

Growth rate of hexagonal SiGe multi-quantum wells

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Direct-band-gap light sources from group IV materials are highly desired for photonic integration. Among the possible candidates is the hex-Si_{1-x}Ge_x system, which has been demonstrated to have a direct band gap at least for $x > 65\%$. Recently, Ge-rich heterostructures of hex-Si_{1-x}Ge_x/Si_{1-y}Ge_y ($x < y$) have been demonstrated to have a type-I alignment. This unlocks opportunities for devices utilizing low-dimensional structures, such as quantum wells (QWs), based on group IV materials. Multi-quantum-well (MQW) structures can significantly reduce lasing thresholds, and a full understanding of the growth kinetics is necessary to grow MQWs with identical well thicknesses. Here, we report a study of hex-Si_{1-x}Ge_x multishells, of which the growth rate increases by a factor 4 over a shell thickness of 500 nm, indicating nucleation-limited growth. We present a model that hints that the increasing growth rate is due to the accumulation of cubic stacking faults at the growth front. This proposed mechanism is generic for the hex-SiGe system, and it has to be considered for any multilayer structure of hex-SiGe.

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Introduction. Low-dimensional semiconductor heterostructures revolutionized optoelectronic devices [1–3]. In lasing devices, charge carriers get localized in the quantum well (QW), which induces population inversion. Apart from single QWs, multi-quantum-well (MQW) structures can offer increased benefits. The overlap between the gain medium and the optical mode can be increased, which decreases the lasing threshold [4]. However, to maximize the effect, each QW in the MQW structure should emit at the same wavelength. This requires the QWs to have an equal thickness and equal compositional profile. Therefore, a complete understanding of the growth kinetics is essential.

Here, we study the growth kinetics of hex-Si_{1-x}Ge_x/Si_{1-y}Ge_y ($x < y$) multishell structures. Alloys of Si_{1-x}Ge_x with a hexagonal crystal structure have a direct band gap for $x > 65\%$ [5]. Moreover, Ge-rich heterostructures, such as Ge/Si_{0.2}Ge_{0.8}, have recently been demonstrated to have type-I band alignment between layers [6]. Both ingredients make hex-Si_{1-x}Ge_x a prime candidate as a silicon-based material for lasers and optical amplifiers. The realization of hex-Si_{1-x}Ge_x/Si_{1-y}Ge_y MQW structures is the next milestone on the roadmap towards a room-temperature laser.

The growth kinetics of cubic Ge/SiGe MQWs, grown as thin films, are well understood, and identical Ge/Si_{0.18}Ge_{0.82}

QWs in a MQW structure can be realized by utilizing the same growth time for each repetition [7]. When growing hex-Si_{1-x}Ge_x MQWs, we face two major differences compared to its cubic counterpart. First, hex-Si_{1-x}Ge_x alloys are currently grown as a shell within a nanowire (NW) geometry. For radial QWs in NWs, each subsequent QW grows on a larger circumference of the NW. Consecutive radial QWs of equal thickness are thus made of a larger crystal volume due to the increase in circumference. In the case of III-V NW QWs, each successive radial QW is grown with increased growth time [4,8,9], indicating that these quantum wells are grown in a mass-transport-limited regime. The second difference compared to cubic Si_{1-x}Ge_x is the metastable crystal phase of the hex-Si_{1-x}Ge_x alloys, implying that stacking faults have to be considered. It has been shown that Si_{1-x}Ge_x grows faster on a mixed-phase structure compared to a hexagonal one [10]. Moreover, during the growth of hex-Si_{1-x}Ge_x, cubic inclusions are accumulated, known as the I_3 basal stacking faults [11]. Thus, the fraction of hexagonal (mixed-phase) stacking at the growth front is expected to decrease (increase) with shell thickness. It is not yet distinctively proven that the I_3 stacking faults, which increase the mixed-phase stacking at the growth front, induce a faster growth rate. However, *in situ* transmission electron microscopy (TEM) studies of hex-Ge growth tentatively indicate that nucleation of a new monolayer is preferable on the local cubic structure of the I_3 stacking fault [12]. In this case, the increasing nucleation probability with accumulating stacking fault density would result in a rising growth rate with shell thickness, being strikingly different from the mass-transport-limited growth regime.

