

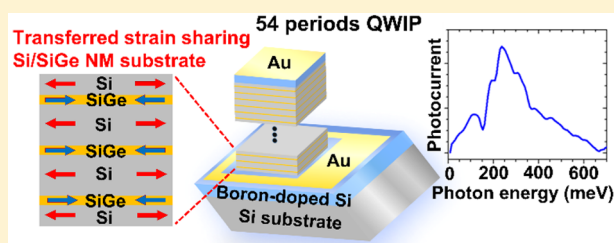
SiGe Nanomembrane Quantum-Well Infrared Photodetectors

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Supporting Information

ABSTRACT: SiGe quantum wells are promising candidates for the development of intersubband light emitters and photodetectors operating at mid- and far-infrared wavelengths. By virtue of their inherent compatibility with the Si microelectronics platform, these devices may be integrated seamlessly within complex optoelectronic systems for sensing and imaging applications. However, the development of high-quality SiGe intersubband active layers is complicated by the large lattice mismatch between Si and Ge, which limits the number of quantum wells that can be grown on bulk Si substrates before the onset of structural degradation due to inelastic strain relaxation. To address this issue, we investigate the use of lattice matched growth templates consisting of quantum-well nanomembrane stacks that were at one point free-standing, allowing for the internal stress to be relaxed via elastic strain sharing rather than defect formation. SiGe quantum-well infrared photodetectors (QWIPs) based on this approach are developed and characterized. Efficient current extraction from these ultrathin devices is obtained by bonding the nanomembranes directly on a doped Si substrate. Pronounced photocurrent peaks at mid-infrared wavelengths are measured, with improved responsivity compared to otherwise identical devices grown simultaneously on the supporting Si substrate.

KEYWORDS: semiconductor nanomembranes, SiGe/Si quantum wells, intersubband transitions, strain engineering



In the past several years, extensive research efforts have been devoted to the development of optoelectronic devices based on silicon, germanium, and related alloys (the leading materials platform of microelectronics) as a way to enable the large-scale on-chip integration of electronic and photonic functionalities in a CMOS compatible process. As a result of these activities, the use of silicon photonics for data communications at fiber-optics transmission wavelengths is already entering commercialization. At the same time, group-IV semiconductor devices operating at longer wavelengths (into the mid- and far-infrared spectral regions) are also being explored for the development of highly integrated systems for biochemical sensing and thermal imaging.¹ While Si and Ge are transparent over a wide portion of the mid/far-infrared spectrum, they can be used for active optoelectronic devices at these wavelengths using intersubband (ISB) transitions in quantum wells (QWs).²

Early work in this context has focused on SiGe quantum-well infrared photodetectors (QWIPs),^{3–8} that is, unipolar photoconductive devices where ISB absorption is used to promote carrier escape out of the QWs. Intersubband electroluminescence has also been reported,^{9–13} following various theoretical proposals,^{14,15} as a promising first step toward the demonstration of SiGe quantum cascade lasers. The potential of terahertz light emission^{11,13} is particularly significant in these group-IV semiconductors, owing to their nonpolar nature.

Specifically, existing GaAs THz quantum cascade lasers are intrinsically limited to cryogenic operation and incomplete coverage of the THz spectrum, due to thermally activated ISB nonradiative decay and Reststrahlen absorption caused by polar optical phonons.¹⁶ In SiGe, such electron–phonon and photon–phonon interactions are significantly weaker, so that, in principle, both limitations can be overcome. However, progress in the development of group-IV ISB devices has so far been hindered by the large lattice mismatch (4.2%) between Si and Ge and the resulting formation of extensive strain-induced defects in SiGe QWs.² This issue is especially problematic in quantum cascade lasers, which require a large number of QWs with sufficiently different well and barrier compositions to provide strong quantum confinement for the carriers involved in the laser transitions. Specifically, these devices require sufficiently deep QWs to avoid any significant thermal escape from the upper laser subbands into the unconfined states over the barriers. When these structures are grown epitaxially on a rigid Si substrate, the accumulated strain eventually relaxes plastically through the formation of misfit dislocations and other structural defects, causing severe degradation in the electrical and optical properties of the QWs.²

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In the present work, we describe the use of “lattice-matched substrates” consisting of elastically relaxed nanomembrane (NM) stacks,¹⁷ as a way to address this strain accumulation problem for the growth of high-quality SiGe/Si multiple-QW structures. The lattice-matched substrates are fabricated by first growing a few periods of the desired SiGe/Si/SiGe/Si... QW structure (with overall thickness well below the threshold for plastic relaxation) on a Si-on-insulator (SOI) wafer. The resulting epitaxial stack is then released from the handle wafer by etching the underlying SiO₂ layer, to form a free-standing NM.¹⁸ In the process, the compressive strain in the SiGe layers caused by the lattice mismatch between Si and SiGe is partially relaxed through the introduction of tensile strain in the Si layers, such that the global average strain in the stack becomes zero. A wet transfer process followed by an anneal step is then employed to transfer and bond this elastically relaxed NM to a new Si host wafer. Finally, the same NM is used as the substrate for the subsequent growth of an arbitrary number of additional QW periods. If these QWs are identical to those in the released and elastically relaxed substrate, they develop the same strain conditions (tensile in the Si layers and compressive in the SiGe layers), so that no net strain accumulates in the entire epitaxial film. Under these conditions, the sample growth is therefore fully strain compensated. In the initial heterostructures fabricated with this approach,¹⁹ we demonstrated the growth of high-quality QWs, and measured strong far-infrared ISB absorption with line widths considerably narrower than those for similar QWs grown on rigid Si substrates, indicative of improved crystalline structure. Here we report the development of the first electrically addressed ISB devices based on this approach (specifically strain-compensated NM QWIPs) and show how the favorable growth conditions enabled by the lattice-matched NM substrates lead to improved device performance.

