Tensile-Strained Self-Assembly of InGaAs on InAs(111)A

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ABSTRACT
We have determined a reproducible set of growth conditions for the self-assembly of tensile-strained In$_{1-x}$Ga$_x$As quantum dot (QD) nanostructures on (111)A surfaces. During molecular beam epitaxy, In$_{1-x}$Ga$_x$As islands form spontaneously on InAs(111)A when the Ga content $x \geq 50\%$. We analyze the structure and composition of InGaAs/InAs(111) samples using atomic force microscopy, transmission electron microscopy, and electron energy loss spectroscopy. We demonstrate control over the size and areal density of the islands as a function of In$_{1-x}$Ga$_x$As coverage, In$_{1-x}$Ga$_x$As composition, and substrate temperature. We calculated the conduction and valence band energy values for these QDs in an InAs matrix. This work supports the efforts to establish InAs(111)A as a platform for future incorporation with other (111)-oriented materials from the 6.1 Å family of semiconductors.

I. INTRODUCTION
Interest in III-V semiconductors with the (111) orientation has increased in recent years. The unique structural and electronic properties of semiconductors with this crystallographic orientation are the main driver behind this renewed interest. The symmetry of the (111) surface means these III-V materials are well-suited to integration with 2D materials such as transition metal dichalcogenides, V$_2$-VI$_4$ topological insulators, and certain IV-VI rock salt semiconductors such as PbSe. Transistors with a (111) orientation may offer ballistic electron transport in both $\Gamma$ and $L$ valleys to overcome the bottleneck in the density of states. Quantum dots (QDs) with a (111) orientation have negligible fine-structure splitting, making them ideal entangled photon sources.

For this latter application, Stranski–Krastanov (SK) self-assembly of QDs on (111) surfaces must be driven by tensile strain. The specific combination of tensile strain and surface orientation creates an energy barrier to dislocation nucleation and glide, resulting in the spontaneous formation of defect-free QDs. The presence of residual tensile strain in self-assembled QDs also modifies their properties, lending them some unusual characteristics. Tensile strain breaks the valence band degeneracy, lifting the light-hole band above the heavy-hole band and raising the prospect of light-hole exciton formation in QDs for quantum media conversion applications. Tensile strain also reduces the semiconductor bandgap energy ($E_g$), pushing QD light emission and absorption toward the infrared (IR) band. This reduction in $E_g$ typically exceeds the increase in the QD ground state transition energy that comes from quantum confinement. The result is light emission below the band edge of the bulk material from which the QDs are composed.

The ability to grow QDs with redshifted light emission presents us with an opportunity to create novel optoelectronic devices
for IR applications. By synthesizing tensile-strained QDs from III-V semiconductors such as In$_{1-x}$Ga$_x$As that have a narrow bulk bandgap, we could, in principle, engineer highly tunable mid-IR light sources. Laser devices based on self-assembled QDs (e.g., Ref. 19) are significantly quicker, easier, and cheaper to grow than quantum cascade structures.  

In this paper, we explore the self-assembly of tensile-strained In$_{1-x}$Ga$_x$As QDs by molecular beam epitaxy (MBE) on InAs(111)A. This is a rich material system: changing the Ga content, $x$, adjusts both the In$_{1-x}$Ga$_x$As bandgap and the amount of tensile strain due to lattice mismatch with the InAs matrix. Raising $x$ increases $E_g$ in In$_{1-x}$Ga$_x$As, but also increases the tensile strain, which serves to simultaneously reduce $E_g$. When we combine these composition-related effects with the additional degree of freedom that comes from quantum confinement, it is easy to see just how tunable the band structure of tensile In$_{1-x}$Ga$_x$As QDs could be.

A few research groups have previously explored tensile-strained GaAs and In$_{1-x}$Ga$_x$As QDs and quantum wells (QWs).  

These reports all focused on GaSb(001) as the substrate material. Despite the fact that efficient tensile strain relief can be an issue for achieving defect-free self-assembly on (001) surfaces, these researchers were able to demonstrate QD and QW light emission in the 2–2.3 μm range by keeping the nanostructures small. More recent studies show that moving from (001) to (111) substrates will give us more flexibility in the QD size, since dislocation formation is retarded in (111)-oriented semiconductors under tensile strain.

