific facets (see also Figures S4 and S5 in Supporting Information). In order to obtain a more precise estimation of the Sn incorporation and distribution, cross-sectional STEM measurements are performed and compared with APT measurements. As shown in the EDX compositional map in Figure 2c,d, the $\sim 120 \text{ nm}$ thick GeSn shell is made of six wide $\{112\}$ facets and six narrow $\{110\}$ corner facets. The higher sensitivity of APT compared to EDX allows a highly detailed chemical mapping of the Ge/GeSn interface in Figure 2e.²¹ Importantly, no clustering or local precipitation of Sn is measured (see Figure S3 in Supporting Information).^{23,24} The Sn compositional profile integrated along the $\langle 112 \rangle$ radial direction in the EDX and APT measurements are shown in Figure 2f. Within the first 10 nm away from the Ge core the Sn content steeply increases to 8%, followed by an almost linear increase to $13 \pm 1\%$ (from the integrated area of the yellow rectangle in Figure 2d) in the next 110 nm. We observe the formation of stripes with a much lower (6%) Sn content and a full width at half-maximum (FWHM) of $5.2 \pm 0.1 \text{ nm}$ along the six equivalent $\langle 110 \rangle$ directions in Figure 2e (see also Figure S5 in Supporting Information). This morphology is similar to that observed in shells of ternary III–V alloys^{25–27} and relates to the different surface energies and adatom mobility of the $\{112\}$ and $\{110\}$ facets,²⁸ possibly influenced by strain.²⁹

The large lattice mismatch between the Ge core and the epitaxial GeSn shell introduces strain, which could result in the formation of defects.^{5–8} To study strain and the crystalline properties of our samples, tens of Ge/ $\text{Ge}_{0.87}\text{Sn}_{0.13}$ (and Ge/ $\text{Ge}_{0.895}\text{Sn}_{0.105}$) core/shell NWs have been examined in TEM along the $[110]$ zone axis, which would reveal the presence of crystalline defects. In the TEM images in Figure 3a,b, no defects, such as cracks or partial dislocations,^{4,6} are observed in the core/shell wires, leading to a pure diamond cubic crystal structure. However, contrast variations in Figure 3b could be associated with the presence of strain and/or surface roughness in the GeSn shell. Therefore, in order to obtain more information on the crystal structure of the NWs aberration-corrected high-resolution TEM (HRTEM) is performed. The HRTEM image in Figure 3c confirms the pure diamond cubic crystal structure of the GeSn shell. However, defect imaging in HRTEM is challenging due to the large sample thickness (above 200 nm), which does not allow to directly investigate the core/shell interface. When imaging in HRTEM the shell in the thinner NW top section, discontinuities in the $\{111\}$ planes are observed, which could be associated with the presence of dislocations in the GeSn shell. Since the strain energy in a core/shell system increases with shell thickness,³⁰ enhanced nucleation of misfit dislocations might be observed at larger GeSn shell thicknesses, which are not accessible in HRTEM. Therefore, more advanced structural studies will be required to decouple thickness, strain, and roughness effects and allow defect imaging in Ge/GeSn core/shell NWs. Lastly, it is important to mention that no protruding dislocations extending from the core–shell interface to the outer edge are observed in the investigated samples, which is in agreement with the observation of optical emission up to room-temperature as discussed later in the text.

In order to evaluate the presence of residual strain, we determine the (local) lattice constants of the GeSn shell in cross sectional view (along the $[111]$ zone axis) from selective area electron diffraction (SAED) patterns and compare this value with bulk GeSn (see Supporting Information S6 for more details). We note that in cross sectional TEM we can estimate

Figure 3. TEM characterization of the nanowire crystal structure along [110] zone axis. (a,b) Low-magnification TEM image of the entire (a) and central segment (b) of a Ge/Ge_{0.87}Sn_{0.13} core/shell NW. The contrast variations in (b) are highlighted using a dashed elliptical shape. (c) Aberration-corrected HRTEM image of the outer portion of the GeSn shell (green square in (b)). Inset: indexed FFT pattern calculated from (c).