An important challenge related to these devices is how to achieve efficient vertical current injection into and extraction from such ultrathin NMs. In typical multiple-QW devices, the active region is grown on a sufficiently thick (a few 100 nm) highly doped contact layer, which is then electrically addressed either through a conductive substrate or via etching and metallization from the top side. However, such contact layers cannot be incorporated in the NMs under study without at the same time impeding the elastic strain relaxation that is needed for the subsequent strain-compensated growth of the multiple-QW active region. The approach that we have developed to address this challenge is to transfer and bond the elastically strain-relaxed NM stack that will serve as the growth substrate onto a highly doped Si substrate, and then rely on charge transport across the NM-Si interface. The SiGe mid-infrared QWIPs developed in the present work provide a relatively simple ISB device structure to investigate the resulting electrical properties. Our measurement results indicate that the bonded NM-Si interface allows for unobstructed current injection/extraction, and does not seem to affect the device characteristics in any significant way. Furthermore, pronounced photocurrent peaks are obtained, with improved responsivity compared to otherwise identical devices grown simultaneously on the supporting Si substrate.

RESULTS AND DISCUSSION

The QWIP active-material design is illustrated schematically in Figure 1. Similar to most prior work on SiGe ISB devices, hole ISB transitions are employed in this structure because of the

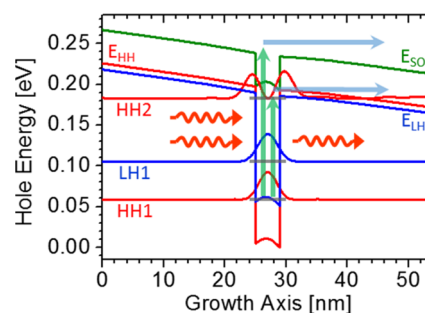


Figure 1. Calculated valence-band lineup of the QWIPs developed in this work under bias and squared envelope functions of the lowest-energy bound states. The gray horizontal lines show the energies at the bottom of the respective subbands. The green vertical arrows indicate hole transitions that define the main mechanisms for photocurrent generation (blue horizontal arrows). The red wavy arrows represent photons flowing through the device. The vertical axis corresponds to hole (rather than electron) energy. As shown, the wells are 4 nm thick.

exceedingly small conduction-band offsets of SiGe/Si QWs.² In passing, we note that, because of several potential advantages including larger oscillator strengths and longer nonradiative decay lifetimes, electronic ISB transitions in the *L* valleys of Ge/SiGe QWs have also started to gain attention in recent years.^{20–23} The development of these QWs has so far been limited by the lack of high-quality SiGe substrates with sufficiently high Ge concentration (>80%) and, therefore, could also benefit strongly from the use of elastically strain-relaxed NM substrates (initially grown on Ge-on-insulator wafers to obtain the required high Ge content).

The QWIP active material described in the present work consists, nominally, of 4 nm thick Si_{0.73}Ge_{0.27} wells separated by 50 nm thick Si barriers grown along the (001) direction. The first and last 2 nm thick sections of each Si barrier (immediately adjacent to the neighboring SiGe wells) are doped *p*-type with boron to the level of $5 \times 10^{18} \text{ cm}^{-3}$. In this structure, the Ge content of the well layers is large enough to produce substantial lattice mismatch with a rigid Si substrate. At the same time, the average Ge concentration over each period is relatively small (~2%), because of the thick Si barriers used, following standard QWIP design rules, to avoid any dark current due to interwell tunneling.²⁴ In any case, the structural and optical characterization results presented below both indicate improved materials quality of the NM devices compared to identical QWIPs grown simultaneously on a rigid Si substrate. The valence-band lineup of one period of this device structure under bias is shown in Figure 1, where E_{LH} , E_{HH} , and E_{SO} indicate the band edges of the light-hole (LH), heavy-hole (HH), and spin–orbit–split-off (SO) valence bands, respectively. Also plotted in the figure are the squared envelope functions of the lowest HH and LH bound states referenced to their respective energy levels (indicated by the gray horizontal lines), which were computed using a 6×6 Luttinger–Kohn model²⁵ coupled to a Poisson-equation solver.