Ge/Si_{0.2}Ge_{0.8} multishells. The growth rate of hex-Si_{1-x}Ge_x is investigated in arrays of core-multishell NWs [Fig. 1(a)], which are grown using metal-organic vapor phase epitaxy (MOVPE) [see the Methods section of the Supplemental

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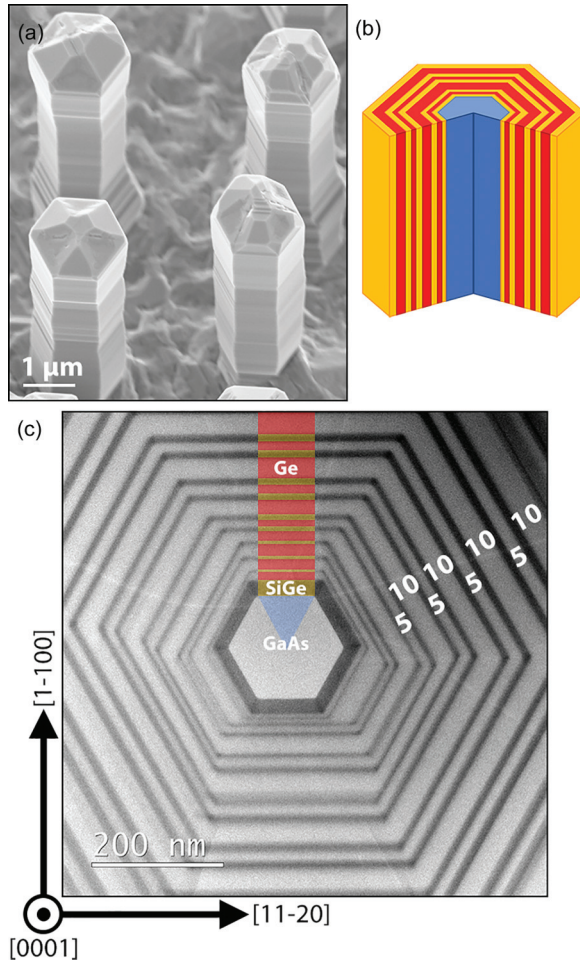
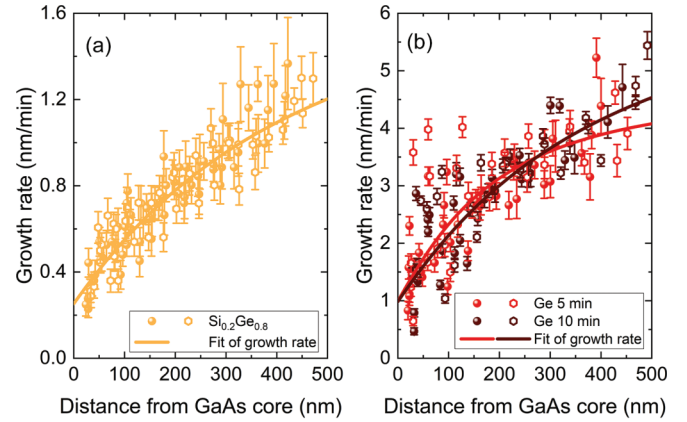


FIG. 1. Ge/Si_{0.2}Ge_{0.8} multishell structures. (a) 30° tilted SEM image of a NW array. (b) Schematic of the NWs, which have a core-multishell structure. (c) High-angle annular dark-field scanning transmission electron microscopy (HAADF-STEM) image of a cross-sectional lamella, viewing the multishell structure in the [0001] zone axis. Several Ge shells that are grown for 5 or 10 min are labeled.