the strain level, but eventual misfit dislocations cannot be resolved. For the Ge_{0.87}Sn_{0.13} shell we obtain a lattice parameter $a_0 = 5.73 \pm 0.02$ Å, implying a residual compressive strain of $\sim 0.7\%$. By comparing this with the expected value of $\sim 2\%$ for a Ge_{0.87}Sn_{0.13} layer on a Ge (100) substrate³¹ we can conclude that the NW geometry allows partial strain accommodation in a Ge_{0.87}Sn_{0.13} shell that is only ~ 120 nm thick. This might be achieved with plastic strain relaxation and the formation of misfit dislocations at the GeSn–Ge interface, as discussed above. We note that both the compositional gradient of Sn and the 1D geometry of the nanowires may help to relieve the strain in the shell.

The optical properties of the Ge/GeSn core/shell NWs have been investigated using temperature-dependent integrating-sphere Fourier transform infrared (FTIR) absorption measurements and photoluminescence (PL) as shown in Figure 4.^{32,33} The nature of the band gap and the band gap energy in Ge/Ge_{0.87}Sn_{0.13} NW arrays are first determined by absorption measurements in Figure 4a. The absorption of the Ge/Ge_{0.87}Sn_{0.13} core/shell NW sample is strongly red-shifted and increased with respect to a sample with core Ge NWs (50 nm diameter) only. The increased absorption results from the increased diameter, the tapered NW geometry, and possibly from the direct bandgap Ge_{0.87}Sn_{0.13} shell.^{34–36} The high absorption, close to 100%, allows a precise estimation of the

absorption coefficient α (see Supporting Information S7 for more details). The linear increase of α^2 with energy in Ge/Ge_{0.87}Sn_{0.13} NWs (at both 4 and 300 K) demonstrates that this sample has a direct band gap.³⁷ The band gap energy is determined from a linear extrapolation of α^2 as shown in Figure 4a for different temperatures. The obtained energies are plotted versus temperature in the range from 4 to 300 K and fitted by the Vina's equation in Figure 4c.³⁸ For the sample with lower Sn content (Ge/Ge_{0.895}Sn_{0.105}), we find a deviation from a linear slope of α^2 indicating that both indirect and direct band gap absorption processes take place, which can be explained by compositional gradients in the sample around the indirect/direct transition point (see Figure S7 in Supporting Information).

We now focus on the emission properties of the Ge/Ge_{0.87}Sn_{0.13} NWs sample measured by photoluminescence using an excitation power density of 8 W/cm². At low temperature, the PL spectrum of Ge/Ge_{0.87}Sn_{0.13} NWs consists of two peaks at 0.493 eV (X_1) and 0.468 eV (X_2) (see Figure 4b) with a FWHM of around 60–70 meV, possibly determined by random alloy broadening.^{39,40} The emission peaks X_1 and X_2 are 25 and 50 meV below the band gap energy, determined by absorption spectroscopy at 4 K. In Figure 4c, the PL emission energies are plotted versus the temperature together with the band gap values as determined from absorption measurements. When the temperature is increased from 4 to 100 K the emission of both X_1 and X_2 peaks is red-shifted. Just above 100 K the X_1 peak steeply jumps toward a value close to the band gap energy, followed by a red-shift similar to that of the band gap, and reaches 0.465 eV at room-temperature. Similarly, the X_2 peak also redshifts with temperature and its emission intensity quenches at 170 K. These measurements show that at temperatures below 100 K, the X_1 – X_2 emissions originate from localized states, while at room temperature the emission is associated with the GeSn direct band gap. The slightly higher emission energy in Ge/Ge_{0.87}Sn_{0.13} core/shell NWs with respect to state-of-the-art bulk GeSn with similar Sn contents at room temperature possibly results from the residual $\sim 0.7\%$ compressive strain in the GeSn shell.⁸