The device operation principles are illustrated schematically by the arrows in Figure 1. At the nominal doping level of this structure, a significant hole population is obtained only in the ground-state subband of each QW (denoted HH1 in the figure). Photocurrent can then be produced via two distinct ISB absorption processes, indicated by the vertical arrows. First, holes can be promoted from HH1 into the first excited HH subband (HH2), followed by tunneling into the continuum of

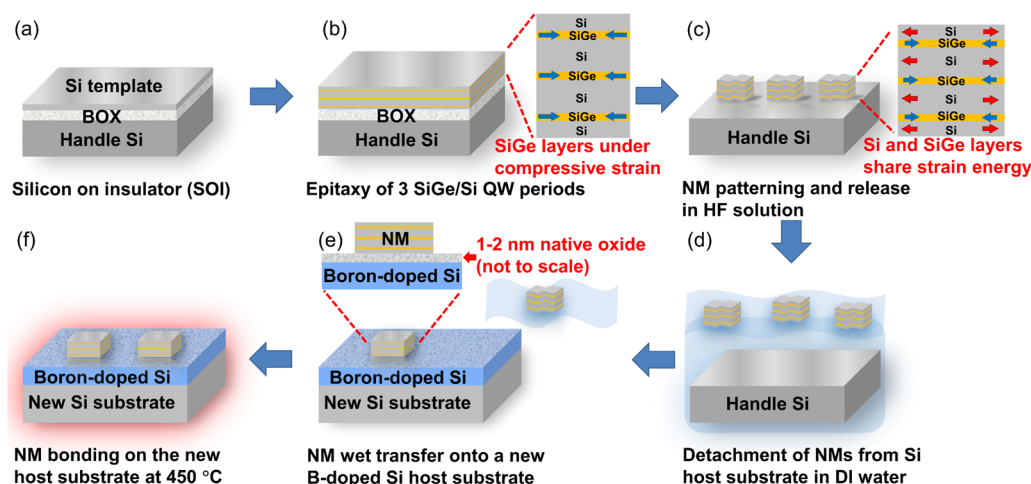


Figure 2. Fabrication of lattice-matched SiGe/Si NM substrates. (a) SOI starting material. (b) Epitaxy of three SiGe/Si QW periods on SOI. The in-plane lattice of the SiGe layers is compressively strained to fit the lattice constant of the SOI Si template. (c) NM patterning by photolithography and release by etching the BOX using HF. Upon release, all Si and SiGe layers partially relax elastically, leaving the SiGe layers less compressively strained and the Si layers slightly tensilely strained, so that the global average strain is zero. (d) Detachment of the released NMs from their host substrate upon dipping in DI water. The NMs float to the water surface. (e) Transfer of the relaxed NM stacks to a new host, a Si substrate with a highly B-doped film and a thin native oxide layer grown via a UV ozone treatment. (f) Anneal to 450 °C to strengthen the bond between the NM and the new host substrate.

unbound states over the adjacent barrier. These bound-to-bound transitions can only couple to TM-polarized light, following the polarization selection rules of ISB absorption.²⁶ The corresponding absorption peak is centered at a photon energy of 122 meV (i.e., at a wavelength of 10.2 μm) based on the simulation results of Figure 1. Second, holes can also be photoexcited from HH1 directly into the unbound HH and mixed LH/SO states over the barriers, via the absorption of photons of sufficiently high energy. Such processes are not strongly polarization dependent, as they involve both ISB transitions between HH states (which are only allowed for TM light) and intervalence-band transitions between HH and mixed LH/SO states (which have larger oscillator strength for TE light).²⁶ The relative magnitudes of these bound-to-bound versus bound-to-continuum contributions to the QWIP photocurrent depend on a well-known trade-off:²⁴ bound-to-bound transitions feature significantly larger absorption strength (due to the larger spatial overlap of the initial and final states), at the expense of reduced escape efficiency (because some of the photoexcited carriers in HH2 may fall back into the ground-state subband before tunneling out of the QWs).

The QWIP active material was grown using low-pressure chemical-vapor deposition (LPCVD). Details are found in Methods. The first fabrication step involves the epitaxy of a small number of the desired SiGe/Si QW periods (three 4 nm thick $\text{Si}_{0.73}\text{Ge}_{0.27}$ wells separated by 50 nm thick Si barriers in the heterostructures described in the current work) on the template layer of a (001) SOI wafer (Figure 2a,b). At these conditions, no structural degradation due to inelastic strain relaxation is expected in the heteroepitaxial film stack. In fact, high-resolution X-ray diffraction (HRXRD) data confirm that the three-period SiGe/Si QW film is coherently grown on the SOI substrate. Next, rectangular pieces of this film stack are completely released from the SOI handle wafer (Figure 2c) using well-established techniques.¹⁸ Specifically, the SOI buried oxide (BOX) layer is dissolved with a wet etch in a hydrofluoric acid (HF) solution, after the NM boundaries are defined by

optical lithography. As the NMs are released from the SOI substrate, their internal stress due to the lattice mismatch between Si and SiGe is elastically relaxed via strain sharing.²⁷ As indicated above, the compressive strain in the SiGe well layers is partially relaxed through the introduction of tensile strain in the enveloping Si layers, until the in-plane lattice constant reaches an equilibrium value that produces zero net strain per period (Figure 2c).

