Defect Filtering for Thermal Expansion Expanded Dislocations in III-V Lasers on Silicon

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Growing III-V semiconductor lasers directly on silicon circuitry will transform information networks. Currently, dislocations limit performance and lifetime even in defect tolerant InAs quantum-dot (QD)-based devices. Although the QD layers are below the critical thickness for strain relaxation, they still contain long, previously unexplained misfit dislocations which lead to significant non-radiative recombination. This work offers a mechanism for their formation, demonstrating that the combined effects of thermal-expansion mismatch between the III-V layers and silicon and precipitate and alloy hardening effects in the active region generate the misfit dislocations during sample cooldown following growth. These same hardening effects can be leveraged to mitigate the very problem they create. The addition of thin, strained, indium-alloyed trapping layers displaces 95% of the misfit dislocations from the QD layer, in model structures. In full lasers, performance benefits from adding trapping layer now both above and below the QD layers include a twofold reduction in lasing threshold currents and a threefold increase in output powers. These improved structures may finally lead to fully integrated, commercially viable silicon-based photonic integrated circuits.

**MAIN**

Silicon-based photonic integrated circuits will dramatically increase data network bandwidth and energy efficiency and enable new paradigms in chip-scale sensing, detection, and ranging. Direct crystal-growth methods integrating III-V semiconductor lasers with silicon promise cost-effectiveness and scalability,\textsuperscript{1,2} however, fabricating reliable, high-performance GaAs- or InP-based lasers on silicon has proven challenging.\textsuperscript{3–4} Lattice constant mismatch between the silicon substrate and III-V film generates dislocation line defects, including ‘threading’ dislocations, which rise upward through the film.\textsuperscript{4} Where they intersect the device’s active region, they facilitate non-radiative recombination, degrading performance. The energy released causes dislocations to lengthen during device operation, a run-away degradation process ending in device failure.\textsuperscript{2,5–7} Despite decades of work to reduce threading dislocation densities to $10^6$–$10^7$ \text{cm}^{-2} (refs [8–10]) \textsuperscript{8–10} and to develop dislocation-tolerant active materials such as InAs-quantum dots (QDs) in quantum wells (QW) (dots in a well or DWELL),\textsuperscript{11–17} threading dislocations continue to stifle the development of commercially viable III-V lasers on Si.\textsuperscript{7,18,19} We have recently identified the root of this contradiction using plan-view scanning transmission electron microscopy (PV-STEM): unexpected dislocations found lying flat along the uppermost and lowermost QD layers, in even record lifetime QD lasers.\textsuperscript{20,21}

These dislocations, termed ‘misfit’ dislocations, have a far larger interaction area with the active region than threading dislocations as they lie horizontally in the film and additionally are potent non-radiative recombination centers.\textsuperscript{22} As they serve to mediate lattice mismatch, misfit dislocations are traditionally understood to form in layers exceeding a certain “critical thickness” during growth.\textsuperscript{23,24} While active-region-adjacent misfit dislocations in QW lasers on silicon have also been reported, they have gone unexplained.\textsuperscript{25} This is because, in both QW and DWELL lasers, active region layers are designed to be below critical thickness and therefore misfit dislocation free.\textsuperscript{26} Furthermore, in DWELL structures, the QDs provide a precipitate hardening effect, which increases the critical thickness.\textsuperscript{27} These misfit dislocations are also easily overlooked: traditional cross-sectional transmission electron microscopy (XTEM) renders them nearly invisible (see Figure S1, Supporting Information), their strain contrast masked by the QDs. Thus, as misfit dislocations are both unexpected and obscured from view, they have gone unaddressed in QD systems and a formation mechanism has yet to be offered.

Here, we remedy not only this lack of clarity, but also demonstrate a viable solution to the underlying problem. We identify two key components that enable the formation of misfit dislocations: lattice hardening in the active region and tensile stress in the film from thermal expansion mismatch. We further identify that the mechanical hardening we observe is in part a result of semiconductor alloy hardening\textsuperscript{28,29} in the DWELL.
With this knowledge, we engineer thin, strained indium-alloyed “trapping layers” inserted a short distance both above and below the laser active region to extend the mechanically hardened region and displace the defects from the QDs. Our atypical defect filtering technique successfully removes 95% of misfit dislocations from the QDs in model structures. In full lasers, this filtering technique substantially reduces active-region defect densities and yields dramatic improvements in laser performance.