We, therefore, use InAs substrates with a (111)A orientation to help minimize the plastic relief of the tensile strain. Our choice of InAs instead of GaSb allows us to use arsenic as a common anion for both the QD and the matrix. This means that we can cap the QDs immediately without the need for complex shutter sequences at the III-As/III-Sb interfaces, which helps minimize any QD annealing effects. We investigate how one can control the size and areal density of the self-assembled In$_{1-x}$Ga$_x$As/InAs(111)A QDs as a function of their composition and MBE growth conditions. We use a combination of computational modeling and photoluminescence (PL) to explore the electronic structure of these QDs. This work sets the stage for future studies of the optical properties of tensile-strained In$_{1-x}$Ga$_x$As(111) QDs.

II. METHODS

All samples were grown via solid-source MBE on unintentionally doped, nominally on-axis ($\pm 0.5^\circ$) InAs(111)A substrates. We use high-purity indium metal to mount the InAs(111)A substrates onto molybdenum blocks, which ensures excellent temperature uniformity across each sample. We monitor substrate temperature ($T_{sub}$) using a thermocouple behind the substrate and an infrared pyrometer, calibrated by reflection high-energy electron diffraction (RHEED) against known changes in surface reconstruction. We calculate growth rates in monolayers per second (ML/s) on (111)A from RHEED intensity oscillations performed on the (001) surface. We calculate the composition of the In$_{1-x}$Ga$_x$As layers using RHEED intensity oscillations and ex situ x-ray diffraction.

We begin by heating the InAs(111)A substrates under As$_x$ to $T_{sub} = 495^\circ C$ to remove the native oxide. We use As$_x$ for consistency with previous studies of InAs(111)A homoepitaxy. We then anneal the substrate for 600 s at this temperature, followed by a further annealing step at 500 °C for 180 s. After oxide removal and annealing, RHEED shows a bright (2 x 2) surface reconstruction. To smooth the substrate surface, we grow a 100 nm InAs buffer at $T_{sub} = 500^\circ C$, with a growth rate of 0.12 ML/s and an As$_x$/Ga flux ratio of 48. We have optimized these conditions for InAs(111)A homoepitaxy previously.  

We then adjust $T_{sub}$ and cell temperatures as required for a specific In$_{1-x}$Ga$_x$As/InAs(111)A deposition experiment. In this study, we explore In$_{1-x}$Ga$_x$As coverage from 2 to 4 ML, In$_{1-x}$Ga$_x$As composition from $x = 0.25$ to 1.00, and $T_{sub}$ from 410 to 500 °C. We adjusted the flux ratio for Ga and In elements in each case to maintain a growth rate of 0.1 ML/s, and a V/III ratio of 300. Depending on the purpose of the experiment, the In$_{1-x}$Ga$_x$As QDs are either left exposed for atomic force microscopy (AFM) or buried with an InAs capping layer for PL spectroscopy. InAs capping layers are grown under the same conditions as the underlying InAs buffer. During the transition between layers, there is no pause other than the minimum required to change the substrate temperature and Ga, In, or As flux as required by the different growth series.

We use NANOscope software to analyze multiple AFM images of surface In$_{1-x}$Ga$_x$As QDs from each sample to find the average QD height, diameter, and areal density. We calculate error bars for each parameter by dividing the standard deviation of each parameter population by the square root of the number of QDs measured, which amount to the total visible QDs in each AFM scan. We examine the structure, morphology, and compositional distribution of our samples using cross-sectional bright-field (BF) transmission electron microscopy (TEM), and scanning TEM (STEM) combined with electron energy loss spectroscopy (EELS) to map material composition. We prepare TEM samples using a focused ion beam lift-out method and characterize them in a JEOL 200F ARM operating at 200 kV. All TEM and STEM images and EELS maps were taken with the samples tilted to align the electron beam to a [110] zone-axis.

III. RESULTS AND DISCUSSION

A. In$_{1-x}$Ga$_x$As on InAs(111)A

Figure 1 shows a matrix of AFM images from the full set of In$_{1-x}$Ga$_x$As on InAs(111)A samples we grew. Moving from left to right corresponds to increasing the Ga content ($x$) of the In$_{1-x}$Ga$_x$As layer, while moving from top to bottom corresponds to increasing the InGaAs coverage. Below we discuss some of the trends revealed in this matrix.

