Wurtzite-Phased InP Micropillars Grown on Silicon with Low Surface Recombination Velocity

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ABSTRACT: The direct growth of III–V nanostructures on silicon has shown great promise in the integration of optoelectronics with silicon-based technologies. Our previous work showed that scaling up nanostructures to microsize while maintaining high quality heterogeneous integration opens a pathway toward a complete photonic integrated circuit and high-efficiency cost-effective solar cells. In this paper, we present a thorough material study of novel metastable InP micropillars monolithically grown on silicon, focusing on two enabling aspects of this technology—the stress relaxation mechanism at the heterogeneous interface and the microstructure surface quality. Aberration-corrected transmission electron microscopy studies show that InP grows directly on silicon without any amorphous layer in between. A set of periodic dislocations was found at the heterointerface, relaxing the 8% lattice mismatch between InP and Si. Single crystalline InP therefore can grow on top of the fully relaxed template, yielding high-quality micropillars with diameters expanding beyond 1 μm. An interesting power-dependence trend of carrier recombination lifetimes was captured for these InP micropillars at room temperature, for the first time for micro/nanostructures. By simply combining internal quantum efficiency with carrier lifetime, we revealed the recombination dynamics of nonradiative and radiative portions separately. A very low surface recombination velocity of 1.1 × 10^3 cm/sec was obtained. In addition, we experimentally estimated the radiative recombination B coefficient of 2.0 × 10^-10 cm^3/sec for pure wurtzite-phased InP. These values are comparable with those obtained from InP bulk. Exceeding the limits of conventional nanowires, our InP micropillars combine the strengths of both nanostructures and bulk materials and will provide an avenue in heterogeneous integration of III–V semiconductor materials onto silicon platforms.

KEYWORDS: InP nanowire/nanoneedle/nanopillar, silicon, interface, wurtzite, surface recombination velocity

Silicon has been the prevalent backbone for the existing microelectronic infrastructures and the foundation of complementary metal-oxide—semiconductor (CMOS) transistor devices. It is difficult, however, to utilize silicon for optoelectronic devices because it is an indirect bandgap material. Direct bandgap materials that exhibit superior efficiencies in optical processes are traditionally group III–V compound semiconductors, which on the other hand, are not as cost-effective as silicon. Integrating III–V semiconductors onto silicon therefore can pave the way toward a highly compact optoelectronic platform combining the strengths of both materials. Bottom-up growth of III–V materials on silicon substrate provides a possible pathway toward this goal. However, it remains highly challenging due to the fundamental incompatibilities mainly in lattice mismatch and synthesis temperatures between these two materials. As semiconductor nanostructures have been the key enablers in a variety of applications, their unique properties help bypass these roadblocks and open up a new landscape in optoelectronics. Tomioka and Fukui et al. have provided comprehensive reviews of III–V nanowires on silicon substrate, including pioneer work on metal-catalyzed vapor–liquid–solid growth, catalyst-free growth, as well as selective-area growth. However, these methods of growing nanowires require either the use of gold catalyst, or relatively high growth temperature (usually >550 °C), which are both CMOS postprocess incompatible. In addition, the existence of critical dimensions, leading the diameters of these nanowires to be well below 200 nm in general, results in poor optical confinement and large surface to volume ratio, thus hampering their optoelectronic performance. Thereby, a complete photonic integrated circuit (PIC) has been an unfulfilled feat with traditional nanowires grown on silicon.