We first examine a model structure without a trapping layer by scanning electron microscopy (SEM)-based cathodoluminescence spectroscopy (CL) to directly observe the effects of misfit dislocations and explain their formation. Throughout the paper, we refer to structures with trapping layers as “trapping layer” structures and those without as “baseline” structures. Here, the baseline consists of a single molecular-beam-epitaxy-grown QD layer capped with 100 nm of GaAs and grown on a GaAs-on-Si template (see Supporting Information Figure S2a for full structure and S2b for buffer structure). Using SEM CL, we observe a network of dark lines and spots, corresponding to misfit dislocations and threading dislocations, respectively. In Figure 1a, the sharp, dark lines in the luminescence map of the wetting layer (a thin conformal InAs layer beneath the QDs) indicate that the misfit dislocation segments lie adjacent to the QDs, leading to lower light emission in the vicinity. The InAs QD ground-state luminescence map (Figure 1b) has these same dark features, however they appear more diffuse from inhomogeneous strain and uneven carrier confinement.

Our proposed mechanism for the formation of these misfit dislocations follows. In addition to being 4% lattice-mismatched to silicon, GaAs has a larger coefficient of thermal expansion (\(\alpha_{\text{GaAs}} - \alpha_{\text{Si}} \approx 3 \times 10^{-6} \, \text{K}^{-1}\)). During post-growth cooldown, the film contracts more quickly than silicon substrate, which generates up to 0.1% biaxial tensile strain in the GaAs layers as they approach 300 °C. The existing threading dislocations—extensions of III-V/Si interfacial misfit dislocations that mediate the lattice mismatch during growth—experience a net glide force in GaAs layers thicker than a few hundred nanometers. The indium containing QD layer, however, locally pins the threading dislocation segment where it cuts through (blue box in Figure 1c).\[130] The pinning shear, a result of resistive forces imparted by both precipitate-like InAs QDs and alloy fluctuations in the In\(_{0.15}\)Ga\(_{0.85}\)As QW, inhibits threading-dislocation glide in the QD layer. Thus, only the free threading segment in the GaAs buffer glides and, in doing so, lays down a misfit dislocation segment at the QD layer interface (Figure 1c). Additionally, since glide kinetics at these intermediate temperatures limit dislocation motion, the total misfit dislocation length at the QD layer should be proportional to the threading dislocation density.

Lacking practical methods to uniformly halt or enable dislocation glide throughout the structure we insert a 7-nm In\(_{0.15}\)Ga\(_{0.85}\)As misfit trapping layer 100 nm below the QDs to extend the lattice hardened region and prevent misfit dislocation formation at the QD layer. This second model structure, nearly identical to the baseline structure but with the trapping layer added (see Supporting Information Figure S2c), shows a 95% reduction in total dark line length (Figure 1d and 1e). The few remaining short dark-line segments indicate that some misfit segments have not been trapped. We attribute the faint broad dark lines to misfit dislocations displaced to the trapping layer where they only slightly reduce the light emission from the QDs. As the trapping layer introduces an additional dislocation pinning point (blue box) (Figure 1f) and the thin intermediate GaAs is below critical thickness, dislocation glide and misfit formation occur almost exclusively below the trapping layer. Pinning by the trapping layer is apparently not perfect meaning some threading segments can glide through the trapping layer while remaining pinned in the QD layers. For additional information on the difference in these effects see Figure S3 (Supporting Information).

While CL reveals the optoelectronic impact of both misfit and threading dislocations, we use another SEM-based characterization technique, electron-channeling contrast
imaging (ECCI), to gain additional insight into their structural characteristics. Using ECCI, we can observe in-situ a continuation of the misfit-dislocation formation process that occurs during cooldown because at room temperature the films retain a 0.15% tensile strain; sluggish glide kinetics block relaxation below ~300 °C in GaAs based structures. However, under electron-beam irradiation in an SEM, electron-hole pairs are generated and can recombine non-radiatively at dislocations. This process releases enough energy to revive dislocation glide, a phenomenon known as recombination-enhanced dislocation motion (REDM). Figure 2a shows an ECCI time-lapse evolution of a single threading dislocation in the baseline, which, by chance, did not glide to form a misfit dislocation during cooldown. Assisted by REDM, the dislocation, which is pinned at the QD layer, glides below it to form a misfit segment that lengthens over time. See Video S1 (Supporting Information) for a large-area time-lapse video and Figure S4 (Supporting Information) for initial and final still frames. That this process continues at room temperature provides direct evidence that thermal strain buildup during cooldown and local pinning drive misfit dislocation formation. The observed dislocation structure agrees with the illustration in Figure 1c: the threading dislocation’s point contrast is only visible where it exits the film surface; on the other end, its contrast gradually fades as the dislocation sinks below the ECCI detection depth.