The resulting strain-relaxed NMs are detached from their host substrate upon dipping in deionized (DI) water, where they float to the surface (Figure 2d). Next, they are transferred and bonded onto a 500 nm thick highly *p*-doped ($\sim 1 \times 10^{19} \text{ cm}^{-3}$) Si contact layer previously grown on a separate Si substrate (Figure 2e). Prior to this step, a UV-ozone treatment is used to create a thin native-oxide layer on the target substrate surface, to make it more hydrophilic so as to enable the transfer of large-area NMs without the formation of wrinkles. A 1 h anneal at 450 °C is finally used to improve the NM bonding to the Si contact layer (Figure 2f). Even though the resulting NM/host interfaces contain the aforementioned thin native-oxide layer (about 1–2 nm thick), they are sufficiently conductive to produce devices with electrical characteristics similar to those of otherwise identical samples grown on the same B-doped contact layer, but where the oxide has been removed. Prior detailed measurements of charge transfer across similar Si-NM/Ge interfaces also show that the interfacial layer does not have any significant deleterious effects.²⁸ After a mild cleaning procedure (described in Methods), the Si substrate carrying the NMs is transferred back into the LPCVD chamber where 51 additional QWs are grown under the same conditions as the initial three periods. With this process, flat, wrinkle-free QW stacks are obtained, even without the need for cleanroom conditions for all fabrication steps other than the photolithography.

Because of the aforementioned elastic strain sharing, the QWs grown on the NMs are fully strain compensated so that no net strain accumulates. A similar result could in principle be obtained with compositionally graded SiGe substrates (grown

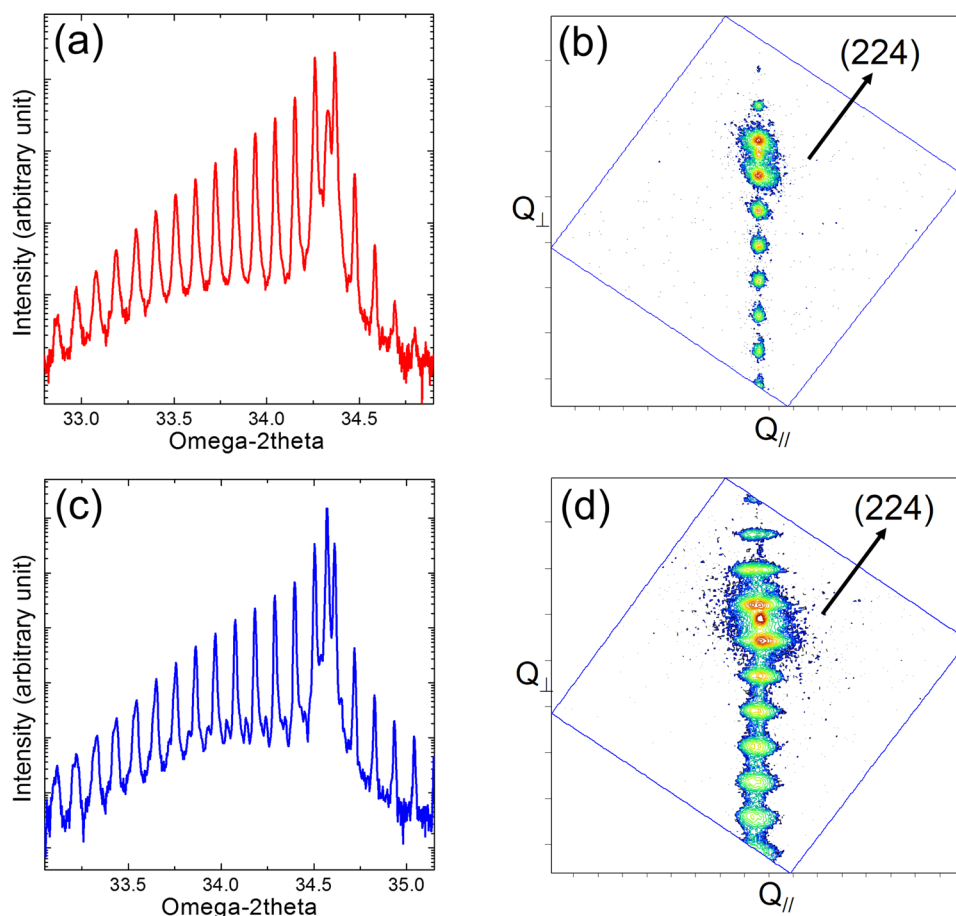


Figure 3. HRXRD results for the QWIP active material grown on a NM-stack substrate transferred to a Si host wafer, and for the same active material grown simultaneously on the host wafer. (a, c) $\omega/2\theta$ scans along the (004) direction measured on the NM and bulk-Si areas, respectively. Many orders of sharp superlattice peaks indicate excellent periodicity of the QWs grown on both areas. (b, d) HRXRD reciprocal-space maps (plotted in the same reciprocal-unit scale) measured along the (224) direction for heterostructures grown on the NM and bulk Si areas, respectively. Note the significant peak broadening observed on the film grown on bulk Si.

on bulk Si), which have also been investigated for the development of SiGe ISB devices.^{7,8,12,29} However, these substrates suffer from a significant density of threading dislocations, as well as mosaic structure and nonuniform lateral strain, all of which are necessarily created during strain grading³⁰ and then propagate through any epitaxial film grown over the substrate. Strain compensated NM substrates are free of such defects and therefore can provide higher crystalline quality,³¹ particularly for the growth of complex multiple-QW systems.