Our group has reported InGaAs-based and InP-based nanoneedle/nanopillar/micropillar structures integrated on single and polycrystalline silicon substrates, synthesized by low-temperature (400–450 °C) metalorganic chemical vapor deposition (MOCVD) without external catalysts. The
growth condition was verified to be CMOS-postprocess-compatible. The unique core–shell growth mode allows the base diameters of micropillars to scale far beyond lattice-mismatched critical dimensions while still preserving single crystalline quality in the body. In the text below, we refer to the pillars with diameters above or close to 1 μm as “micropillars”. Being able to scale up in size overcomes the long-standing drawback of nanowires whose large surface-to-volume ratio induces severe surface recombinations. The large size empowers the optical performance of our micropillars vastly, eclipsing traditional nanowires in optoelectronic functionality. This integrated material template has enabled high-performance optoelectronic devices for light emission, light detection, and photovoltaics. An integrated optical link was also demonstrated, showing the potency of the micro-/nanostructures for optical interconnect with higher speed and lower power consumption than their traditional electrical counterparts. This can be crucial, to the increasing bandwidth demands of communication network. In the rapidly growing field of photovoltaics, cost is becoming the major competition factor. Although III–V nanowires offer strong light absorption and potential for high energy conversion efficiency, the use of native substrates is yet very expensive. Integration of III–V micro- and nanowires on silicon substrate incurs minimal real estate cost while still offering full functionality and, thus, promises the realization of high-efficiency and cost-effective solar cells.

Located at the center of “Energy Gap versus Lattice Constant” chart for semiconductors, InP is an ideal candidate as a growth template to bridge silicon and alloys of compound semiconductors. Hence, InP micropillars grown on silicon are particularly suitable for developing communication wavelength on-chip laser sources, as well as multijunction high efficiency solar cells. Most importantly, with a very low surface recombination velocity (SRV), InP-based nanostructure is less susceptible to recombination at free surfaces. Being able to scale the single crystalline structures with diameter exceeding 1 μm further facilitates high photon luminescence yield and outstanding optical characteristics. Given the superior material quality, a thorough material characterization is necessitated in order to understand the growth mechanism and to better harness it. In this paper, we aim at answering the following questions through various characterization techniques: (1) how InP grows on silicon and how stress is relaxed around the heterointerface; (2) how InP micropillar remains single crystalline when its diameter is far beyond the critical value; (3) how the optical characteristics are improved owing to the larger size of these micropillars compared to conventional nanowires; (4) how surface recombination velocity is characterized and how it compares with the reported values for bulk InP.

Growth of InP nanostructures studied here occurred at 450–460 °C for 15 min via MOCVD on a (111)-silicon substrate (refer to ref 23 for details), resulting in nano- and micropillars with tapered sidewalls (size scalable by growth time). The substrate was deposited with SiO2 and exposed with photolithography, the surface was then chemically treated with TMAH (tetramethylammonium hydroxide). Inside the exposed areas, the pillars are randomly distributed. As shown in the large scale scanning electron microscopy (SEM) image in Figure 1a, both upright and slanted micropillars stem along the equivalent (111) directions of silicon. We obtained at least one upright InP pillar per 40 μm2 area. Because each single micropillar is large enough, it will be processed and used as a single device, as evident by the optically pumped laser. Hence, different from the cases where an ensemble of nanowires would be required to function as a laser, or a single nanowire needed to be removed from the native substrate and optically pumped from the side, high density may be neither necessary nor desirable. However, periodic and precise-position-controlled...
growth is indeed more desired and is the next goal to work toward, upon which better control of pillar orientation can also be realized. Despite the stochastic nature, our nanostructures can be repeated for more than 500 runs. This reproducibility signifies the potential of this work as an important pathway for heterogeneous integration. Figure 1b exemplifies a micropillar grown at 450 °C, which possesses a diameter of about 1 μm, and height of ∼8.9 μm. The pillars assumed a hexagonal pyramidal shape, similar to our previously reported InGaAs-based structures grown on silicon. However, there is a distinct difference between the growth of InGaAs and InP micropillars. In the former case, a continuous polycrystalline film is found to cover the entire substrate and embrace the base of the single crystalline pillars. In fact, this poly-InGaAs layer grows simultaneously with the InGaAs pillars and prevents the very bottom part of the pillars from expanding. As the growth proceeds, an inversely tapered "root" forms at the bottommost portion of the pillar. Such tapering effectively shrinks the exact InGaAs/Si contact area to ∼60 nm in diameter, thus reducing the strain energy accumulating within the III–V structure.25 In the growth of InP pillars, however, no continuous poly-InP layer is observed.23 This can be attributed to the long diffusion length of indium adatoms, as well as the large lattice mismatch between InP and silicon. Compared to InGaAs (lattice mismatch ∼6% with Si), InP experiences a much higher misfit strain (∼8%) with the heterogeneous substrate. The much higher elastic energy due to larger lattice mismatch in the latter can effectively inhibit the wetting of InP on silicon. In addition, the presence of Ga in the ternary compound also changes the wetting behavior of the material. Because Ga has a much shorter diffusion length than indium, we believe that Ga settles down readily on the substrate without much clustering and acts as nucleation sites for the subsequent nanostructure and polycrystalline layer growth. Indium, which can diffuse much longer distances, tends to form large and sparsely distributed clusters before anchoring to the substrate for pillar nucleation. These two factors result in the formation of continuous layer in InGaAs but not in InP. In the absence of a poly-InP layer wrapping around, the entire pillar base is in direct contact with the substrate. With a lattice mismatch as high as 8% between InP and silicon,19 a large footprint implies a huge amount of elastic energy built up in the structure. A stress relaxing mechanism must exist around the interface such that the InP pillars can remain single crystalline when the lateral size scales up to close to a micron.