Figure 2b-2e use ECCI to compare misfit dislocation densities between the baseline and the trapping layer structure before and after heavy electron-beam irradiation. In the as-grown baseline structure (Figure 2b), misfit dislocations, marked with black arrows, are present following growth and cooldown. Their sharp contrast indicates that they are near the film surface, likely just below the QD layer, consistent with Figure 1a and 1b. Electron-beam irradiation enables continued film relaxation (Figure 2c), so new sharp-contrast misfit dislocation segments, marked with orange arrows, form and grow. Compared to baseline, the as-grown trapping-layer structure (Figure 2d) has a much lower density of high-sharpness misfit dislocations. As sharpness indicates depth in ECCI, we measure a 95% reduction from baseline to total misfit dislocation length near the QDs (over a 2500-μm² area), in agreement with CL. Electron-beam irradiation results in a high-density network of diffuse-contrast lines (Figure 2e) because, as in the baseline, REDM enables continued strain relaxation. However, their diffuse contrast indicates that these dislocation segments lie farther down in the structure, likely at the trapping layer. Notably, the density of high-sharpness dislocations remains constant, indicating that SEM irradiation does not increase misfit dislocation length at the QD layer. As REDM is a common failure mechanism in semiconductor lasers, this is a good sign for laser reliability.

To investigate the efficacy of misfit trapping layers in full laser devices, we fabricated InAs DWELL ridge structures on (001) Si with trapping layers in the epitaxial stack, shown schematically in Figure 3a, alongside a baseline sample with no trapping layers. Unlike the model structure, the material above the active region is sufficiently thick to relax during cooldown, so we insert a single 7-nm layer of In₀.₁₅Ga₀.₈₅As above and In₀.₁₅Al₀.₈₅As below the active region to prevent misfit-dislocation formation on both sides (trapping layers marked with red boxes). These selected alloys minimize electron and hole barriers formed by unfavorable band alignments in the laser. We first grew a laser structure with trapping layers separated from the nearest QD layer by 80 nm (TL80). See Figure S5a and S5b (Supporting Information) for detailed baseline and TL80 epitaxial stacks, respectively. Figure 3b and 3c show the effect of trapping layers on misfit-dislocation formation via bright-field (BF) cross-sectional STEM. The sample lift-out geometry relative to the misfit array
ensures that all misfits appear as equal-length horizontal lines. Misfit dislocations, marked with black arrows, lie at both the upper and lower trapping layers, successfully displaced away from the active region. Figure 3c shows this more clearly at higher magnification, and Figure S6 (Supporting Information) provides additional cross-sectional evidence of misfit dislocation trapping.

We additionally illustrate the differences between the TL80 and baseline structures at the single-dislocation level using cross-sectional strain-contrast electron tomography generated from multiple BF plan-view (PV)-STEM images taken across a range of tilt angles. Although tomography is traditionally performed by tilting along a single axis, we used a double-tilt holder (see Experimental Section) to follow the \( g = 220 \) Kikuchi band; this maximizes dislocation contrast in our thick samples. A sample PV-STEM image for the baseline laser (Figure 3d) shows a misfit dislocation amid a field of QDs. The tomographic reconstruction (Figure 3e) resolves the five QD layers and shows that this misfit dislocation lies at the uppermost QD layer. Figure S7 (Supporting Information) presents additional evidence of misfit dislocations at the uppermost and lowermost QD layers. For a TL80 laser, Figure 3f and 3g show a PV-STEM image and a tomographic reconstruction, respectively, of a dislocation with a misfit segment and a terminating threading segment. Although strain-contrast tomography cannot resolve the trapping layer itself, the misfit segment clearly lies above the QD layers at the trapping layer’s height. The threading segment travels downward.

Figure 3. (a) Schematic of a quantum dot (QD) laser with trapping layers (red boxes) above and below the QD layers. Baseline samples are equivalent but lack trapping layers. (b) Cross-sectional bright-field (BF) STEM image ([100] zone axis, convergence angle = 10.5 mrad) of a TL80 laser. Inset shows orientations of foil relative to misfit dislocations. Arrows mark misfit segments at the trapping layers. (c) High-magnification image of (b). (d–e) Baseline: (d) BF plan-view (PV)-STEM image (\( g = 220 \), convergence angle = 10.5 mrad) showing a misfit dislocation among QDs. (e) Cross-sectional tomographic strain-contrast reconstruction showing the misfit dislocation at the fifth QD layer. (f–i) TL80 laser: (f) BF PV-STEM image showing a misfit segment terminating in a threading segment. (g) Reconstruction shows the misfit segment lying at the trapping layer. (h) Misfit segments at two heights with a threading dislocation end. (i) Tomographic reconstruction reveals a short misfit dislocation segment at the top QD layer with the rest lying at the trapping layer.
through the five QD layers without forming additional misfit segments. Figure 3h also shows a PV-STEM image of a dislocation in a TL80 laser, but here, there is a short downsloping section along the misfit segment, which indicates a change in height. The tomographic reconstruction (Figure 3i) shows that these two segments lie at the trapping layer and uppermost QD layer. This configuration forms because threading segments, normally pinned by the trapping layer, can become unpinned during cooldown and glide briefly before becoming pinned again. Since dislocation glide cannot occur within QD layers, a misfit segment forms at the outermost QD layer. This is consistent with the incomplete misfit reduction observed by CL and ECCI (Figure 1c, 1d, 2d, and 2e). Nevertheless, just as in the model structures, the trapping layers are very successful here: most misfit dislocation length lies at the trapping layer (see Supporting Information Figure S8).