The high structural quality of the samples developed in this work is illustrated by the HRXRD data of Figures 3a,b. Specifically, Figure 3a shows an $\omega/2\theta$ scan measured along the (004) direction from the QWIP active material grown on a NM. Many orders of sharp superlattice peaks are observed, indicative of excellent periodicity of the multiple-QW structure. A HRXRD reciprocal-space map of the same sample measured along the (224) direction is plotted in Figure 3b, showing coherent film growth without plastic relaxation, as indicated by the vertical alignment of the superlattice peaks, with minimal broadening. Fitting of these data with RADS Mercury software also indicates that the layer thicknesses and compositions are near the target values (see Supporting Information, Figure S1). We observe no threading dislocations with optical microscopy or scanning electron microscopy (SEM); however, a few

pinholes per several hundred μm^2 are typically evident (see Supporting Information, Figure S2). Such defects are not caused by lattice-mismatch strain, and in fact are also observed in homoepitaxial films grown in the same chamber. Their origin is likely related to contamination initially present on the substrate surface and not fully removed by the cleaning procedures.

Figure 3c,d show HRXRD spectra measured on the identical QWIP structure grown at the same time on the bulk Si area surrounding the strain-relaxed NMs (with the 500 nm thick heavily B-doped Si contact layer). These data are indicative of a structurally worse film compared to the strain-compensated QWs of Figure 3a,b. Many superlattice orders are again clearly observed from the $\omega-2\theta$ scan, suggesting good periodicity of the film (Figure 3c). However, additional weaker peaks are also evident, located between the main superlattice features. Moreover, the corresponding reciprocal-space map (Figure 3d) shows a significant broadening of the superlattice peaks relative to that measured on the heterostructure grown on the NM (Figure 3b). The details of interpretation of these data are left for a separate paper, but briefly, we believe that a laterally quasiperiodic 3D morphology forms in these heterostructures as a result of the accumulated strain.³²

For the electrical-characteristics and photocurrent measurements, rectangular mesas with lateral dimensions of 300 to 400

μm are fabricated by reactive ion etching through the multiple-QW active material into the doped-Si layer. Metal contacts consisting of a Ti/Au bilayer are then deposited on top of and around each mesa. The resulting device structure is shown schematically in Figure 4a, together with a top-view optical

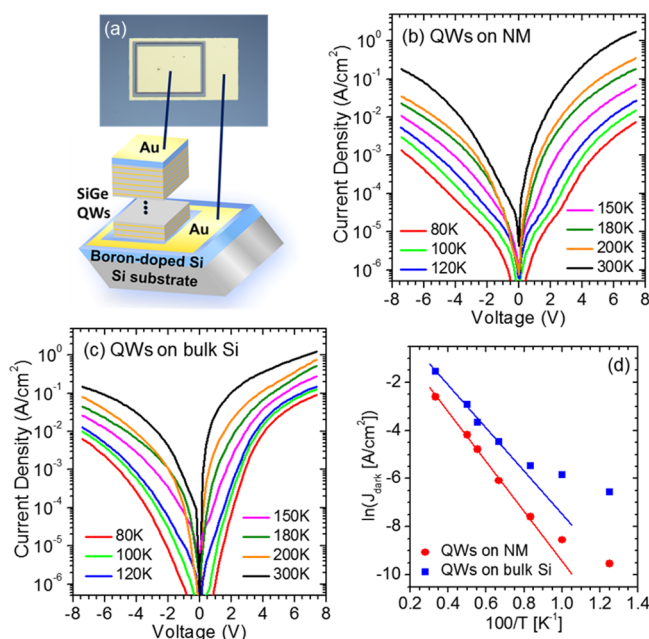


Figure 4. Electrical-characterization results. (a) Schematic device structure (bottom panel) and top-view optical micrograph (top panel) of a QWIP fabricated in this work. (b) Temperature-dependent dark electrical characteristics of a $400 \times 400 \mu\text{m}^2$ device grown on a strain-relaxed NM. (c) Temperature-dependent dark electrical characteristics of a $300 \times 300 \mu\text{m}^2$ device grown directly on the Si contact layer. (d) Arrhenius plot of the dark-current density J_{dark} from (b) (circles) and (c) (squares) at a fixed bias voltage of 3 V. Solid lines: linear fits to the high-temperature (i.e., $T > 120$ K) data.

micrograph. These devices are fabricated simultaneously on the QWs grown over the strain-relaxed NM stacks, and on the QWs grown on nearby regions of the bulk Si substrate. Figure 4b and c show the measured dark-current density J_{dark} versus voltage V of, respectively, a device grown on a NM and a device grown directly on the Si wafer, for several heat-sink temperatures. The shapes of these traces are consistent with standard models of carrier transport in QWIPs,²⁴ including the

observation of a significant decrease in current with decreasing temperature, which indicates predominant thermal activation of the dark carrier transport through the QWs. The asymmetry observed in both figures with respect to the bias polarity may be attributed to some degree of doping migration, as reported previously with GaAs QWIPs.³³ Importantly, the two devices of Figure 4b,c show sufficiently similar electrical characteristics that we can rule out any major detrimental effect from the bonded NM–Si interface in the former sample. In particular, their dark-current densities measured at room temperature and high voltage (i.e., when the transport is least limited by thermal barriers) are nearly the same. The temperature variations of their $J_{\text{dark}}-V$ traces are also qualitatively similar, although with somewhat different rates.