The sum of the integrated PL intensities ($X_1 + X_2$) in the temperature range 4–300 K is shown in Figure 4d (see Supporting Information S8 for more details). The PL decreases by a factor ~ 14 when increasing the temperature in this range due to nonradiative recombination and possibly intervalley tunnelling.^{6,41} The increase in PL efficiency above 170 K can be correlated with the quenching in intensity of the X_2 peak and should thus be attributed to thermal detrapping from the localized X_2 states into a higher energy band (see also Figures S8–S9 in Supporting Information).³² From fitting the Arrhenius plot in Figure 4d considering multiple recombination channels (see Supporting Information S8 for more details), we find an activation energy of 20–30 meV for the X_1 emission peak, in agreement to the difference in energy between X_1 and the band gap. Based on these optical studies we attribute X_1 and X_2 to bandgap-related localized states, such as alloy fluctuations (X_1) and acceptors (X_2), broadened by disorder, and could be associated with, e.g., residual aluminum or oxygen in the NWs (see Figure S3 in Supporting Information) or interstitial Sn atoms.^{21,32,33,42}

Finally, we compare the temperature-dependent PL and the estimated quantum efficiency of the Ge/GeSn core/shell NW samples with a state-of-the-art planar Ge_{0.88}Sn_{0.12}/Ge/Si VS sample (see Figure S10 in Supporting Information).^{7,8} The

Figure 4. Temperature dependence of the optical properties of the nanowires. (a) Absorption coefficient α^2 for Ge/Ge_{0.87}Sn_{0.13} core/shell NWs as a function of the energy at 4 and 300 K. Inset: Absorption measurements for Ge NWs and Ge/Ge_{0.87}Sn_{0.13} core/shell NWs acquired at 4 and 300 K. (b) Photoluminescence spectra of Ge/Ge_{0.87}Sn_{0.13} core/shell NWs in the temperature range 4–300 K. Measurements are performed using an excitation power density of 8 W/cm². (c) X₁ and X₂ PL peak energies (spheres) and band gap from absorption data (hollow spheres) as a function of the temperature. The dashed line shows the temperature dependence of the direct band gap emission (eq 3 in Supporting Information S8). PL error bars are estimated from fitting the data with Lorentzian peak functions. Absorption measurements error bars are estimated from the linear fit of α^2 . (d) Integrated PL intensity for the sum of the X₁ + X₂ peaks as a function of the inverse temperature (Arrhenius plot). Inset: enlarged view in the high temperature range. Solid curves represent the fit using two competing recombination channels (eq 1 in Supporting Information S8).

temperature-dependent PL shows very similar behavior, and the ratio between 4 and 300 K emission is nearly identical for all these samples. This provides additional evidence that both the Ge/Ge_{0.87}Sn_{0.13} and the Ge/Ge_{0.895}Sn_{0.105} NW samples feature a direct bandgap in emission. This demonstrates that the lowest Γ -minimum in both NW samples is far enough below the L-minimum to prevent intervalley tunneling.^{32,41} In the Ge/Ge_{0.885}Sn_{0.105} NW sample, the direct band gap emission with the presence of both direct and indirect absorption edges (see Figure S7 in Supporting Information) may be explained by the gradual increase in Sn content toward the outer portion of the shell (see Figure S2 in Supporting Information), progressively reducing the band gap. Since carriers are confined toward lower band gap regions (i.e., higher Sn content) the measured optical emission originates from the outer portion of the GeSn shell, while absorption takes place in the whole

structure. The quantum efficiency has been estimated in our calibrated FTIR setup, leading to a very similar quantum efficiency for GeSn NWs and bulk samples being in the 10^{−3}–10^{−4} range, indicating the presence of nonradiative channels. Since the radiative efficiency of GeSn is independent of the NW or planar geometry, it suggests that these channels are not at the surface, but rather in the bulk of the material, and future work is required to address this point more in detail.

In summary, we have demonstrated the growth of Ge/GeSn core/shell NW arrays with Sn contents up to 13%, high crystalline purity, and direct band gap room temperature emission at 0.465 eV. The enhanced light absorption above 98% in the NW geometry makes Ge/GeSn core/shell nanowire arrays an excellent candidate for photodetectors operating in the SWIR and thermal imaging sensors.^{5,9,10} The very low GeSn NW growth temperature of 300 °C may accelerate the

development of SiGeSn-based materials. With the formation of ternary SiGeSn compounds, enhanced tunability of the direct band gap structure up to 0.9 eV (i.e., 1.4 μm wavelength) becomes possible,^{5,43,44} providing a promising alternative to the 1.55 μm telecommunication infrastructure based on III–V semiconductors.