Nevertheless, this incomplete-trapping phenomenon emphasizes the importance of trapping layer optimization. Here, we begin to explore the design space by investigating the effect of a single parameter—trapping layer height. Placing trapping layers farther from the QD layers likely reduces non-radiative recombination losses from trapping-layer misfit dislocations, but also may exacerbate incomplete pinning and return some misfit segments to the QD layers. To better understand the optimum spacing, we introduce an additional laser design, identical to TL80 but with 180 nm trapping layer spacing (TL180) and compare the electro-optic properties of all three laser types. For the full epitaxial stack of TL180, see Figure S5c (Supporting Information). All three sample types were fabricated together and grown on pieces of the same buffer with a dislocation density of 3x10^7 cm^-2.

Photoluminescence spectroscopy and light output-current-voltage (LIV) curves of a representative high performing device from each design are shown in Figure 4a and 4b, respectively. Introducing trapping layers increases photoluminescence intensity by approximately 2x in TL80 and 1.5x in TL180.

![Figure 4](image)

**Figure 4.** Comparison of TL80 (red), TL180 (blue), and baseline (black) lasers. (a) Photoluminescence intensity comparison of TL80, TL180, and baseline lasers. (b) Single-facet output power (mW) (solid) and voltage (V) (dashed) as a function of current (mA). A reduction in threshold currents, with simultaneous increases in both slope efficiency and peak output powers are observed in both trapping layer devices as compared to baseline. Current-voltage (IV) curves are comparable across trapping layer and baseline device designs. (c-e) Histograms comparing TL80, TL180, and baseline along key performance metrics of (c) threshold current (mA) and threshold current density (A/cm^2), (d) single-facet slope efficiency (W/A), and (e) single-facet output power (mW). Both trapping layer designs show clear improvement over baseline across all metrics.
compared to baseline (Figure 4a). These results agree qualitatively with the marked improvements in single-facet output power (Figure 4b) in fully fabricated TL80 and TL180 lasers over baseline. Histograms comparing the structures further support these performance improvements showing lower currents to begin lasing (threshold current: Figure 4c), more rapidly increasing output powers with input current (slope efficiency: Figure 4d), and higher peak single-facet output powers (Figure 4e).

Both trapping layer designs show a 2x reduction in median threshold current below baseline. The lowest threshold current, 15 mA on a TL180 laser, represents a 40% decrease from baseline minimum. This is also 25% lower than state-of-the-art lasers on Si, with identical device design but 4x lower threading dislocation density.\(^{[34]}\) We additionally observe an impressive 60% (40%) increase in median slope efficiency and a 3.4x (2.6x) increase in median peak single-facet output powers for TL80 (TL180) lasers. Finally, the median electrically dissipated power, in W, at rollover of TL80 (0.85) and TL180 (0.76) is nearly twice that of the baseline (0.46) (see Supporting Information Figure S9 for histograms). Assuming comparable thermal impedances, the inclusion of trapping layers appears to have increased another critical parameter in these lasers: optical amplification (gain).

The TL80 laser design outperforms TL180 across all measured performance metrics aside from minimum and median threshold current. Although we cannot yet conclusively determine which design is superior, we anticipate that differences may emerge during long-term reliability studies due to these competing factors. Additionally, we cannot yet ascertain whether trapping layers adversely impact electrical transport in these lasers due to large variability in the series resistances across all devices. Particularly, higher-than-usual specific contact resistances across all devices (p: \(2.3 \times 10^{-4} \ \Omega \text{cm}^2\), n: \(5.5 \times 10^{-5} \ \Omega \text{cm}^2\)) represent a limiting factor on output power, so processing modifications will likely further improve device performance.