1. In$_{0.5}$Ga$_{0.5}$As/InAs(111)A coverage series

The column with the dashed outline in Fig. 1 highlights a series of In$_{0.5}$Ga$_{0.5}$As samples grown at $T_{sub} = 410^\circ C$ where we increase the coverage from 2 to 4 ML. We focus on this series in Fig. 2.

The sample grown with 2 ML In$_{0.5}$Ga$_{0.5}$As exhibits a comparatively smooth surface, for which flat 2D islands are the predominant morphological feature [Fig. 2(a)]. When we raise the coverage...
to 3 ML, we observe a dramatic change in the morphology, such that the surface is now covered in self-assembled 3D InGaAs QDs [Fig. 2(b)]. This 2D-to-3D transition is a hallmark of the SK growth mode.

We do see evidence in Fig. 2(a) of a low density of small islands appearing as bright spots, which could suggest the onset of 3D nucleation and QD growth. A previous study based on scanning tunneling microscopy of InAs on GaAs(001) showed that QD nucleation actually begins at deposition amounts lower than the established critical thickness of 1.6 ML. It, therefore, seems likely that the surface is now covered in self-assembled 3D InGaAs QDs [Fig. 2(b)]. This 2D-to-3D transition is a hallmark of the SK growth mode.

The tensile lattice mismatch between In0.25Ga0.75As and InAs(111)A is 3.3%. A critical thickness of ~2 ML is consistent with the 2–2.5 ML reported for the InAs/GaAs(001) QD system that has a compressive mismatch of 3.2%. By staying below this critical thickness, we can, therefore, produce 2D In0.25Ga0.75As QWs under large tensile strains, with potential IR optoelectronic applications. As we raise the In0.25Ga0.75As coverage from 3 to 4 ML, the 3D QDs increase in average height (0.57 ± 0.03 to 0.69 ± 0.14 nm), average diameter (7.9 ± 0.2 to 8.7 ± 1.0 nm), and areal density (2.3 to 6.8 × 10¹⁰ cm⁻²) [Fig. 2(c)].

We can tune the critical thickness for the SK growth mode by modifying the magnitude of the tensile strain. The presence of QDs in Figs. 1(b) and 1(c) means that the transition to 3D self-assembly has already occurred by the time we deposited 2 ML InGaAs. In0.25Ga0.75As and GaAs are lattice mismatched to InAs by 5.0% and 6.7%, respectively. Compared with the In0.25Ga0.75As sample in Fig. 1(a), the resulting increase in tensile strain in these cases has reduced the critical thickness below 2 ML. Again, this is consistent with other highly strained SK systems such as InAs/GaAs(001) QDs where 7.2% compressive lattice mismatch results in a critical thickness of ~1.6 ML. In general, the higher the strain, the lower the critical thickness for SK growth. The ternary nature of the tensile InGaAs QD system gives us a great deal of control in this respect.

2. InₓGa₁₋ₓAs/InAs(111)A composition series

The row with the black outline in Fig. 1 highlights a 3 ML InₓGa₁₋ₓAs sample series grown at Tsub = 410 °C, in which we tune the Ga concentration: x = 0.25, 0.50, 0.75, and 1.00. We focus on this series in Fig. 3.

At x = 0.25, the tensile strain resulting from the 1.67% lattice mismatch between the In0.75Ga0.25As and the InAs(111)A substrate is insufficient to drive QD self-assembly [Fig. 3(a)]. Although a very low density of 2D islands show up as isolated bright spots, these do not develop into 3D QDs upon deposition of more material. Indeed, Fig. 1(h) shows that a 2D planar surface persists, even after 4 ML In0.75Ga0.25As deposition. As we have already mentioned, the ability to grow smooth tensile-strained InGaAs layers could be useful for QW-based IR optoelectronics.

As we raise x, we reduce the InₓGa₁₋ₓAs lattice constant. The result is a greater lattice mismatch with the InAs(111)A substrate and hence higher tensile strain. For a Ga concentration of x = 0.5,
the tensile strain is large enough to produce spontaneous QD formation from 3 ML coverage [Fig. 3(b)]. As we further increase the Ga concentration to $x = 0.75$ [Fig. 3(c)] and then to $x = 1.0$ [Fig. 3(d)], we see that the 3 ML In$_{1-x}$Ga$_x$As QDs become larger and more densely packed. InGaAs QD areal density increases by almost six times ($2.32 \times 10^{11}$ to $13.6 \times 10^{11}$ cm$^{-2}$) as we raise $x$ from 0.5 to 1.0 [Fig. 3(e)], with smaller increases in QD height and diameter over the same compositional range [Fig. 3(f)].