■ ASSOCIATED CONTENT

Supporting Information

The Supporting Information is available free of charge on the ACS Publications website at DOI: 10.1021/acs.nanolett.6b04627.

Additional information on growth conditions and structural and optical characterization (PDF)

■ AUTHOR INFORMATION

Corresponding Author

*E-mail: simone.assali@polymtl.ca.

ORCID

S. Assali: 0000-0002-3919-9112

S. Koelling: 0000-0002-6606-9110

L. Gagliano: 0000-0003-4214-8639

N. von den Driesch: 0000-0003-0169-6110

E. P. A. M. Bakkers: 0000-0002-8264-6862

Author Contributions

#Department of Engineering Physics, Polytechnique Montréal, C. P. 6079, Succ. Centre-Ville, Montréal, Québec H3C 3A7, Canada.

Author Contributions

S.A. and A.L. planned the epitaxial growth experiments, S.A. grew the Ge/GeSn core/shell nanowires, A.L. and S.A. optimized the Ge nanowires, L.G. prepared the Ge nanoimprint substrate, A.L. and M.A.V. performed the TEM characterization, S.K. prepared the TEM lamella and performed the APT measurements, S.A. and A.D. performed the optical measurements, N.D. and D.B. provided the planar $\text{Ge}_{0.88}\text{Sn}_{0.12}/\text{Ge}/\text{Si}$ VS sample, and P.M.K., J.E.M.H., and E.P.A.M.B. supervised the experiments and coordinated the data interpretation. All authors discussed the results and contributed in writing the manuscript.

Notes

The authors declare no competing financial interest.

■ ACKNOWLEDGMENTS

The authors thank P. J. van Veldhoven for the technical support with the CVD reactor and Sonia Conesa-Boj for the fruitful discussion. This work was supported by the Dutch Organization for Scientific Research (NWO-VICI 700.10.441), the Dutch Technology Foundation (STW 12744), and Philips Research. NWO is also acknowledged for funding the Atom Probe and the Nanowire Growth facility. Solliance is acknowledged for funding the TEM facility.

■ REFERENCES

- (1) Soref, R.; Buca, D.; Yu, S.-Q. *Opt. Photonics News* **2016**, 27 (1), 32–39.
- (2) Paniccia, M. *Nat. Photonics* **2010**, 4 (8), 498–499.
- (3) Süess, M. J.; Geiger, R.; Minamisawa, R. a.; Schiefler, G.; Frigerio, J.; Chrastina, D.; Isella, G.; Spolenak, R.; Faist, J.; Sigg, H. *Nat. Photonics* **2013**, 7 (June), 466–472.