To study how single crystalline InP stems on silicon, we studied the InP/Si interface with high resolution scanning transmission electron microscope (HR-STEM). All the high resolution high angle annular dark field (HAADF) images were taken with the transmission electron aberration corrected microscope (TEAM I) operated at 80 kV at the National Center for Electron Microscopy, Lawrence Berkeley National Lab. In STEM mode, the best resolution can be as small as 0.1 nm. Sample preparation involves the use of focused ion beam (FIB) and micromanipulator to create a thin slice, or a lamella, of the heterointerface. In particular, we chopped off the top part of the InP pillar with ion beam such that only the bottommost ∼400 nm remained. Pt was then deposited onto the frustum to
passivate the heterointerface in the subsequent top-down ion beam milling. These special treatments help minimize curtaining effect (i.e., roughening of the lamella surface) that would have occurred if the pillars were thinned down in full length. Figure 1c shows an SEM image of the pillar cross-section halfway through the milling process. The entire pillar base, with a width of ~900 nm, is indeed in direct contact with silicon without any inverse tapering. The sample was first aligned along the [1100] zone axis of InP. Because ZB [211] looks exactly the same as WZ [1100], any stacking faults would be invisible along this orientation. This simplifies the identification of any misfit defects that may appear at the interface. Figure 1d displays an HAADF image of the InP/Si interface. Because indium has a much larger atomic number than silicon, InP appears to be much brighter than silicon in the z-contrast image. We note that the heterointerface is a little blurry due to the nanoscopic roughness induced by TMAH etch prior to the growth to facilitate pillar nucleation. The lattice continues seamlessly from silicon to InP in spite of the blurriness at the interface. In particular, silicon is aligned with InP in a way that Si [211] || InP [1100] and Si [111] || InP [0001]. These observations indicate that InP pillars do nucleate directly on silicon without any amorphous materials in between. Such direct contact is a necessary condition for good electrical conductivity across the heterointerface, which is crucial to fabricating real devices on silicon.

Information in the reciprocal space can reveal the strain state of InP close to the heterointerface. Figure 1e displays the fast Fourier transform (FFT) of the HAADF image in Figure 1d. Two sets of diffraction spots, belonging to InP and silicon, can be clearly seen. Diffractions from InP (wurtzite lattice) and silicon (diamond lattice) are registered with 4-digit Bravais—Miller indices and 3-digit Miller indices, respectively. In particular, the diffractions of InP 1122 and Si 131 are collinear with the transmitted beam at the center. This indicates that InP and silicon are relaxed to their own lattice spacing rather than strained to accommodate for the lattice mismatch elastically. To find out the origin of the stress relaxation mechanism, we applied filtering in the spatial frequency domain such that only the InP 1120 and Si 022 diffraction pairs remain. Figure 1f illustrates the corresponding filtered image in real space, revealing vertical (1120) planes in InP and (022) planes in silicon only. Most of the InP (1120) planes in the upper half are seen to stem seamlessly from the Si (022) planes below. However, there exist some locations at which one InP (1120) is connected to two Si (022), as indicated by red arrows. These are in fact misfit dislocations that propagate along the [1100] direction. When connected together, the five misfit dislocations outline the contour of the silicon surface, attesting to the roughening effect of TMAH etch during sample preparation. Moreover, the planes in the vicinity of the misfit dislocations are considerably bent, revealing the presence of strain in these regions. Interestingly, these dislocations appear periodically in every 12th or 13th InP (1120) planes. On average, 27 Si (022) planes are matched with 25 InP (1120) planes in the presence of dislocations. The ratio of the two numbers, 1.08, accounts for exactly 8% lattice mismatch between silicon and InP. We believe that these periodic misfit dislocations effectively relax the misfit stress between the two materials leading to the high-quality growth in the bulk material of the pillar that is remote from the heterointerface.