The relative performance improvements we observe here, comparing trapping layer lasers to baseline, are comparable to our own previous performance gains achieved with a full order-of-magnitude reduction in threading dislocation density (7x\(10^5\) cm\(^{-2}\) to 7x\(10^6\) cm\(^{-2}\)).\(^{[35]}\) From a practical perspective, minimizing device thickness is crucial, so it highly advantageous that performance improvements from adding ultrathin filter layers compare favorably to those arising from many hundreds of nanometers of traditional dislocation filters. Furthermore, Jung et al. additionally found that an inverse cubic relationship exists between thread density and device lifetime,\(^{[35]}\) reducing threading dislocation density by one order of magnitude (7x\(10^5\) cm\(^{-2}\) to 7x\(10^6\) cm\(^{-2}\)) resulted in a nearly four order-of-magnitude increase in device lifetimes. Now, knowing that the misfit dislocation density is directly tied to the threading dislocation density, we anticipate similar lifetime increases may be available with a 95% reduction in misfit dislocation length in laser active regions. As previous state-of-the-art QD lasers on silicon fall just short of commercial lifetime requirements at 60 °C operating temperature, eliminating misfit dislocations may enable epitaxially grown lasers to meet the requirements of many telecom and high-performance computing applications.

Researchers in the decades-long study of III-V on silicon integration have duly focused on reducing the density of threading dislocations. Our research demonstrates these threading dislocations additionally give rise to highly damaging but independently addressable misfit dislocations. We resolve this by introducing misfit-dislocation trapping layers both above and below the lasing active region. This represents a major departure from traditional defect filtering and device design while remaining synergistic with ongoing efforts in threading-dislocation reduction. Recognizing that these misfit dislocations form not due to QD and QW layers exceeding critical thickness, as previously thought, but rather due to an unusual mix of thermal and lattice hardening effects occurring during sample cooldown, we provide clear pathways for continued laser performance improvements. The trapping layers presented here and the underlying advances in structural materials science both present opportunities to improve a wide variety of heterogeneously integrated semiconductor devices, such as photodetectors, solar cells, and LEDs. For silicon photonics, this may finally clear the path to commercially viable, monolithically integrated, III-V-on-silicon photonic integrated circuits.

**EXPERIMENTAL SECTION**

**Materials Growth and Laser Fabrication.** All samples were grown using molecular beam epitaxy. GaAs/Si Template: The III-V/Si buffer was grown on a commercially available GaP/Si template available from NAsP III/V, GmbH. A 100-nm GaAs nucleation layer was grown at 500 °C with 0.1 μm/h growth rate. Next, a 1500-nm GaAs buffer was grown at 580 °C and 1 μm/h. Twelve cycles of thermal annealing were then performed (max. temp. = 700 °C, min. temp. = 400 °C). After annealing, a 200-nm In\(_{0.15}\)Ga\(_{0.85}\)As layer was grown. DWELLs and Trapping Layers: DWELLs were grown at 490 °C and annealed at 580 °C for 5 min before growing the spacer layer. Each DWELL had a V/III ratio of 15 in the 2-nm In\(_{0.15}\)Ga\(_{0.85}\)As below the dots, a V/III of 35 for the dots themselves (2.55 ML of InAs deposited at .11 ML/s), and a V/III of 35 in the In\(_{0.15}\)Ga\(_{0.85}\)As cap. The spacers were then grown at 530 °C. Trapping layers were grown at 530 °C with a V/III of ~35 and a growth rate of 2.12 Å/s. Lasers: Layers prior to the QD layers were grown at 580 °C, and those following were grown at 540 °C. The active region p-type modulation doping level is 5x\(10^{17}\) cm\(^{-3}\). 3 μm wide, deeply etched ridge structures with two top-side contacts were fabricated and cleaved to a length of 1500 μm.

**STEM Imaging and Sample Preparation:** Focused Ion Beam Sample Preparation: All BF STEM samples were prepared using standard lift out procedures on a FEI Helios Dualbeam Nanolab 600. PV-STEM and X-STEM samples were 0.7–1.0 μm and 250–650 nm, respectively. BF STEM Imaging: All STEM images were acquired on a ThermoFisher Talos G2 200X TEM/STEM, using an excitation voltage of 200 kV, standard BF STEM circular detector, double tilt holder, and beam convergence angle of 10.5 mrad. X-STEM samples in Figure S1 (Supporting Information), 3b, and 3c were made with [100] foil
orientation and acquired on zone to ensure the capture of misfit dislocation segments from the orthogonal (110) dislocation networks. Misfit dislocations lie 45° to the foil thickness and their observed length is proportional to it. The X-STEM images in Figure S6 (Supporting Information) show [110] foil orientation and were acquired in a g = 002 diffraction condition. We performed tomographic reconstructions of PV-STEM foils containing the laser active regions using a series of two-beam diffraction-contrast images obtained by tilting along the (220) Kikuchi band (Figure 3d-i and Figure S8). The detector served as a virtual aperture. Samples were tilted (α) from -35° to 35° (baseline) or from -28° to 28° (increased foil thickness of TL80 made ±35° tilt impractical). Remaining on Bragg condition necessitated small changes (approximately 4°) in the azimuth angle (β). We used Tomviz (https://tomviz.org) to obtain the cross-sectional reconstructions, manually aligning and optimizing the tilt axis for the sets of 9–11 images and using the ‘Simple back projection’ algorithm. The image presented in Figure S7 (Supporting Information) was taken using (g = 220).