- (4) Conesa-Boj, S.; Hauge, H. I. T.; Verheijen, M. A.; Assali, S.; Li, A.; Bakkers, E. P. A. M.; Fontcuberta i Morral, A. *Nano Lett.* **2015**, 15 (5), 2974–2979.
- (5) Moontragoon, P.; Soref, R. A.; Ikonik, Z. *J. Appl. Phys.* **2012**, 112 (7), 73106.
- (6) Wirths, S.; Geiger, R.; von den Driesch, N.; Mussler, G.; Stoica, T.; Mantl, S.; Ikonik, Z.; Luysberg, M.; Chiussi, S.; Hartmann, J. M.; Sigg, H.; Faist, J.; Buca, D.; Grützmacher, D. *Nat. Photonics* **2015**, 9 (2), 88–92.
- (7) Stange, D.; Wirths, S.; Von Den Driesch, N.; Mussler, G.; Stoica, T.; Ikonik, Z.; Hartmann, J. M.; Mantl, S.; Grützmacher, D.; Buca, D. *ACS Photonics* **2015**, 2 (11), 1539–1545.
- (8) Von Den Driesch, N.; Stange, D.; Wirths, S.; Mussler, G.; Holländer, B.; Ikonik, Z.; Hartmann, J. M.; Stoica, T.; Mantl, S.; Grützmacher, D.; Buca, D. *Chem. Mater.* **2015**, 27 (13), 4693–4702.
- (9) Hansen, M. P.; Malchow, D. S. *Proc. SPIE* **2008**, 6939, 69390I–69390I–1.
- (10) Dhar, N. K.; Dat, R.; Sood, A. K. In *Optoelectronics - Advanced Materials and Devices*; InTech, 2013; Chapter 7.
- (11) Wirths, S.; Buca, D.; Mantl, S. *Prog. Cryst. Growth Charact. Mater.* **2016**, 62 (1), 1–39.
- (12) He, G.; Atwater, H. A. *Phys. Rev. Lett.* **1997**, 79 (10), 1937–1940.
- (13) Olesinski, R. W.; Abbaschian, G. J. *Bull. Alloy Phase Diagrams* **1984**, 5 (3), 265–271.
- (14) Takeuchi, S.; Sakai, A.; Nakatsuka, O.; Ogawa, M.; Zaima, S. *Thin Solid Films* **2008**, 517 (1), 159–162.
- (15) Attiaoui, A.; Moutanabbir, O. *J. Appl. Phys.* **2014**, 116 (6), 063712.
- (16) Day, R. W.; Mankin, M. N.; Gao, R.; No, Y.-S.; Kim, S.-K.; Bell, D. C.; Park, H.-G.; Lieber, C. M. *Nat. Nanotechnol.* **2015**, 10 (4), 345–352.
- (17) Biswas, S.; Doherty, J.; Saladukha, D.; Ramasse, Q.; Majumdar, D.; Upmanyu, M.; Singha, A.; Ochalski, T.; Morris, M. A.; Holmes, J. D. *Nat. Commun.* **2016**, 7, 11405.
- (18) Meng, A. C.; Fenrich, C. S.; Braun, M. R.; McVittie, J. P.; Marshall, A. F.; Harris, J. S.; McIntyre, P. C. *Nano Lett.* **2016**, 16, 7521.
- (19) Wagner, R. S.; Ellis, W. C. *Appl. Phys. Lett.* **1964**, 4 (1964), 89–90.
- (20) Assali, S.; Gagliano, L.; Oliveira, D. S.; Verheijen, M. A.; Feiner, L.-F.; Plissard, S. R.; Bakkers, E. P. A. M. *Nano Lett.* **2015**, 15 (12), 8062–8069.
- (21) Koelling, S.; Li, A.; Cavalli, A.; Assali, S.; Car, D.; Gazibegovic, S.; Bakkers, E. P. A. M.; Koenraad, P. M. *Nano Lett.* **2017**, 17 (2), 599–605.
- (22) Seidman, D. N. *Annu. Rev. Mater. Res.* **2007**, 37 (1), 127–158.
- (23) Kumar, A.; Komalan, M. P.; Lenka, H.; Kambham, A. K.; Gilbert, M.; Gencarelli, F.; Vincent, B.; Vandervorst, W. *Ultra-microscopy* **2013**, 132, 171–178.
- (24) Kumar, A.; Demeulemeester, J.; Bogdanowicz, J.; Bran, J.; Melkonyan, D.; Fleischmann, C.; Gencarelli, F.; Shimura, Y.; Wang, W.; Loo, R.; Vandervorst, W. *J. Appl. Phys.* **2015**, 118 (2), 25302.
- (25) Heiss, M.; Ketterer, B.; Uccelli, E.; Morante, J. R.; Arbiol, J.; Fontcuberta i Morral, A. *Nanotechnology* **2011**, 22 (19), 195601.
- (26) Rudolph, D.; Funk, S.; Döblinger, M.; Morkötter, S.; Hertenberger, S.; Schweickert, L.; Becker, J.; Matich, S.; Bichler, M.; Spirkoska, D.; Zardo, I.; Finley, J. J.; Abstreiter, G.; Koblmüller, G. *Nano Lett.* **2013**, 13 (4), 1522–1527.
- (27) Zheng, C.; Wong-Leung, J.; Gao, Q.; Tan, H. H.; Jagadish, C.; Etheridge, J. *Nano Lett.* **2013**, 13 (8), 3742–3748.
- (28) Migas, D. B.; Borisenko, V. E.; Rusli; Soci, C. *Nano Conver.* **2015**, 2 (1), 16.
- (29) Grönqvist, J.; Søndergaard, N.; Boxberg, F.; Guhr, T.; Åberg, S.; Xu, H. Q. *J. Appl. Phys.* **2009**, 106, 053508.
- (30) Trammell, T. E.; Zhang, X.; Li, Y.; Chen, L. Q.; Dickey, E. C. *J. Cryst. Growth* **2008**, 310 (12), 3084–3092.
- (31) Gencarelli, F.; Vincent, B.; Demeulemeester, J.; Vantomme, A.; Moussa, A.; Franquet, A.; Kumar, A.; Bender, H.; Meersschaut, J.