That we see larger, higher density QDs in Figs. 3(e) and 3(f) for a constant InGaAs coverage of 3 ML confirms our finding from the InGaAs coverage series that critical layer thickness is dependent on the strain. Because tensile strain increases as we raise $x$, the critical thickness required for the 2D-to-3D SK transition is reduced. A thinner wetting layer means that a greater proportion of the 3 ML InGaAs will end up in the 3D QDs, and so their size and areal density increase accordingly.

### 3. In$_{0.5}$Ga$_{0.5}$As/InAs(111)A substrate temperature series

We use a third set of samples to explore how $T_{\text{sub}}$ affects the morphology of 3 ML In$_{0.5}$Ga$_{0.5}$As QDs [Figs. 4(a)–4(d)]. As we raise $T_{\text{sub}}$ from 410 to 500 °C, the areal density of the In$_{0.5}$Ga$_{0.5}$As QDs decreases by almost 20× from 2.32 to $0.12 \times 10^{11}$ cm$^{-2}$ [Fig. 4(e)].

The ability to tune QD density by as much as an order of magnitude with InGaAs composition (i.e., tensile strain) [Fig. 3(e)] and $T_{\text{sub}}$ [Fig. 4(e)] is desirable for future optoelectronic applications. High QD density is suited to high intensity photon emission
for LEDs/lasers, while low QD density is appealing for single-photon generation where light collection from individual QDs is needed.

As we raise $T_{\text{sub}}$ over the same range, the average diameter of the InGaAs QDs increases monotonically by a factor of 3 [Fig. 4(f)]. Average QD height also initially increases, peaking at $\sim 440^\circ$C [Fig. 4(f)]. These trends are consistent with ripening effects observed in previous studies of both compressive- and tensile-strained QD systems. As we increase $T_{\text{sub}}$, adatom surface diffusion length is enhanced, resulting in the formation of large, low density QDs that are more efficient at minimizing the strain energy than small, high density QDs. However, further raising $T_{\text{sub}}$ to 470–500$^\circ$C, reduces the average QD height. These observations indicate the formation of larger, flatter InGaAs islands [Fig. 4(d)] at high $T_{\text{sub}}$. A similar temperature-dependence of QD aspect ratio has been seen during SK growth of tensile-strained GaAs/InAlAs(111)A QDs (Ref. 18) and droplet epitaxy of InAs/GaAs(111)A QDs, where it was attributed to longer adatom diffusion and an enhanced ability for adatoms to migrate from the tops of islands to the terrace below.

4. Transmission electron microscopy of In$_{0.5}$Ga$_{0.5}$As

We used cross-sectional TEM to examine 3 and 4 ML In$_{0.5}$Ga$_{0.5}$As QDs grown at 410$^\circ$C, i.e., equivalent to those in Figs. 2(b) and 2(c) but capped with InAs. We used low magnification TEM to survey wide areas of these samples and examine their film quality (Fig. 5), and strain contrast from the presence of the InGaAs QD and wetting layers is visible at the interface between the InAs buffer and cap layers. For the 3 ML sample, we see that the InAs capping layer is free of threading dislocations originating at the In$_{0.5}$Ga$_{0.5}$As QD layer [Fig. 5(a)]. This lack of defects indicates that tensile strain in the 3 ML InGaAs QDs is not high enough to cause widespread relaxation. We do, however, see several defect types originating in the substrate, in the InAs buffer, and at the interface between the two. These defects and strain contrast visible along that interface are unrelated to the InGaAs QDs and instead suggest that the quality of the InAs(111)A substrate itself and the pregrowth surface treatment was not fully optimized for these specific samples.

When we raise the In$_{0.5}$Ga$_{0.5}$As coverage to 4 ML [Fig. 5(b)], we see more pronounced contrast in the InGaAs layer than for the 3 ML InGaAs sample, consistent with the larger InGaAs QDs in
this sample (Fig. 2). However, we also observe a high density of defects originating at the InGaAs layer, consistent with increased strain from the larger InGaAs QDs in this sample. This was a high enough stress to cause relaxation via widespread dislocation nucleation and glide.12 For self-assembly of defect-free In$_{0.5}$Ga$_{0.5}$As QDs, one must, therefore, remain below this upper deposition limit of 4 ML.