For a more detailed analysis, the symbols in Figure 4d show an Arrhenius plot of the dark-current density of both devices at fixed bias voltage (3 V). In the high-temperature region (i.e., for $T > 120$ K), both traces are in good agreement with a linear fit, as shown by the solid lines in the figure. The activation energies inferred from the slopes of these lines are 89 ± 4 and 77 ± 5 meV for the NM- and Si-substrate devices, respectively. Both values are smaller than the effective barrier height $\Delta E_{\text{B}} = E_{\text{B}} - E_{\text{F}}$ (where E_{F} is the Fermi energy in each well and E_{B} is the HH valence-band edge in the barrier downstream), which is computed to be 125 meV based on the simulation results of Figure 1. This discrepancy suggests a deviation from purely thermionic behavior, which may be quantified with an ideality factor η given by the ratio between the expected and measured activation energies (in analogy with the theory of transport at metal–semiconductor junctions³⁴). A value of $\eta = 1.4$ and 1.6 is obtained, respectively, for the NM- and Si-substrate devices of Figure 4. Given the location of the doping impurities in these samples (in the first and last 2 nm thick sections of each Si barrier), the additional transport mechanism responsible for this deviation may be attributed to thermally activated tunneling through the acceptor impurity levels in the barriers. Finally, at lower temperatures the dark-current densities of Figure 4d exhibit a much weaker temperature dependence, which indicates the presence of some leakage, possibly involving surface states on the mesa sidewalls or defects through the active material. In this respect, the smaller low-temperature dark current density of the NM QWIP compared to the device grown on bulk Si in Figure 4b–d is consistent with the higher structural quality of the strain-compensated QWs grown on the NMs demonstrated in Figure 3.

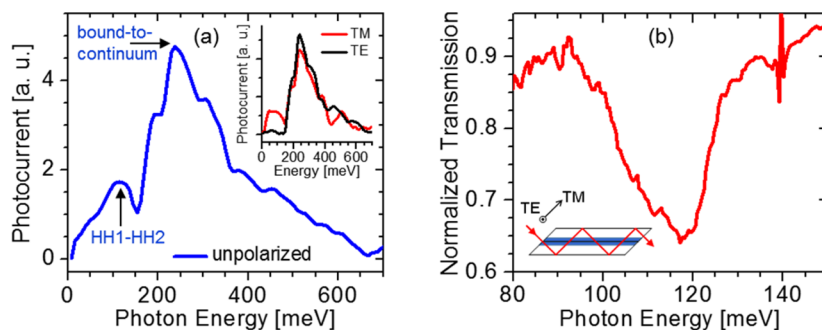


Figure 5. Optical-characterization results. (a) Photocurrent spectrum of a $400 \times 400 \mu\text{m}^2$ device grown on a strain-relaxed NM stack, measured with unpolarized light. The heat-sink temperature and applied voltage are 80 K and 3 V, respectively. Inset: photocurrent spectra of the same device measured with TM- and TE-polarized light. (b) Ratio of the TM- and TE-polarized transmission spectra of the QWIP material. The transmission dip observed in this plot confirms the presence of the HH1–HH2 ISB transition near $10 \mu\text{m}$. Inset: experimental geometry used in this measurement.

The device photocurrent spectra were measured using a Fourier transform infrared spectrometer (FTIR) equipped with a global source, via step-scan phase modulation and lock-in detection. In these measurements, the input light is coupled into the Si substrate at normal incidence through a facet lapped at 45° and eventually reaches the QWIP mesa after a few bounces inside the substrate. This configuration allows probing the HH ISB transitions, which could not be excited by light incident along the mesa surface normal due to their TM-polarized nature. Figure 5a shows the photocurrent spectrum of a NM device at 80 K under an applied voltage of 3 V, measured with unpolarized incident light. As indicated by the arrows, two distinct features can be identified in this spectrum, peaked at about 238 and 115 meV photon energy, which are attributed to bound-to-continuum and HH1–HH2 ISB transitions, respectively. This assignment is consistent with the simulation results of Figure 1, where the HH1–HH2 absorption energy is 122 meV, in close agreement with the peak photon energy of the long-wavelength photocurrent feature. At the same time, the short-wavelength peak extends over a broad range of energies above all the relevant barrier heights, and displays the characteristic asymmetric shape of bound-to-continuum transitions.²⁶