Material (SM) I [13]]. These NWs consist of wurtzite (WZ) GaAs cores, which are used as templates to copy the hexagonal stacking to Si_{1-x}Ge_x shells [5]. This crystal transfer method [14] can also be used to copy the hexagonal stacking over multiple interfaces [15,16], which we use to grow alternating shells of hex-Si_{0.2}Ge_{0.8} and Ge [Fig. 1(b)]. Within such a structure, the Ge and Si_{0.2}Ge_{0.8} layers would respectively be the wells and barriers of the MQW structure. To get more information on the growth of the well, we decided to alternate the growth time of the Ge shell between 5 and 10 min, while keeping the growth time of the Si_{0.2}Ge_{0.8} barriers constant. This creates a structure with alternating thin and thick Ge shells separated by Si_{0.2}Ge_{0.8}, of which a cross section is shown in Fig. 1(c) (an additional NW is shown in SM II [13]).

The cross section in Fig. 1(c) clearly demonstrates an increasing shell thickness, i.e., growth rate, for both Ge and Si_{0.2}Ge_{0.8} when going further away from the core (Fig. 2). The growth rate increases with a factor 4 during 500 nm of shell growth, while the input material fluxes and temperature



	Si _{0.2} Ge _{0.8}	Ge 5 min	Ge 10 min
α (-)	1.2 ± 0.2	1.6 ± 0.6	1.2 ± 0.4
c_0 (10^4 sec)	4 ± 4	9 ± 30	1 ± 3
$\langle t_{\text{completion}} \rangle$ (sec)	10 ± 3	4 ± 1	3 ± 2

FIG. 2. Growth rate increase in the Ge/Si_{0.2}Ge_{0.8} multishell. (a) Increase of growth rate of the Si_{0.2}Ge_{0.8} sections in the Ge/Si_{0.2}Ge_{0.8} shells, as a function of distance to GaAs core. Data from two NWs are included, with the solid/open symbols. The nucleation-limited model of Eq. (5) provides a good fit to the data. (b) Growth rate increase for Ge shells. The growth rate also increases with distance to the core, although large deviations in the growth rate between different facets are observed close to the core. The two growth times are indistinguishable. Shell thicknesses that could not be measured, due to changing shell position along the [0001] direction, are excluded from the analysis. (c) Table of fit parameters obtained from fitting Eq. (5) to the data.

are kept constant. Thus, both the circumference and the shell thickness increase with distance from the core, indicating a strong increase in the deposited volume. A mass-transport-limited regime, which includes adatom diffusion from the substrate [17], cannot explain an increase in deposited volume with nanowire radius. Instead, it indicates a nucleation-limited regime.

Additional evidence for a nucleation-limited regime is found by examining the growth rate variation along the nanowire. The growth rate is higher near the bottom, but also near the top (see SM III [13]). The increased growth rate near the bottom could be consistent with a mass-transport-limited growth, since the substrate diffusion of precursors on the substrate provides additional material to the bottom of each NW. However, the enhanced growth rate near the top cannot be explained with mass-transport-limited growth. Here, we observe that the growth of cubic Si_{1-x}Ge_x on top of the [0001] facet of GaAs is faster. The faster growth of cubic material results in the emission of hexagonal growth steps from the top downwards, highlighting the importance of step nucleation (see SM IV [13]). In the remainder of this Letter, we will focus on the growth rate in the middle of the NW, i.e., at several μm distance from both the top and the bottom of the NW.

There are two factors that may induce a faster nucleation rate. The first factor is a geometric effect, since the sidewall area scales with the NW radius, and the probability of forming a two-dimensional (2D) nucleus *anywhere* on the sidewall

scales with the area of the sidewall. The second factor is an accumulating stacking fault density at the growth front, and we indeed see that the outer shells have more local cubic ABC stacking (see SM III [13]). We note that, despite the high number of stacking faults, photoluminescence from these samples is still observed (SM V [13]).