Figures 6(a) and 6(d) show BF STEM images from the 3 and 4 ML InGaAs QD samples, respectively. Figures 6(c) and 6(f) show corresponding EELS maps of the Ga L-edge signal from the boxed regions of the 3 and 4 ML samples in Fig. 6(a) and 6(d). Figures 6(b) and 6(e) show high-resolution TEM images of QDs from the 3 and 4 ML samples respectively, with numbered QD locations that correspond to positions on the EELS maps.

Figures 6(d) and 6(e) show that the strain contrast around each 4 ML QD extends much farther than the EELS map in Fig. 6(f) shows the actual QD size to be. Conversely, for the 3 ML QD sample, the apparent QD size in the BF TEM/STEM [Figs. 6(a) and 6(b)] better matches the QD size indicated by EELS mapping [Fig. 6(c)] because the QDs are not straining the surrounding InAs lattice as much.

We note that while both the 3 and 4 ML QDs are visible in BF TEM and BF STEM imaging modes due to their strain contrast, they are not readily visible in the annular dark-field STEM imaging mode used concurrently with EELS mapping. Therefore, to allow us to correlate the EELS maps with BF TEM and BF STEM images of the same areas, we intentionally chose EELS mapping locations close to easily identifiable features or defects such as that on the
left-hand side of Fig. 6(a). For the 3 ML InGaAs sample, in particular, Fig. 5(a) shows that such defects are not representative of the crystal quality of the sample as a whole.

The EELS mapping reveals that in both samples, InGaAs QDs have low height-to-diameter ratios and that QD size and density increase when moving from 3 to 4 ML coverage. These observations are consistent with our AFM measurements of corresponding uncapped 3ML and 4ML samples in Figs. 2(b) and 2(c).

We were interested whether capping with the InAs top barrier changes the shape of the InGaAs QDs. Previous reports describe the flattening of QD nanostructures during capping due to surface segregation and interdiffusion of atoms between QDs and the surrounding matrix. For example, during the capping of tensile InGaAs QDs with GaSb(001), the layer of 3D QDs becomes completely “smeared out” into a 2D QW.\(^6\) This transformation of discrete QDs into a highly uniform QW with abrupt interfaces was attributed to the Sb acting as a surfactant during capping with GaSb. For the InGaAs/InAs material system described here, the absence of Sb would suggest this process is unlikely to occur. Indeed, as we have already noted, the TEM micrographs in Figs. 5 and 6 show that well-defined 3D InGaAs nanostructures remain even after capping with InAs.

For the case of compressively strained InAs/GaAs(001) QDs, a less extreme flattening process occurs. The exchange of In and Ga atoms during capping results in the QDs becoming truncated, with a flat top and steep sides.\(^6\) Interestingly, our tensile InGaAs/InAs(111)A QD system presents the reverse arrangement, truncated, with a flat top and steep sides.\(^40\) Interestingly, our tensile InGaAs/InAs(111)A system reveals the reverse arrangement, with the In atoms more mobile than In could actually serve to stabilize the discrete InGaAs QDs into a highly uniform QW with abrupt interfaces. To explore this idea, we compared mobile Ga atoms relative to the Sb acting as a surfactant during capping with InAs. For the InGaAs/InAs(111)A QD system, the absence of Sb would suggest this process is unlikely to occur. Indeed, as we have already noted, the TEM micrographs in Figs. 5 and 6 show that well-defined 3D InGaAs nanostructures remain even after capping with InAs.

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5. Electronic structure of In\(_{1-x}\)Ga\(_x\)As/InAs(111)A QDs

Having established the MBE conditions required for the defect-free growth of these novel, tensile-strained InGaAs/InAs(111)A QDs, we moved on to explore their electronic structure. Although the bulk bandgap of In\(_{1-x}\)Ga\(_x\)As is larger than InAs for all compositions, we know that tensile strain can cause profound changes in the semiconductor band structure.\(^9,10,13,16\) We were, therefore, interested to see whether for specific compositions, the tensile strain could reduce the InGaAs/InAs(111)A QDs bandgap sufficiently to produce carrier confinement.