While misfit dislocations can be clearly identified along [1100], stacking disorders can only be seen along [1120]. Figure 2a displays a HAADF image of the interface taken along the [1120] zone axis. InP is observed to grow directly on silicon along the entire 90 nm interface, although the border is a bit uneven due to nanoroughening during TMAH etch. Figure 2b shows the heterointerface at higher magnification. Immediately above the InP/Si interface is a transition region composed of ZB-phased crystal with stacking faults and twinings. InP then stabilizes into WZ phase above this 4 nm transition region. We attribute the formation of stacking disorders to the very high growth rate during the initial nucleation stage. The high deposition rate possibly inhibits the adatoms from migrating to the most energy favorable sites before incorporating into the lattices. When the growth slows down and stabilizes later on, the large surface-to-volume ratio favors the formation of WZ lattice, which subsequently becomes the dominant crystal phase in the micropillar growth. Figure 2c displays the zigzag arrangement of the pure WZ InP lattice under high magnification, resolving individual atomic columns clearly in the binary compound. Because indium is much heavier than phosphorus, the former appears much brighter and larger in size in the z-contrast image. Nevertheless, one can still observe the polarity clearly from the highly asymmetric dumbbells; the tiny phosphorus atom is sitting on top of the gigantic indium atom in the pair. This configuration implies that the top facet of the pillar is terminated with group V. A group V terminated surface is believed to be more stable and is consistent with what we observed in InGaAs nanopillars and other nanowire work. To further confirm the crystal phase purity, we performed microphotoluminescence (μ-PL) on single pillars by delivering continuous-wave (CW) excitation from a 660 nm diode laser to individual as-grown InP pillars (see ref 35 for more details). As displayed in the spectrum in Figure 2d, InP micropillar demonstrates a bandgap of 1.41 eV at room temperature (298 K), which is about 70 meV larger than that of ZB-phased InP bulk substrate (1.34 eV). Remarkably, pure WZ phase is very hard to achieve for InP nanowires, except for a few cases reported on InP substrate with relatively high growth temperatures. When growing on a lattice-mismatched substrate under a CMOS-postprocess-compatible temperature, it is extra challenging. These metastable microstructures therefore provide lots of possibilities for exploration of new physics as well as unprecedented functionalities.

In addition to the crystalline quality inside the micro-/nanopillar body, our focus is shifted to the pillar surface to study another crucial factor impacting the performance of optoelectronic materials. Surface quality has been a longstanding problem for nanostructures due to their large surface-to-volume ratio and thus propensity for more severe nonradiative recombinations at the surface states. Surface recombination velocity (SRV) is historically used to evaluate the surface quality of materials. Its importance sparks great interest in the realm of InP nanowires, with SRV values reported via different characterization methods. However, the precision of these values suffers from undesirable effects induced by polytypism and oversimplification through some level of estimations. Here, we demonstrate a new but straightforward methodology to characterize the surface quality of InP micro/nanopillars as-grown on silicon, and compare our results with the well-known values of InP bulk.