Model. The growth rate increase must be due to an increasing nucleation rate in a nucleation-limited growth regime. A model is necessary to identify the responsible mechanism of nucleation.

First, we postulate that the radial shell growth occurs by the addition of one single monolayer at a time. Similar to birth-and-spread models [18,19], these monolayers are assumed to grow in two sequential stages. First, local 2D nuclei are formed at different locations on the NW sidewall after a time $\langle t_{\text{nucleus}} \rangle$. In the second stage, the 2D nuclei expand to form a complete monolayer in a time $\langle t_{\text{completion}} \rangle$. The latter is treated as a constant, since the capture area of the facet increases with the circumference, assuming a negligible contribution from substrate diffusion to the growth rate in the middle of the NW, while the amount of material required to form a complete monolayer is also proportional to the NW circumference. Thus, both effects cancel each other out. Moreover, the nucleation and layer completion time are assumed to be independent of each other, such that the time to form a complete monolayer is

$$\langle t_{\text{ML}} \rangle = \langle t_{\text{nucleus}} \rangle + \langle t_{\text{completion}} \rangle. \quad (1)$$

The time to form a nucleus $\langle t_{\text{nucleus}} \rangle$ can be argued using classical nucleation theory [20], which states that the rate of nucleation R is given by

$$\frac{1}{\langle t_{\text{nucleus}} \rangle} = R \propto n_s \exp\left(-\frac{\Delta G}{k_B T}\right), \quad (2)$$

with n_s the number of nucleation sites and ΔG the energetic barrier for forming a 2D nucleus on the NW sidewall. Here, we assume that the energetic barrier ΔG does not change with shell thickness. The energetic cost of forming an island is a function of the surface and interface energies of the nucleation island and the NW sidewall. These energies change due to both a changing lattice mismatch and a changing $\text{Si}_{1-x}\text{Ge}_x$ composition due to a gradient in composition (for example, from 20% to 30% Si over several hundred nm [6]). Note that for Ge growth, the compositional gradient does not occur. The influence of lattice mismatch and compositional gradient on the island formation is assumed to be smaller than uncertainties of the experiment, and the exponential term is thus treated as a constant for each of the different growth conditions.

On the contrary, the number of nucleation sites will depend on the NW radius. We introduce a power law between the number of nucleation sites n_s and the radius of the NW r , according to

$$n_s \propto r^\alpha. \quad (3)$$

With this power law, we can identify three scenarios, each with a different nucleation mechanism, based on the value of α .

(1) $\alpha = 0$: n_s is radius independent, the growth is not nucleation limited.

(2) $\alpha = 1$: n_s scales linearly with r , as the circumference of the NW.

(3) $\alpha > 1$: n_s scales superlinearly with r , i.e., defect mediated.

The value of α , obtained from fitting the model to the data, thus provides information on the mechanism of 2D nucleus formation on the NW sidewall. For $\alpha = 1$, the number of nucleation sides scales exactly with the NW sidewall area, with no preferential nucleation on cubic or hexagonal stacking. In contrast, we identify the last scenario, $\alpha > 1$, as a defect-mediated mechanism. Here, the number of nucleation sides increases faster with the r than the circumference of the NW. We ascribe this condition to the preferential nucleation on the cubic stacking sequences at the growth front, such as the I_3 defects. The amount of nucleation sides, i.e., the number of atoms with the defected I_3 stacking at the growth front, does increase superlinearly with r . This is because the number of atoms in a cubic position from an individual I_3 defect grows linearly with thickness [21], and the formation of additional I_3 defects effectively results in the superlinear dependence with r .