We, therefore, calculated the bulk band alignments of In\(_{1-x}\)Ga\(_x\)As/InAs(111)A as a function of composition (Fig. 7). Our model accounts for the effects of tensile strain on the InGaAs band structure by using two different components, an in-plane hydrostatic term and an axial term. The hydrostatic strain acts on the bandgaps of the system, while the axial component modifies the valence bands. We then calculate the energy of any confined states in the QDs by solving the Schrödinger equation for the envelope function in the effective mass approximation. For a more detailed description of our model, see Ref. 10.

For In\(_{1-x}\)Ga\(_x\)As compositions \(x = 0.25–0.75\), we obtain a type-II band structure, with a barrier in the valence band [Figs. 7(a)–7(c)].

![FIG. 7. Simulation of the bulk bandgap of In\(_{1-x}\)Ga\(_x\)As (black curves) under biaxial tensile strain on InAs(111)A substrates. The systems analyzed in each panel correspond to (a) In\(_{0.75}\)Ga\(_{0.25}\)As, (b) In\(_{0.5}\)Ga\(_{0.5}\)As, (c) In\(_{0.25}\)Ga\(_{0.75}\)As, and (d) GaAs. The darker region (orange in color) signifies the experimentally derived thickness of the 2D wetting layer, showing its reduction with increasing strain. For samples with \(x > 0.5\), the lighter region (yellow in color) represents the average QD height measured with AFM. The lines within the InGaAs conduction band (red in color) show the electron ground states for the wetting layer (solid line) and QDs (dashed line), calculated by solving the Schrödinger equation in the effective mass approximation.](https://vigilant-js.062809-7.avs.scitation.org/journal/jvb)
QDs with a type-II band alignment are of interest for various optoelectronic applications due to their enhanced carrier lifetimes and the fact that their emission wavelength is tunable with excitation density.\textsuperscript{41,42} In contrast, for the case where $x = 1.0$, the bandgap of the GaAs is too large for the tensile strain to lower the conduction band edge below that of the InAs(111)A barriers. According to these calculations, the tensile-strained GaAs forms “antidots” with respect to the InAs(111)A, with barriers in both the conduction and valence bands and no confined states [Fig. 7(d)].

For $x = 0.25$, our calculations suggest a weakly confined electron ground state in the tensile-strained QW [Fig. 7(a)]. In the coupled wetting layer-QD systems formed when $x = 0.5$–0.75, the electron ground state appears to be close to or just above the continuum of states of the InAs barriers [Figs. 7(b) and 7(c)].

We used PL spectroscopy to test these predictions of weakly confined electron states in at least some of our samples. In particular, we looked for evidence of recombination between electrons confined within the In$_{1-x}$Ga$_x$As QWs or QDs and holes localized outside of the QDs in the InAs(111)A barriers. However, despite exploring a range of measurement temperatures and excitation densities, we were unable to detect any obvious light emission from the QDs. This lack of a PL signal confirms either the limitations of our model’s ability to capture the effect of tensile strain on band structure or the fact that even at 7 K, the thermal broadening of the confined states was sufficient for the electrons to escape. To overcome this issue in future, we will use this work as the basis for synthesizing these tensile-strained In$_{1-x}$Ga$_x$As(111)A QDs on barriers such as AlGaSb for which even a modest increase in bandgap will ensure robust electron confinement.

IV. CONCLUSIONS

We have established that tensile-strained In$_{1-x}$Ga$_x$As forms either QWs (for $x = 0.25$) or self-assembled QDs (for $x \geq 0.5$) when deposited on InAs(111)A in the range of $T_{\text{sub}} = 410$–500 °C. For a composition of In$_{0.5}$Ga$_{0.5}$As, the critical thickness for the SK growth mode formation of these QDs lies between 2 and 3 ML. The accumulated tensile strain energy from 4 ML In$_{0.5}$Ga$_{0.5}$As coverage exceeds the barrier to dislocation nucleation and glide and so plastic strain relief occurs. We can control both the density and size of the InGaAs QDs as a function of the MBE growth conditions. Our band structure calculations predict a type-II band alignment with weak electron confinement for In$_{1-x}$Ga$_x$As(111)A QWs and QDs where $x \leq 0.75$. We anticipate that the future use of wider bandgap barrier materials will allow us to demonstrate highly tunable light emission from these tensile-strained In$_{1-x}$Ga$_x$As(111)A QDs for various IR applications.

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DATA AVAILABILITY

The data that support the findings of this study are available from the corresponding author upon reasonable request.

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