Carrier recombination lifetime \( \tau \) consists of a radiative portion \( \tau_r \) and a nonradiative portion \( \tau_{nr,d} \) defined as

\[
\tau^{-1} = \tau_{nr}^{-1} + \tau_r^{-1} = \text{SRV} \times (\text{SA: V}) + B \times N
\]
where SA:V is the surface area to volume ratio, $B$ is defined as the radiative recombination coefficient, and $N$ stands for excess carrier concentration. The overall effective lifetime $\tau$ can be measured directly with time-resolved photoluminescence.
At room temperature, nonradiative recombination dominates such that $\tau_{nr}$ is much shorter than $\tau_r$. Hence with $\tau \approx \tau_{nr}$, SRV is usually estimated as $(\text{SA:V} \times \tau)^{-1.26}$ Although neglecting radiative recombination induces inaccuracies in calculating SRV, this estimation is still widely adopted because there was no obvious way to separate $\tau_{nr}$ and $\tau_r$. Here, we propose a straightforward method to extract $\tau_{nr}$ and $\tau_r$ from $\tau$, by simply considering another equation of these two unknowns. It is worth noting that this method was also proposed in ref 48 for another material system and purpose, during the time when our study was being conducted. Also, very recently, Gao et al. proposed a different method to characterize IQE and $\tau$, which requires a steady-state condition. Another relevant study under a pulsed excitation condition was also conducted and would be of interest here as well.

Figure 3a lays out the experimental setup used to implement this method. It consists of the aforementioned $\mu$-PL setup (blue background), in conjunction with time-correlated single-photon counting (TCSPC) facility (pink background) for TRPL measurement (see ref 35 for details). In order to calculate SRV, carrier recombination lifetime $\tau$ is first obtained from TRPL measurement at 298 K, as shown in Figure 3b, highlighting the impulse response of the system. Under each pump power, the “effective” internal quantum efficiency (IQE) is then evaluated within the same time window (12.5 ns) by comparing the PL intensities (integrated across the whole spectrum from Figure 3d) at 298 K to that at 4 K, displayed in Figure 3e. Assuming 100% of IQE is reached at 4 K under each pump power ($P$), IQE at room temperature is defined as $I_{298K}(P)/I_{4K}(P)$. It is worth noting that the exact same pump conditions need to be used for Figure 3c and e, which is critical for our method here.

In Figure 4a, we show the dependence of the overall carrier lifetime $\tau$ as a function of average pump power density. In the low injection regime, nonradiative recombination is the dominant factor. As the carrier density increases, more of the nonradiative recombination centers become saturated, $\tau$...
therefore increases rapidly with pump power in this regime. At higher carrier injection level, however, \( \tau \) becomes dominated by radiative recombination whose lifetime decreases with carrier concentration. Notably, this two-phase dependency has only been reported for InP bulk material.\(^{32}\) The increase of lifetime in the low injection region was usually out of reach for nanostructures due to the low photoluminescence yield at room temperature. The key development enabling the capture of data lies in both the bigger size of our pillars and the compact \( \mu \)-PL setup with high collection efficiency. Obtained under the same experimental conditions, the power-dependence of IQE is depicted in Figure 4b. Theoretically, IQE can be expressed as

\[
\text{IQE} = \frac{1}{\tau} - \frac{1}{\tau_r} = \frac{1}{\tau_n} + \frac{1}{\tau_f}
\]

Combining eqs (1) and (2), the two unknowns \( \tau_n \) and \( \tau_r \) can be resolved, revealing the fascinating physics underlying. The extracted \( \tau_n \) (red) and \( \tau_r \) (green) are plotted against average pump power in Figure 4c. The effective lifetime \( \tau \) (blue), the same as that shown in Figure 4a, is also included in Figure 4c for easier comparison. It can be observed that at low injection level, \( \tau_n \) is almost equal to \( \tau \) and increases until all nonradiative recombination centers are saturated at around 0.5 kW/cm\(^2\), where \( \tau \) is about 1.0 ns. As carrier injection gradually rises, \( \tau \) starts to drop, whereas \( \tau_n \) stays at around 1.2 ns. This validates that radiative recombination should not be neglected especially at higher carrier concentration.