**Photoluminescence Spectroscopy:** Photoluminescence measurements were carried out at room temperature using a non-resonant 780-nm pump laser with normal incidence pumping and detection.

**Cathodoluminescence Spectroscopy:** All CL maps (Figure 1a-b and 1d-e) were collected at room temperature using an Attolight Quanta 400 electron microscope with an accelerating voltage of 10 kV, ≈10-nA probe current, and 0.1 s per pixel exposure. The QD wetting layer and ground state emission maps were collected using an Andor Newton CCD DU920P-Bx-DD silicon detector and an Andor iDus InGaAs detector, respectively.

**Electron-Channeling Contrast Imaging:** ECCI was performed using a ThermoFisher Apreo S SEM (Figure 2a) and an FEI Quanta 400F SEM (Figure 2b-c and 2d-e), aligning to the intersection of the (400) and (220) channeling conditions at 30-kV accelerating voltage and 3-nA beam current. Imaging for time-lapse sequence scanned a 200 μm² area for 66 min. Electron-beam pumping (between Figure 2b and 2c and Figure 2d and 2e) was performed at 30-kV acceleration voltage and 100-nA beam current, scanning a 1725 μm² area for 14 min.

**Laser Characterization:** LIV measurements were conducted at 20 °C.

**SUPPORTING INFORMATION**

See Supporting Information for a document containing: (1) XSTEM ([100] zone axis) showing the appearance of misfit dislocations in QD layers, (2) full structures of (a) the baseline model structure, (b) the buffer, and (c) the trapping layer model structure, (3) schematic representations of (a) the stresses a dislocation experiences in our misfit trapping layer structures and the approximate stress landscapes in the (b) alloy-hardened In0.15Ga0.85As misfit trapping layer and (c) the QD layer, (4) full-view first and final frames from time-lapse ECCI. (Figure 2a shows a small portion of this) and Video, (5) full epilaxial laser stacks for (a) baseline, (b) TL80, and (c) TL180 structures, (6) two tilted cross-sectional bright field STEM images of threading dislocations that have formed trapped misfit segments in a TL80 laser (g = 002), (7) misfit dislocations above and below the QD layers using an offcut STEM foil preparation method, (8) large-area plan-view STEM of the trapping layer structure (used for 3D-tomography), and (9) histograms for electrically dissipated power at rollover for TL80, TL180, and baseline lasers.

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CONFLICT OF INTEREST
The authors have filed a provisional patent application on this work.

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

Defect filtering for thermal expansion induced dislocations in III-V lasers on silicon

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Cross-sectional STEM of baseline structure

Figure S1. (a) Cross-sectional scanning transmission electron microscopy image of misfit dislocations in the active region of a quantum dot laser imaged along the [100] zone axis. The lift-out geometry (inset) ensures all misfit dislocations appear the same length (see also description in Experimental Section), so we can identify them by their length. Black arrows mark misfit dislocations adjacent to the uppermost QD layer; the white arrow marks a misfit dislocation elsewhere in the structure. The contrast of this image was carefully adjusted to maximize misfit dislocation visibility. (b) At higher magnification (boxed area in (a)), the strain contrast from a misfit dislocation is rendered nearly invisible by the strain contrast from the dots themselves.

By intentionally lifting out our foil along the [100] direction, all misfit dislocations will appear the same length, which helps in identifying them within the QD layer. In more conventional [110] lift-out orientations, misfits will appear as black dots (see Ref.[1][1]) that are nearly indistinguishable from the dots themselves or as long segments of slightly increased contrast. Depending on the length of the misfit dislocation segments parallel to the [110] oriented foil, the slight increase in contrast can extend for the length of the foil. Even with the unusual lift-out geometry and contrast enhancements employed here, the strain contrast from the misfit dislocations is very difficult to distinguish from the contrast due to the dots themselves. Without specifically searching for these defects, they are extremely difficult to identify in QD structures imaged in cross-section.
**Full model structure**

(a) Schematic structure of the baseline model structure. (b) The buffer design used for all structures in this paper. Mismatched interfaces where misfit dislocations nucleate are marked with the type of strain relieved. (c) Schematic structure of the trapping layer model structure. See **Experimental Section** for growth details.
Pinning mechanism

Figure S3. (a) Schematic showing a dislocation traveling upwards through a GaAs-based film on Si, pinned in the trapping layer and the QD layer. In the In\textsubscript{0.15}Ga\textsubscript{0.85}As trapping layer and the QD layer, $\tau_{\text{line}}$ represents the shear due to dislocation line tension; $\tau_{\text{misfit}}$, the shear due to lattice mismatch between GaAs and the strained indium-alloyed layers; and $\tau_{\text{alloy}}$, a resistive alloy hardening shear due to alloy compositional fluctuations. (b-c) Rough sketch of the effective stress landscape in (b) the In\textsubscript{0.15}Ga\textsubscript{0.85}As trapping layer where pinning is relatively weak and (c) the QD layer, where the combination of QDs inside a QW results in strong pinning.