The power law [Eq. (3)] is substituted in the nucleation theory [Eq. (2)], resulting in

$$\langle t_{\text{nucleus}} \rangle = c_0 \left(\frac{r}{\frac{1}{2}\sqrt{3}a} \right)^{-\alpha}. \quad (4)$$

Here, the term between parentheses represents the number of radial monolayers in the NW and we introduced the proportionality constant c_0 , which links the rate of nucleation events to the radial growth rate. The thickness of one monolayer on the $\{1\bar{1}00\}$ sidewall is $\frac{1}{2}\sqrt{3}a$, with a the lattice constant of the hexagonal lattice. This thickness gets added in a growth time of $\langle t_{\text{ML}} \rangle$, which is found by combining Eqs. (1) and (4), resulting in a growth rate $\Delta r / \Delta t$ of

$$\frac{\Delta r}{\Delta t} = \frac{\frac{1}{2}\sqrt{3}a}{\langle t_{\text{ML}} \rangle} = \frac{\frac{1}{2}\sqrt{3}a \left(\frac{2}{3}\sqrt{3}\frac{r}{a} \right)^\alpha}{c_0 + \langle t_{\text{completion}} \rangle \left(\frac{2}{3}\sqrt{3}\frac{r}{a} \right)^\alpha}. \quad (5)$$

The growth rate model is fitted to the data, resulting in fit parameters reported in Fig. 2(c). For both compositions, a value of $\alpha > 1$ is obtained, suggesting that preferential nucleation on the accumulating stacking faults at the growth front causes the increase in growth rate. The model fits best to $\text{Si}_{0.2}\text{Ge}_{0.8}$, and we calculate that $\langle t_{\text{nucleus}} \rangle$ and $\langle t_{\text{completion}} \rangle$ are equal around 340 nm from the GaAs core. Beyond this distance, according to the model, the time for nucleus formation still decreases, but the layer completion time becomes the growth rate-limiting step. Consequently, the growth rate flattens at large distances from the GaAs core. The model predicts that a transition occurs from a nucleation-limited regime towards a layer completion-limited, e.g., mass-transport-limited, regime for thicker shells.

Compared to $\text{Si}_{0.2}\text{Ge}_{0.8}$, the model describes the experimental data significantly worse for the Ge shells of both the 5- and 10-min growth time [Fig. 2(b)]. The uncertainty on the fit parameters is larger, with the proportionality constant c_0 even having a larger error bar than the value itself [Fig. 2(c)]. We attribute the large uncertainty in c_0 to the growth rate variations of the inner shells. In the inner shells, the thickness