On the other hand, while appearing much longer than \( \tau_n \), \( \tau_r \) always decreases with carrier concentration based on the relation \( \tau_r^{-1} = B \times N \) for undoped materials. By plotting \( \tau_r^{-1} \) as a function of \( N \) in Figure 4d, the radiative recombination \( B \) coefficient can be obtained through linear fitting in the high injection regime. We note that the carrier concentration is the most critical for the fitting of \( B \) coefficient. We adopted the method of estimating carrier concentration from Fermi level split, which was generated directly from PL intensity instead of pump power.\(^{53}\) With a pulsed pump laser, the carrier concentration calculation here represents an averaged condition (see Supporting Information). The fitting result is just a preliminary estimation, and further work needs to be done as to improve the accuracy. Nevertheless, a value of \( 2.0 \times 10^{10} \) cm\(^3\)/sec is attained, which has never been reported experimentally for WZ-phased InP before. The fact that this is within the realm of the state-of-the-art numbers reported for ZB-phased InP before. The increase of lifetime at the saturation point for each trace in Figure 5a, we plot the corresponding TRPL decay curves, displayed in Figure 5b. The effective carrier lifetime \( \tau \) of each trace is obtained by single exponential decay function fitting. We took full set of data as in Figure 5a–b for around 60 pillars, the lifetime spread is summarized as below. For pillars grown at 450 °C, \( \tau = 0.98 - 1.15 \) ns with an average of 1.03 ns. For pillars grown at 455 °C, \( \tau = 1.30 - 3.36 \) ns with an average of 2.01 ns. And for pillars grown at 460 °C, \( \tau = 3.59 - 8.58 \) ns with an average of 5.55 ns. The table in Figure 5c summarizes the important parameters, including \( \tau_n \), \( \tau_r \), and SRV, of the example pillars grown at different temperatures studied in Figure 5a–b. It is worth noting that the lifetime of micropillars synthesized at 460 °C can be as long as 7.4 ns, with the nonradiative portion calculated as 10.4 ns. This is the longest lifetime detected at room temperature for InP micro/nanostructures, to the best of our knowledge. Moreover, the SRV of the micropillars can reach ~1.1 × 10\(^{10}\) cm/sec, which is favorably comparable with the reported value in InP bulk material.\(^{35,57}\) The exciting results verify that the WZ-phased InP micropillars studied here are endowed with superior surface quality while still possessing the unique advantage of nanostructures, which is of paramount importance for various optoelectronic devices.

Nonradiative recombinations occur mainly at crystal defects and surface states. Embodied by the HR-TEM image in Figure 1f, misfit dislocations are confined at the InP/Si interface where the lattice-mismatch strain is relaxed, leading to excellent material quality in the bulk part remote from the heterointerface. When we broke the root part of micropillars from as-grown silicon substrate and transferred them to a foreign silicon substrate, a remarkable increase in carrier lifetime was observed. Figure 5d exemplifies that after being transferred, the lifetimes of micropillars grown at 450 °C can be doubled compared with that of a typical as-grown micropillar (1.0 ns). Such drastic improvement proves the high concentration of nonradiative recombination centers, that is, misfit dislocations, is indeed well confined in the root part of micropillar and at the InP/Si interface. Carrier lifetime also shows great improvement after surface passivation with wet chemical etching, that is, HCl (2M):H\(_2\)O:citric acid (0.5M) = 1:1:0.5.\(^{58}\) For micropillars grown at 455 °C, lifetimes before and after surface passivation were compared for the same pillar, showcasing in Figure 5e an improvement from 1.9 to 2.9 ns. Combining these two experiments, we conclude that the nonradiative recombination takes place both at the defect centers inside the pillar body and at free surface outside the pillar. Therefore, a more rigorous form of Equation 1 should be written as

\[
\tau^{-1} = \tau_n^{-1} + \tau_r^{-1} = \tau_{body}^{-1} + \tau_{surf}^{-1} + \tau_r^{-1} = \tau_{body}^{-1} + \text{SRV} \times (\text{SA}: V) + B \times N
\]

