Tensile strained island growth at step-edges on GaAs(110)

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We report the growth of tensile strained GaP islands on a GaAs(110) surface. Three-dimensional island formation proceeds via a step-edge nucleation process. To explain the dislocation-free nature of these islands, we consider the kinetics of strain relief within the context of a model for dislocation glide as a function of surface orientation and sign of strain. © 2010 American Institute of Physics. [doi:10.1063/1.3498676]

The majority of research into self-assembled islands offering quantum confinement has focused on compressively strained materials grown on (001) substrates. However quantum confinement within tensile strained materials has also been investigated. The general trend within the III-V semiconductors of increasing band gap with smaller lattice constant would seem to preclude quantum confinement within tensile materials. However, confinement of one or more dimensions well below 1 ML, as shown in Figs. 1(b)–1(f) cannot be an artifact of anion exchange reactions and arise solely from P$_2$ and Ga codeposition. The tensile strained islands self-assemble at GaP thicknesses well below 1 ML, as shown in Figs. 1(b)–1(d). As a result there can be no continuous GaP wetting layer present below the islands. The VW mechanism describes 3D island growth in the absence of wetting layer formation and was reported for tensile Si/Ge(111). VW growth is anticipated when $\gamma_{\text{film}} > \gamma_{\text{substrate}}$, where $\gamma$ is the surface energy. From the limited literature data available, $\gamma_{\text{GaP}}(110) \approx \gamma_{\text{GaAs}}(110)$ (0.61–1.23 eV/atom and 0.57–0.63 eV/atom, respectively). Implies that VW growth may be expected. However, a surface

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energy description of island nucleation such as this does not account for surface inhomogeneities such as steps. We hypothesize that step-edges mediate island nucleation. Figures 1(g) and 1(h) show that GaP islands form only at step-edges on the GaAs(110) buffer surface, implying that Ga adatoms do not remain in the middle of terraces long enough to condense into islands. Such behavior may be plausible given that Ga adatom mobility on GaAs(110) is thought to be 10–15 times larger than on GaAs(001). Adatoms reaching a step-edge encounter a potential barrier limiting their migration to neighboring terraces. The result is locally enhanced adatom concentration at the step-edges. Hence, at 460 °C, even though total GaP coverage is ≲1 ML, adatom density becomes sufficiently high at step-edges to trigger island nucleation.

Following nucleation at step-edges, islands grow continuously without the formation of further islands. In Fig. 2(a) we see that mean island height and diameter both increase as a function of GaP thickness, whereas Fig. 2(b) shows that island density remains constant to within error for the range of GaP thicknesses studied here. By changing Ga flux exposure time we demonstrate control of island size over the ranges 3–12 nm in height and 40–200 nm in diameter. A mean island density of ~7 × 10^7 cm^-2 is significantly lower than typical densities of >10^9 cm^-2 in InAs/GaAs(001) (Ref. 1) and Ge/Si(001). Also supporting the argument that no wetting layer is formed, Fig. 2(c) shows that total island volume per unit area is linearly dependent on GaP thickness. The intercept extremely close to the origin confirms that all deposited GaP is incorporated in islands.

Step-edge mediated nucleation has been previously linked with reduced dispersion in island size, and could explain the uniformity of our GaP islands. A histogram of island heights measured on the 6.4 ML GaP/GaAs(110) sample is shown in Fig. 2(d). It exhibits a single peak, confirming an absence of the bimodal distribution often observed in (001)-oriented systems; monomodal distributions of island size were seen for all our samples. Narrow, monomodal size distributions have been frequently observed in InAs/GaAs(001) (Ref. 21) and Ge/Si(001) (Ref. 18) but the thermodynamic and kinetic factors leading to the distributions in GaP/GaAs(110) remain unclear. Despite this, monomodal distributions of island size are consistent with heterogeneous nucleation at step-edges, followed by island growth.

The GaP islands form a coherent interface with the GaAs(110) buffer. Figure 3(a) shows a high-resolution XTEM image of a single 0.7 ML GaP island. g-b invisibility analysis of ~10 such islands reveal that all are dislocation-free. Relaxation in highly mismatched materials can be treated as a competition between elastic deformation via stain-induced roughening, and plastic deformation via dislocation nucleation. For growth of coherent islands, the onset of roughening must precede dislocation nucleation. Most strain-relieving dislocations in zinc-blende materials are 60° dislocations that can dissociate into a pair of 30° and 90° Shockley partial dislocations, since such a reaction reduces the net energy by up to a third. For strain relief, the

### References
1. Ref. 1
2. Ref. 2
3. Ref. 3
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**Figures**

FIG. 1. (Color online) [(a)–(f)] (1 μm)^2 AFM images showing evolution of 3D GaP/GaAs(110) islands as GaP thickness is increased from 0.0 to 6.4 ML (z scale: 20 nm). (g) (25 μm)^2 AFM image of 6.4 ML GaP/GaAs(110) (z scale: 5 nm): island formation is preferred at terrace edges parallel to [112]. (h) (5 μm)^2 AFM image of 0.7 ML GaP/GaAs(110) (z scale: 2 nm) demonstrating the uniformity typical of these islands.