As further evidence of this interpretation, in the inset of Figure 5a we show the photocurrent spectra of the same device under polarized illumination. The long-wavelength feature completely disappears if the input light has TE polarization (i.e., parallel to the plane of the QWs), consistent with the polarization selection rules of HH ISB transitions. In contrast, the short-wavelength feature has comparable amplitudes for TE and TM input polarizations (where TM is linear at 45° with respect to the growth axis in the present experimental configuration). This observation indicates significant contributions to the bound-to-continuum photocurrent from intervalence-band transitions between HH and mixed LH/SO states, as well as ISB transitions between HH states. We have also measured the ISB absorption spectrum of an unprocessed sample (i.e., without any mesa patterned through the QWs) from the same epitaxial material, using a multipass wedge geometry with two opposite facets lapped at 45° , as illustrated in the inset of Figure 5b. Specifically, this sample has an area of about $4 \times 4 \text{ mm}^2$ and contains a single $2 \times 3 \text{ mm}^2$ NM, so that the measured spectra are the result of a spatial average between the areas with and without the NM. In Figure 5b we plot the ratio between its TM- and TE-polarized transmission spectra at 80 K. The transmission dip due to HH1–HH2 ISB absorption is clearly resolved at a photon energy of about 117 meV, in excellent agreement with the photocurrent results of Figure 5a. The sharp cutoff on the high-energy side of this dip is also consistent with the aforementioned effective barrier height $\Delta E_B = 125 \text{ meV}$.

The spectrum of Figure 5a seems to indicate that bound-to-continuum transitions in these devices produce a significantly larger contribution to the photocurrent compared to bound-to-bound transitions. However, it should be noted that the relative magnitudes of these two contributions depend not only on the trade-off between ISB absorption strength and escape efficiency discussed above, but also on the wavelength-dependent transmission losses in the Si substrate. In particular, substantially larger losses can be expected in the spectral vicinity of the HH1–HH2 transition energy due to multiphonon processes in Si,³⁵ as well as free-carrier absorption (which increases with the wavelength squared) in the doped

contact layer. As a result, in measurements where the QWs are illuminated through the substrate (as in Figure 5a), the long-wavelength peak is significantly decreased compared to the bound-to-continuum feature. In any case, the simultaneous observation of bound-to-bound and bound-to-continuum photocurrent in these devices is in agreement with prior reports of similar SiGe QWIPs.^{5,6}

Figure 6 summarizes our findings regarding the peak responsivity R_p of the samples under study. This parameter is

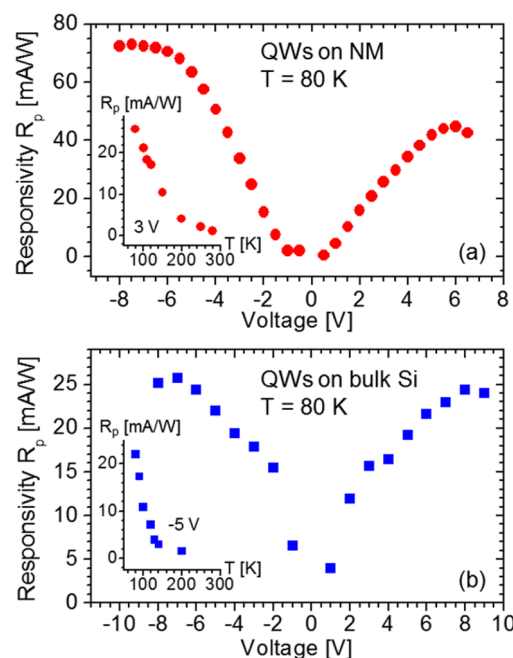


Figure 6. Measured peak responsivity of (a) a NM QWIP and (b) a nearby device grown directly on the Si contact layer, plotted as a function of applied voltage for a heat-sink temperature of 80 K. The inset in each panel shows the peak responsivity of the corresponding device plotted as a function of temperature at fixed bias voltage (3 V in (a) and -5 V in (b), chosen to produce roughly the same 80 K responsivity in the two devices).

evaluated by repeating the same FTIR measurements with a calibrated InSb photodetector, and then comparing the resulting photocurrent signal to the SiGe QWIP data at the photon energy of peak detection (near $5 \mu\text{m}$ wavelength). In Figure 6a we plot the peak responsivity of the device of Figure 5a at 80 K, as a function of bias voltage, showing a maximum value of about 73 mA/W. The observed saturation in R_p with increasing voltage can be attributed to a concomitant saturation in drift velocity, as well as heating of the device active layer by the dark current. As shown in the inset of Figure 6a, the peak responsivity also decreases monotonically with increasing temperature and can be resolved up to about 280 K. Figure 6b shows the same data measured with a nearby QWIP on the same chip, grown directly on the Si contact layer. Compared to the NM-grown sample of Figure 6a, this device has a slightly smaller area ($400 \times 300 \mu\text{m}^2$ vs $400 \times 400 \mu\text{m}^2$), but the responsivity values shown in the figures are scaled accordingly. These plots illustrate the best performance observed with both types of devices. The bulk-grown device of Figure 6b exhibits the same qualitative behavior as the NM QWIP of Figure 6a. Its peak responsivity, however, is noticeably smaller, by up to a factor of over 2, consistent with the higher structural quality of

the strain-compensated QWs grown on the NMs. This result is also in agreement with the ISB absorption spectroscopy study of ref 19, where a similar increase in peak absorption coefficient was measured in NM-grown SiGe/Si QWs relative to a nominally identical sample grown on a rigid Si substrate. Finally, we note that the data plotted in Figure 6a are comparable with prior reports of SiGe QWIPs based on QWs of similar thicknesses and compositions,^{4–6} albeit grown by molecular beam epitaxy, which generally provides superior control of the QW growth parameters.