We analyze a simplified case of pinning (Figure S3a) where a threading dislocation is completely mobile in GaAs and pinned in both the In\textsubscript{0.15}Ga\textsubscript{0.85}As trapping layer and the QD layer above it. To glide in GaAs, the threading dislocation segment only needs to overcome the short-range, interatomic Peierls stress, $\tau_p$ (~4 GPa in GaAs\textsuperscript{[2]}). This happens readily with relatively small resolved shear stresses either at elevated temperatures or through REDM processes\textsuperscript{[3]} which we exploit with ECCI in Figure 2.

The stress states in the two indium-alloyed layers are more complex. We therefore employ the concept of an effective stress ($\tau_{\text{eff}}$), where, by our convention, the threading segments in these layers can only glide leftward with the free segment in the GaAs if $\tau_{\text{eff}}$ is positive\textsuperscript{[4-7]}. During cooldown, the sub-critical thickness In\textsubscript{0.15}Ga\textsubscript{0.85}As and DWELL layers remain compressively strained—no misfit dislocations form. The threading segments in these layers experience a shear stress ($\tau_{\text{misfit}}$) due to this strain, but since the layers are below critical thickness, $\tau_{\text{misfit}}$ must by definition be smaller than the line tension of the dislocation ($\tau_{\text{line}}$), or more exactly, smaller than the maximum value of $\tau_{\text{line}}$ which is assumed when the threading segment forms a perpendicular kink ($\tau_{\text{line}}$\textperp). Without any additional resistive shear stresses, the shear from the dislocation line tension would normally drag these short threading segments along with it—again, no misfit segments would form\textsuperscript{[7]}. Clearly, this is not the case for either the trapping layer or the QD layer.

For pinning to occur as we observe, there must be an additional stress that adds to $\tau_{\text{misfit}}$ to at least match the magnitude of $\tau_{\text{line}}$\textperp. The source of this additional stress in the trapping layer is alloy hardening ($\tau_{\text{alloy}}$), which arises in certain semiconductor alloys due to natural compositional variations that generate in-layer stress fluctuations. Such hardening effects have been observed in SiGe\textsuperscript{[8]} GaAsP\textsuperscript{[9]} and low-indium InAlGaAs alloys\textsuperscript{[10]} in agreement with our results. Thus, the stress state in the trapping layer resembles that shown in Figure S3b. Here, if the long-range resistive stress field, $\tau_{\text{alloy}}$, is large enough such that at some point $\tau_{\text{eff}} = \tau_{\text{line}} = (\tau_{\text{misfit}} + \tau_{\text{alloy}}) = 0$, then the threading dislocation segment is pinned in the trapping layer, and a trapped misfit dislocation segment will form. Note that the misfit segment cannot simply glide upward through the trapping layer due to the repulsive compressive strain in that layer.

The magnitude of the stress field in the QD layer is substantially larger than in the trapping layer, as shown in Figure S3c. Alloy hardening in the QW once again provides a resistive shear, but, as Beanland et al. have shown, precipitate-like QDs also provide their own resistive shear, pinning threading segments so effectively that they nearly triple the critical thickness for dislocation glide compared to a QW\textsuperscript{[11]}. These effects agree with metallurgical research.
showing that mechanical properties of both elemental metals and alloys become increasingly temperature independent, or athermal, with increasing temperature. This is because their long-range fluctuating stress fields, generated both by compositional fluctuations and structural features such as precipitates and line defects, are no easier to surmount at high temperatures than at low ones.\textsuperscript{[12–16]}

**Full view of ECCI time-lapse**

![ECCI time-lapse images](image)

**Figure S4.** Full view of the first and final frames of the ECCI time-lapse from the baseline structure. The orange arrow marks the growing misfit dislocation exhibited in Figure 2a. See Video S1 for the full time-lapse video.