FIG. 2. (Color online) (a) Island height (black squares) and diameter (red triangles) as a function of GaP thickness (curves are guides to the eye). (b) Island density with GaP thickness (dashed line is the mean). (c) Total island volume per unit area with GaP thickness. The dashed line shows a regression fit to these data. (d) Example histogram of island heights for the 6.4 ML GaP sample, with a variance of 1.7 nm.

FIG. 3. (a) On-pole XTEM image of a single 0.7 ML GaP island. A lack of dislocations and other defects confirms its crystalline coherence with the underlying GaAs(110) buffer. (b) On-pole XTEM image of a single 6.4 ML GaP island. Dark areas indicate heavily dislocated regions. Note the different scale bars in (a) and (b).
Thompson tetrahedron\textsuperscript{23} shows that one partial must glide first or “lead” to prevent high-energy A|A stacking fault formation, followed by the other which “trails” to complete the lattice displacement.

To explain our observation of dislocation-free, tensile islands on (110) we use the Kvam–Hull model in which plastic relaxation behavior is dictated by both surface orientation and sign of strain.\textsuperscript{11} According to Kvam and Hull, direct nucleation of a 90° partial in tensile (110)-oriented material would result in an A|A stacking fault and so is forbidden; the 30° partial must therefore lead. The trailing 90° partial experiences greater resolved shear stress than the 30° partial, which prevents the 30° and 90° partials from separating. The two partials thus behave essentially as a single 60° dislocation. The Kvam–Hull model shows that this situation is symmetrically equivalent to the well-known case of compressive strain on a (001) surface;\textsuperscript{11} direct nucleation of 90° partials is forbidden and hence plastic deformation must take place via nucleation and glide of 60° dislocations.

While the Kvam–Hull model predicts that 60° dislocations will cause plastic deformation in GaP/GaAs(110), numerous theoretical and experimental studies show that 60° dislocation nucleation is kinetically limited. The dislocation critical surface half-loop nucleation energy, $E_{\text{nuc}}$, is proportional to $|b|^3$, where $b$ is the Burger’s vector.\textsuperscript{24} Computation of accurate values for $E_{\text{nuc}}$ is problematic due to both difficulties in estimating the dislocation core parameter and the possibility of adding or removing a surface step.\textsuperscript{12} Nevertheless, $E_{\text{nuc}}$ for a 60° dislocation ($E_{\text{nuc60°}}$) in ∼4% compressively strained, (001)-oriented material is calculated to be 10–100 eV for typical growth conditions.\textsuperscript{12} The dislocation nucleation rate is proportional to $\exp(-E_{\text{nuc}}/kT)$ and hence the high value of $E_{\text{nuc60°}}$ leads to an extremely low dislocation nucleation rate for a wide range of growth conditions. Since 60° dislocation nucleation is kinetically limited, the preceding strain relief mechanism becomes elastic deformation via strain-induced roughening, provided adatom mobility is sufficiently high. The result is the formation of dislocation-free, 3D islands. Since we observe dislocation-free GaP/GaAs(110) islands [Fig. 3(a)], we infer that, here too, 60° dislocation nucleation is kinetically limited.

Above some size limit, the elastic energy stored in an island overcomes the metastable barrier to dislocation nucleation and dislocations form. Contrast modulation is revealed in XTEM analysis of a large, 6.4 ML GaP island [Fig. 3(b)]. Dark regions within and below the GaP island indicate areas with high dislocation density and confirm incoherence with the GaAs(110) buffer. Further work will be necessary to determine the boundary between coherence and incoherence for GaP/GaAs(110) islands.

In conclusion, we have demonstrated the formation of self-assembled GaP islands at GaAs(110) step-edges. Based upon the evolution of their size, density, and total volume as a function of GaP thickness, we suggest that islands nucleate at step-edges before the completion of a complete monolayer, and subsequently grow without further nucleation. The coherent nature of these tensile strained GaP islands on a (110) surface implies a high dislocation nucleation energy, and is consistent with the Kvam–Hull model of strain relaxation.

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