Vandervorst, W.; Loo, R.; Caymax, M.; Temst, K.; Heyns, M. *ECS J. Solid State Sci. Technol.* **2013**, 2 (4), P134–P137.

(32) Leroux, M.; Grandjean, N.; Beaumont, B.; Nataf, G.; Semon, F.; Massies, J.; Gibart, P. *J. Appl. Phys.* **1999**, 86 (7), 3721–3728.

(33) Assali, S.; Greil, J.; Zardo, I.; Belabbes, A.; de Moor, M. W. A.; Koelling, S.; Koenraad, P. M.; Bechstedt, F.; Bakkers, E. P. A. M.; Haverkort, J. E. M. *J. Appl. Phys.* **2016**, 120 (4), 44304.

(34) Diedenhofen, S. L.; Janssen, O. T. A.; Grzela, G.; Bakkers, E. P. A. M.; Gómez Rivas, J. *ACS Nano* **2011**, 5 (3), 2316–2323.

(35) Attiaoui, A.; Wirths, S.; Blanchard-Dionne, A.-P.; Meunier, M.; Hartmann, J. M.; Buca, D.; Moutanabbir, O. Under review.

(36) Reimer, M. E.; Bulgarini, G.; Akopian, N.; Hocevar, M.; Bavinck, M. B.; Verheijen, M. a.; Bakkers, E. P. A. M.; Kouwenhoven, L. P.; Zwiller, V. *Nat. Commun.* **2012**, 3 (737), 737.

(37) Grundmann, M. *The Physics of Semiconductors*; Springer, 2006.

(38) Viña, L.; Logothetidis, S.; Cardona, M. *Phys. Rev. B: Condens. Matter Mater. Phys.* **1984**, 30 (4), 1979–1991.

(39) Weber, J.; Alonso, M. I. *Phys. Rev. B: Condens. Matter Mater. Phys.* **1989**, 40 (8), 5683–5693.

(40) Schubert, E. F.; Göbel, E. O.; Horikoshi, Y.; Ploog, K.; Queisser, H. J. *Phys. Rev. B: Condens. Matter Mater. Phys.* **1984**, 30 (2), 813–820.

(41) Pezzoli, F.; Isa, F.; Isella, G.; Falub, C. V.; Kreiliger, T.; Salvalaglio, M.; Bergamaschini, R.; Grilli, E.; Guzzi, M.; von Känel, H.; Miglio, L. *Phys. Rev. Appl.* **2014**, 1 (4), 44005.

(42) Querales-Flores, J. D.; Ventura, C. I.; Fuhr, J. D.; Barrio, R. A. *J. Appl. Phys.* **2016**, 120 (10), 105705.

(43) Jiang, L.; Xu, C.; Gallagher, J. D.; Favaro, R.; Aoki, T.; Menendez, J.; Kouvetakis, J. *Chem. Mater.* **2014**, 26 (8), 2522–2531.

(44) Wendav, T.; Fischer, I. A.; Montanari, M.; Zoellner, M. H.; Klesse, W.; Capellini, G.; von den Driesch, N.; Oehme, M.; Buca, D.; Busch, K.; Schulze, J. *Appl. Phys. Lett.* **2016**, 108 (24), 242104.