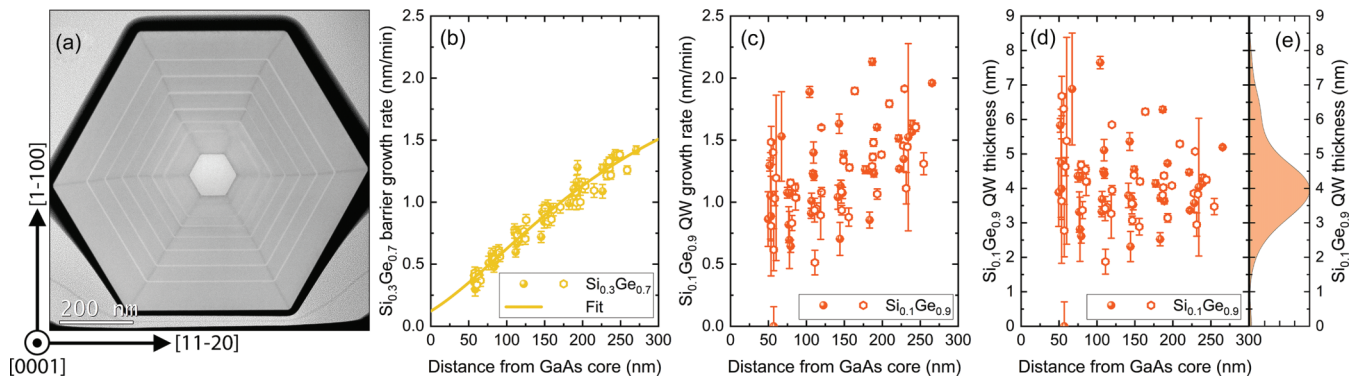


FIG. 3. $\text{Si}_{0.1}\text{Ge}_{0.9}/\text{Si}_{0.3}\text{Ge}_{0.7}$ multi-quantum well structure. (a) HAADF-STEM image of a cross-sectional lamella, viewing the $\text{Si}_{0.1}\text{Ge}_{0.9}/\text{Si}_{0.3}\text{Ge}_{0.7}$ MQW structure along the $[0001]$ zone axis. (b) Growth rate increase in the $\text{Si}_{0.3}\text{Ge}_{0.7}$ shells, as a function of distance from GaAs core. Data from two NW cross sections are included and represented with the solid/open symbols. The nucleation-limited model [Eq. (5)] provides a good fit to the data. (c) Growth rate of the $\text{Si}_{0.1}\text{Ge}_{0.9}$ QWs: The growth rate model does not describe the data accurately, as the variations in growth rate are large. (d) QW thickness as a function of distance from the GaAs core, together with (e) the approximate probability distribution of QW thickness obtained by kernel smoothing, with an average thickness of 4 ± 1 nm.

differs strongly between the six equivalent $\{1\bar{1}00\}$ facets. This anisotropy of the growth rate is larger for the inner Ge shells compared to the $\text{Si}_{0.2}\text{Ge}_{0.8}$ shells, for a reason that has yet to be identified, resulting in a worse fit. A growth model that correctly describes the anisotropic growth rate on the equivalent $\{1\bar{1}00\}$ facets is subject to future research. Yet, all fit parameters of the 5- and 10-min Ge shells are consistent with each other within error bars.

Multi-quantum wells of $\text{Si}_{0.1}\text{Ge}_{0.9}/\text{Si}_{0.3}\text{Ge}_{0.7}$. In addition to Ge/ $\text{Si}_{0.2}\text{Ge}_{0.8}$, we also observed the nucleation-limited regime for seven other compositions between 0% and $\approx 35\%$ Si (SM VI [13]), and use this knowledge to proceed with the growth of $\text{Si}_{0.1}\text{Ge}_{0.9}/\text{Si}_{0.3}\text{Ge}_{0.7}$ MQWs, with an additional 10% Si in both well and barrier compared to Fig. 2. The nucleation-limited growth regime needs to be taken into account for the growth of MQWs, when aiming to have identical QW thickness through decreasing the growth time of consecutive QWs. Therefore, we multiplied the growth time of each subsequent barrier and shell by a factor of 0.9 (SM I [13]). The resulting structure is shown in Fig. 3(a), where the six $\text{Si}_{0.1}\text{Ge}_{0.9}$ QWs can be recognized as brighter rings in the shell.

Using the QWs as markers, we conclude that the growth rate of the $\text{Si}_{0.3}\text{Ge}_{0.7}$ barriers increases strongly with distance from the core [Fig. 3(b)], confirming once more the nucleation-limited growth. The growth rate increases more than compensated for by the growth time, resulting in larger barriers further away from the core. A fit of Eq. (5) results in $\alpha = 1.8 \pm 0.2$, a value clearly above 1, suggesting preferential nucleation on cubic defects. However, the value of α should be interpreted with care, since $[c_0 = (2 \pm 3) \times 10^6 \text{ s}]$ has a value which is smaller than its uncertainty. A c_0 of zero means that there is zero time required to form a nucleus [Eq. (4)]. Possibly, the assumption that there is only a single monolayer growing at each instance does not hold, and mul-

multiple 2D nuclei might form on top of each other on the NW sidewall.