CONCLUSIONS

In conclusion, the results presented in this work illustrate the feasibility and potential advantages of electrically injected SiGe ISB devices grown on strain-relaxed NM stacks. The underlying fabrication process is based on a unique property of multiple-QW NMs released from their original substrate, that is, the ability to minimize internal stress due to lattice mismatch via elastic strain sharing among the constituent layers, rather than through the formation of misfit dislocations or other structural degradation. As a result, an arbitrary number of identical QW periods can then be grown on such NMs without any global strain accumulation, even though each individual layer is still slightly either tensilely or compressively strained. We have furthermore shown that released and therefore elastically strain-relaxed NM stacks can be transferred and bonded onto a conducting substrate in a way that allows for apparently unimpeded current injection into (and extraction from) the QWs. In the present work, this approach has been applied to the development of mid-infrared SiGe QWIPs featuring promising device characteristics. The same idea can be extended to more complex SiGe QW structures, including the gain medium of quantum cascade lasers, where constraints related to lattice mismatch are especially demanding. More generally, similar structures could also be developed for other materials systems, for any application where strain accumulation caused by lattice mismatch with the epitaxial substrate is a limiting factor.

METHODS

SiGe/Si Multiple-Quantum-Well Film Growth. The film growth starts with a (001) SOI wafer having a ~ 25 nm thick Si template layer and a 150 nm thick BOX layer (Figure 2a). The SOI wafer is diced into pieces 8 mm wide and 44 mm long. The standard semiconductor cleaning procedure is performed: (1) 20 s in 10% HF, (2) 10 min in Piranha clean, and (3) 15 min in standard clean 1 (SC1), with a 5 min DI water rinse between each step. Before transferring into the growth chamber, the wafer is dipped in 10% HF for 20 s, which leaves its surface H-terminated. Next, a three-period SiGe/Si QW structure is grown on the strip using SiH_4 , GeH_4 , and B_2H_6 as the precursor gases with a growth pressure of a few mTorr (Figure 2b). The substrate temperature is approximately 700 and 600 °C for the growth of the Si and SiGe layers at a rate of 4 and 2 nm/min, respectively. It should be noted that in this initial SiGe/Si QW structure the (undoped) Si template layer of the SOI wafer is used as the first barrier.

Lattice-Matched Nanomembrane Substrate Fabrication. The as-grown film is patterned by optical lithography to produce several membranes with areas of a few mm^2 (without any etching holes). Next, HF is used to completely etch away the BOX. As a result, the membranes are released in place and

settle on the underlying Si substrate (Figure 2c). Upon dipping into DI water, the NMs release from the wafer and float on the water surface (Figure 2d). The floating NMs are then transferred and bonded onto a 500 nm thick highly *p*-doped ($\sim 1 \times 10^{19} \text{ cm}^{-3}$) Si contact layer (Figure 2e), which was previously grown on a separate Si substrate and exposed to a UV-ozone treatment to make its surface more hydrophilic by creating a thin native-oxide layer. This layer increases the wettability of the new host substrate and therefore enhances the bonding without the presence of wrinkles. A 1 h anneal at 450 °C is then used to improve the NM bonding to the Si contact layer (Figure 2f). Next, the surface is prepared for the subsequent growth with a mild cleaning procedure (consisting of three cycles of UV-ozone exposure, HF dipping, and DI water rinsing) that prevents any NM peel-off. The last step involves dipping the sample into a HF solution for 20 s to remove the exposed native-oxide layer and leave the sample surface H-terminated. The native oxide beneath the bonded NMs is not removed during this step, as HF cannot etch the buried oxide areas in such a short time frame. Finally, 51 additional QW periods are grown under the same conditions described above.

X-ray Diffraction Measurements. The structural properties of the SiGe/Si multiple-QW samples are characterized with high-resolution X-ray diffraction (Phillips Analytical X'Pert PRO). To determine the Ge composition and the SiGe-well and Si-barrier thicknesses, omega-2theta scans along the (004) reflection are measured and then fit to simulation results using RADS Mercury software. Reciprocal-space maps around the (224) reflection are measured to confirm that the film is coherently grown on the substrate and to evaluate its structural uniformity.

QWIP Fabrication and Photocurrent Measurements. The epitaxial sample is diced into $8 \times 5 \text{ mm}^2$ chips, where the QWIP mesas are fabricated by reactive ion etching in SF_6 with a Cr mask. Electron-beam evaporation is used to define the metal contacts, consisting of a bilayer of Ti and Au with thicknesses of 5 and 110 nm, respectively. The devices are then soldered on a Cu block, wire-bonded to Au-plated ceramic pads, and finally mounted on the coldfinger of a continuous-flow cryostat. The photocurrent spectra are measured with an FTIR spectrometer via step-scan phase modulation and lock-in detection, with the device used as the external detector.

ASSOCIATED CONTENT

Supporting Information

The Supporting Information is available free of charge on the ACS Publications website at DOI: 10.1021/acsphotonics.6b00524.

Theoretical fitting of the HRXRD data; optical and SEM images of the surface of the QWIP heterostructure grown on a NM substrate (PDF).

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Notes

The authors declare no competing financial interest.

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