Figure S4 shows the initial and final frames of the electron-channeling contrast imaging (ECCI) time-lapse sequence (baseline structure). In the full time-lapse video (Video S1), misfit dislocations can be seen forming below the QD layer both by lateral (in plane) threading dislocation glide and by vertical (out of plane) misfit dislocation glide. This latter process is apparent from long segments that appear at once, simultaneously increasing in sharpness and translating very slightly laterally (due to upward glide along inclined \{111\} planes).
Full laser structure

Figure S5. Laser schematics for (a) baseline, (b) TL80 (trapping layers 80 nm from quantum dots (QDs)), and (c) TL180 (trapping layers 180 nm from QDs). See Experimental Section for growth and processing details.
Tilted cross-sectional STEM of TL80

Figure S6. Both (a) and (b) show cross sectional scanning transmission electron micrographs of both misfit and threading dislocations in a TL80 laser. Samples were lifted out along [110] direction and imaged at a tilt ($g = \text{002}$) resulting in certain misfits running parallel to the length of the foil and others running perpendicular to it (marked with white). Due to the tilt, the perpendicular misfits appear as vertical lines and the spacer layers between the quantum dots (QDs) described in Figure S5 disappear among the QD strain contrast. Critically, the misfit dislocations are clearly at different heights than the QDs. Threading dislocations, marked with black arrows, give rise to the misfit dislocations, as described.

Figure S6 shows two images of threading dislocations (marked with black arrows) passing through the QD layers to give rise to misfit segments at the top trapping layer (analogous to Figures 3f-3i). Due to the tilt, although the misfit segments connected to the threading dislocations in both images lie at the trapping layer, they appear to differ in height. From this, we can infer that the misfit segments (marked with white arrows) lie at different depths from the surface of the foil. In both images, we can additionally see misfit dislocations lying in the direction of the foil thickness (also marked with white arrows). From the length of these misfits and tilt angles, we can determine the foil thickness. It is worth noting, in both images, that QDs adjacent to the threading dislocation appear slightly different from the others both in density and in appearance. This confirms that the threading dislocation has not moved from its growth position with the QD layers themselves. The consistency between these images and Figures 3b and 3f-3i provides strong evidence for the success of not only the lower trapping layer, but also the unusual upper trapping layer.
Offcut STEM of misfit dislocations above and below DWELL layers

Figure 7. Bright-field plan-view scanning transmission electron microscopy image of a baseline laser bar (g=220). Inset depicts a cross-section view where QD layers are successively removed along the length of the foil due to the offcut orientation. Orange arrows (main figure and inset) mark the intersections of the QD layers and the foil surface; black arrows indicate misfit dislocations that exit the foil surface just above the top QD layer; white arrows indicate misfit dislocations that exit just below the bottom QD layer. Inset shows a side view of the foil.

Figure S7 shows a BF PV-STEM image of a foil cut out slightly off-normal from the film surface. The sample was prepared milling inwards from the lower n-cladding towards the p-cladding. On right side of the image, all layers are intact and imaged through. Moving leftward, QD layers are successively removed starting with the bottom layer until, on the left side, all QD layers are removed and only the upper p-cladding is being imaged. The five wavy contrast bands (orange arrows) correspond to the locations where the foil cuts through each QD layer. Horizontal misfit dislocations (white arrows) are cut off at the same place as the lowermost QD layer, indicating these dislocations lie at the first QD layer. On the other hand, vertical misfit dislocations (black arrows) are not cut off until the uppermost QD layer is removed, indicating they lie at the uppermost QD layer. The difference in the number and direction of misfit dislocations at the lowermost and uppermost QDs suggest that the relaxation processes occur independently in the n and p cladding.
Large-area PV-STEM of trapping layer structure

Figure S8. Large-area plan-view scanning transmission electron microscopy (PV-STEM) of trapping layer foil highlighting incomplete trapping (white arrows) of threading dislocations by the trapping layer. Inset shows a magnified view of the dislocation displayed in Figure 3h.

Figure S8 shows an aerial view of the trapping layer PV-STEM foil used to create tomographic reconstructions in Figure 3g and Figure 3i. The image shows two perpendicular misfit arrays, lying at the trapping layers. Instances of incomplete pinning that form misfit dislocations at the QD layer occur infrequently (white arrows) and the length of misfit dislocation lying near the QD layer tends to be short. The inset shows the location of the dislocation examined in Figure 3h-i.
Electrically dissipated power at rollover

![Histograms of the electrically dissipated power at rollover for TL80 (red), TL180 (blue) and baseline (black).](image)

**Figure S9.** Histograms of the electrically dissipated power at rollover for TL80 (red), TL180 (blue) and baseline (black).

Figure S9 shows histograms comparing the electrically dissipated power at laser rollover (the current at which lasing output begins to decrease with current due to the onset of excited state lasing) between the three laser designs. Although there is some spread in the data, both TL80 and TL180 laser designs show approximate twofold increase over baseline. We attribute the spread in the data to processing variability as described in the discussion of Figure 4.
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