For the $\text{Si}_{0.1}\text{Ge}_{0.9}$ QWs, no clear growth rate increase can be deduced from the data, as the variation in growth rate is too large [Fig. 3(c)]. Consequently, the nucleation-limited growth model does not describe the data accurately. The average growth rate of all QWs is 1.2 ± 0.4 nm/min, and the adjustment of the growth time with every subsequent QW results in a slightly narrower distribution of QW thickness at an average of 4 ± 1 nm [Figs. 3(d) and 3(e)].

Conclusion. The radial shell growth rate of hex- $\text{Si}_{1-x}\text{Ge}_x$ increases with shell thickness. Thus, the growth of hex- $\text{Si}_{1-x}\text{Ge}_x$ occurs in a nucleation-limited regime. Consequently, it is necessary to decrease the growth time of subsequent QWs to achieve MQW structures with equal well thickness. A fit of the growth rate hints that the growth rate increases due to an accumulating I_3 stacking fault density at the growth front. Future work should focus on developing a growth model that also correctly describes both the anisotropic growth rate on equivalent facets, and the $\text{Si}_{1-x}\text{Ge}_x$ compositional profile. Additionally, further insight can be obtained by studying the growth rate of hex- $\text{Si}_{1-x}\text{Ge}_x$ as a function of growth conditions such as temperature and MOVPE flows. Based on this Letter, a growth rate increase may thus even be expected for hex- $\text{Si}_{1-x}\text{Ge}_x$ thin films, rather than NWs, due to the defect-mediated nucleation mechanism.

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Data availability. The data that support the findings of this article are openly available [22].

[1] B. Levine, Quantum-well infrared photodetectors, *J. Appl. Phys.* **74**, R1 (1993).

[2] Y.-H. Kuo, Y. K. Lee, Y. Ge, S. Ren, J. E. Roth, T. I. Kamins, D. A. Miller, and J. S. Harris, Strong quantum-confined Stark