where \( \tau_{body} \) and \( \tau_{surf} \) separate the effects from nonradiative recombinations inside the material body and outside at the surface. From this new form we can tell that the method in our previous context underestimated the value of \( \tau_{surf} \) and thus
overestimated SRV. The real SRV values should be even smaller than the results reported above. In our attempt to quantify $\tau_{\text{body}}$ by measuring nanopillars with different diameters as reported in ref 38, however, similar relation was not found between $\tau_{\text{a}}$ and $d$, indicating $\tau_{\text{body}}$ might also be changing with the diameter. A more complete method to further separate $\tau_{\text{body}}$ from $\tau_{\text{surf}}$ in order to characterize SRV even more accurately is beyond the scope of the present work, and remains a topic for future investigation. However, we observed that SRV values of micropillars grown at 460 °C can be improved by neither mechanical removal nor chemical surface passivation. Thus, the SRV of $\sim$1.1 × 10^3 cm/sec exhibits a good evaluation of material surface quality for our case here. Meanwhile, this again agrees with our hypothesis that higher growth temperature not only reduces the concentration of misfit defects inside micropillar body, but also anneals the surface. Further increasing the growth temperature is also not of interest here, because a lower growth temperature is desired for a more flexible CMOS postprocess thermal budget. 18 All of our optoelectronic devices were based on InP pillars grown at 450 °C, which have already shown excellent performances. 30,31,35

In conclusion, we have presented comprehensive material and optical studies dedicated to answering the questions listed in the beginning of this paper. Extensive electron microscopy studies reveal that periodic dislocations exist at the InP/Si heterointerface, relaxing the 8% lattice mismatch between InP and silicon. High-quality InP pillars exhibiting excellent optical efficiencies therefore can evolve on the mismatched substrate. A straightforward methodology is carried out to characterize the surface quality of the InP micropillars, which is considered as an important figure of merit for micro/nanostructures. The combination of carrier recombination lifetime and IQE values measured at room temperature facilitates the separation of radiative and nonradiative parts of the lifetimes and enables revelation of key information. From nonradiative lifetime, SRV is calculated and found to improve as growth temperature increases. Micropillars grown at 460 °C yields an SRV value of $\sim$1.1 × 10^3 cm/sec, signifying a remarkable surface quality compared with conventional nanowires. Radiative recombination B coefficient is also characterized, for the first time for wurzite-phased InP. Both SRV value and B coefficient are close to the reported values for ZB-phased InP bulk. The above material and optical studies show that the InP micropillars integrated on silicon are endured with advantages of both bulk material and nanostructures. This unprecedented metastable direct growth of InP micropillars monolithically on silicon provides an avenue in the integration of optoelectronics with silicon-based technologies.

# ASSOCIATED CONTENT

* Supporting Information

The Supporting Information is available free of charge on the ACS Publications website at DOI: 10.1021/acs.nanolett.5b02869.

Method to calculate carrier concentration for fitting B constant. (PDF)

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# ACKNOWLEDGMENTS

The work was supported by U.S. DOE SunShot Program (contract DE-EE0005316), DoD NSSEFF Fellowship (contract N00244-09-1-0013 and N00244-09-1-0080), California Advanced Solar Technologies Institute, UC Multicampus Research Program and Institute (MRPI), the Center for Energy Efficient Electronics Science (NSF Award 0939514), Defense Advanced Research Projects Agency (DARPA) University Photonics Research (UPR) (Award HR0011-04-1-0040), Microelectronics Advanced Research Corp (MARCO) Interconnect Focus Center (IFC). Work at the Molecular Foundry (transmission electron microscopy) was supported by the Office of Science, Office of Basic Energy Sciences, of the U.S. Department of Energy under Contract No. DE-AC02-05CH11231. C.C.H. acknowledges support from the Chang Jiang Scholar Endowed Chair Professorship at Tsinghua University, China, and the Li Ka Shing Foundation Women in Science Research Grants. The authors thank Indrasen Bhattacharya, Stephen Adair Gerke, and Weijian Yang for providing valuable suggestions for the manuscript.

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