- effect in germanium quantum-well structures on silicon, *Nature (London)* **437**, 1334 (2005).
- [3] S. Chen, W. Li, J. Wu, Q. Jiang, M. Tang, S. Shutts, S. N. Elliott, A. Sobiesierski, A. J. Seeds, I. Ross *et al.*, Electrically pumped continuous-wave III–V quantum dot lasers on silicon, *Nat. Photonics* **10**, 307 (2016).
- [4] F. Qian, Y. Li, S. Gradečak, H.-G. Park, Y. Dong, Y. Ding, Z. L. Wang, and C. M. Lieber, Multi-quantum-well nanowire heterostructures for wavelength-controlled lasers, *Nat. Mater.* **7**, 701 (2008).
- [5] E. M. T. Fadaly, A. Dijkstra, J. R. Suckert, D. Ziss, M. A. J. Van Tilburg, C. Mao, Y. Ren, V. T. van Lange, K. Korzun, S. Kölling *et al.*, Direct-bandgap emission from hexagonal Ge and SiGe alloys, *Nature (London)* **580**, 205 (2020).
- [6] W. H. J. Peeters, V. T. van Lange, A. Belabbes, M. C. van Hemert, M. M. Jansen, R. Farina, M. A. J. van Tilburg, M. A. Verheijen, S. Botti, F. Bechstedt *et al.*, Direct bandgap quantum wells in hexagonal silicon germanium, *Nat. Commun.* **15**, 5252 (2024).
- [7] S. Assali, S. Koelling, Z. Abboud, J. Nicolas, A. Attiaoui, and O. Moutanabbir, 500-period epitaxial Ge/Si_{0.18}Ge_{0.82} multi-quantum wells on silicon, *J. Appl. Phys.* **132**, 175304 (2022).
- [8] D. Saxena, N. Jiang, X. Yuan, S. Mokkaapati, Y. Guo, H. H. Tan, and C. Jagadish, Design and room-temperature operation of GaAs/AlGaAs multiple quantum well nanowire lasers, *Nano Lett.* **16**, 5080 (2016).
- [9] F. Zhang, X. Zhang, Z. Li, R. Yi, Z. Li, N. Wang, X. Xu, Z. Azimi, L. Li, M. Lysevych *et al.*, A new strategy for selective area growth of highly uniform InGaAs/InP multiple quantum well nanowire arrays for optoelectronic device applications, *Adv. Funct. Mater.* **32**, 2103057 (2022).
- [10] W. H. J. Peeters, M. Vettori, E. M. T. Fadaly, A. Danescu, C. Mao, M. A. Verheijen, and E. P. A. M. Bakkers, Onset of uncontrolled polytypism during the Au-catalyzed growth of wurtzite GaAs nanowires, *Phys. Rev. Mater.* **8**, L020401 (2024).
- [11] E. M. T. Fadaly, A. Marzegalli, Y. Ren, L. Sun, A. Dijkstra, D. De Matteis, E. Scalise, A. Sarikov, M. De Luca, R. Rurali *et al.*, Unveiling planar defects in hexagonal group IV materials, *Nano Lett.* **21**, 3619 (2021).
- [12] L. Vincent, E. M. T. Fadaly, C. Renard, W. H. J. Peeters, M. Vettori, F. Panciera, D. Bouchier, E. P. A. M. Bakkers, and M. A. Verheijen, Growth-related formation mechanism of *I*₃-type basal stacking fault in epitaxially grown hexagonal Ge-2H, *Adv. Mater. Inter.* **9**, 2102340 (2022).
- [13] See Supplemental Material at <http://link.aps.org/supplemental/10.1103/d3zf-jft6> for additional information regarding the methods and TEM imaging.
- [14] R. E. Algra, M. Hocevar, M. A. Verheijen, I. Zardo, G. G. Immink, W. J. van Enkevort, G. Abstreiter, L. P. Kouwenhoven, E. Vlieg, and E. P. A. M. Bakkers, Crystal structure transfer in core/shell nanowires, *Nano Lett.* **11**, 1690 (2011).
- [15] H. I. T. Hauge, S. Conesa-Boj, M. A. Verheijen, S. Koelling, and E. P. A. M. Bakkers, Single-crystalline hexagonal silicon–germanium, *Nano Lett.* **17**, 85 (2017).
- [16] I. Yang, X. Zhang, C. Zheng, Q. Gao, Z. Li, L. Li, M. N. Lockrey, H. Nguyen, P. Caroff, J. Etheridge *et al.*, Radial growth evolution of InGaAs/InP multi-quantum-well nanowires grown by selective-area metal organic vapor-phase epitaxy, *ACS Nano* **12**, 10374 (2018).
- [17] J. Johansson, C. P. T. Svensson, T. Mårtensson, L. Samuelson, and W. Seifert, Mass transport model for semiconductor nanowire growth, *J. Phys. Chem. B* **109**, 13567 (2005).
- [18] A. S. Myerson and R. Ginde, Crystals, crystal growth, and nucleation, *Handbook of Industrial Crystallization* (Elsevier, Amsterdam, 2002), pp. 33–65.
- [19] D. Kashchiev, Two-dimensional nucleation in crystal growth: thermodynamically consistent description of the nucleation work, *J. Cryst. Growth* **267**, 685 (2004).
- [20] R. P. Sear, Nucleation: theory and applications to protein solutions and colloidal suspensions, *J. Phys.: Condens. Matter* **19**, 033101 (2007).
- [21] F. Rovaris, W. H. Peeters, A. Marzegalli, F. Glas, L. Vincent, L. Miglio, E. P. Bakkers, M. A. Verheijen, and E. Scalise, 2H–Si/Ge for group-IV photonics: on the origin of extended defects in core–shell nanowires, *ACS Appl. Nano Mater.* **7**, 9396 (2024).
- [22] E. Bakkers and W. Peeters, Growth rate of hexagonal SiGe multi quantum wells [dataset], Zenodo (2024), <https://doi.org/10.5281/zenodo.14525920>.