

The effect of substrate growth area on misfit and threading dislocation densities in mismatched heterostructures

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We show that mismatched epitaxial growth on small growth areas results in metastable, low dislocation density films. By considering methods of dislocation nucleation in mismatched epitaxial films, we can explain the effect of growth area on the kinetics of misfit dislocation nucleation. We demonstrate this reduction of misfit dislocation density by growth on small areas using the InGaAs/GaAs system, and we also demonstrate a reduction in threading dislocation densities in the GeSi/Si system. InGaAs/GaAs interfaces which have a misfit dislocation density greater than 5000 dislocations/cm for large growth areas are grown nearly misfit-dislocation-free on small growth areas. GeSi/Si structures which have a threading dislocation density of approximately 1000 dislocations/cm² are grown with a 100% threading dislocation density reduction for small growth areas. We discuss the limits of growth on small areas to reduce both interface dislocation densities as well as threading dislocation densities.

I. INTRODUCTION

Lattice mismatched systems have been investigated for a variety of devices.¹⁻⁷ In many of these applications, mildly lattice-mismatched systems are used and the epilayer thicknesses are kept below the critical thickness for defect formation, resulting in totally elastically strained layers. Extension of the coherently strained approach to layers with larger lattice-mismatch and thicker epilayers is of considerable technological interest.

Despite the continued interest in mismatched semiconductor systems, the nucleation mechanisms for misfit dislocations are still relatively unexplored. Most studies of misfit dislocation formation are based on postgrowth observations, and the origin of misfit dislocations must therefore be deduced from dislocation structure and the thickness at which dislocations are first observed,⁸⁻¹⁰ often called the critical thickness. But the criterion of critical thickness is a vague one, due to the varying ability of different measurements to detect the onset of misfit dislocation formation,¹¹ the metastable nature of the epilayers from different systems,¹² and comparison with incorrect theoretical critical thickness expressions.¹³

If experimental results are compared with the critical thickness first proposed by Matthews *et al.*,^{13,14} the metastable nature of the films becomes apparent since the experimental critical thicknesses are greater than Matthews' theoretical critical thickness,^{13,15} especially when the mismatch is less than 2%. A consequence of this metastability due to the kinetics of misfit dislocation formation is that the interface dislocation density depends on the size of the growth area. It has recently been shown that by reducing the growth area on the substrate before the mismatched semiconductor is deposited, one can reduce the number of, or even eliminate, misfit dislocations at mismatched In_{0.05}Ga_{0.95}As/(001) GaAs (mismatch approximately 0.4%) heterostructure interfaces.¹⁶ These results make possible a number of previously unattainable devices, and they demonstrate that growth area is a variable which can be used

to study and determine mechanisms of misfit dislocation nucleation.

II. MOTIVATION FOR DISLOCATION DENSITY REDUCTION

The applications for mismatched epitaxial films can be grouped into two categories: (i) the utilization of the mismatched interface itself [Fig. 1(a)]; (ii) employing the substrate as a template for the growth of subsequent, mismatched device layers [Fig. 1(b)]. The first application is limited primarily by the introduction of misfit dislocations at the heterointerface. Misfit dislocations degrade the electrical transport of carriers in both the vertical and parallel directions to the interface. In a heterojunction bipolar transistor (such as AlGaAs/InGaAs/GaAs), carrier motion is perpendicular to the heterointerface. Even though a low density of misfit dislocations can exist at the InGaAs/GaAs interface without affecting device performance, a degradation in transistor characteristics has been observed when a high density of dislocations is present at the interface.⁷

In parallel transport devices such as GaAs/InGaAs/GaAs high electron mobility transistors, the In concentration is typically less than 20% and the InGaAs layer thickness is below the critical thickness to avoid the formation of deleterious misfit dislocations.¹

The threading dislocation density is of primary concern in the second application. When a substrate is used as a template for a highly mismatched epilayer, the epilayer can be crystalline, yet the threading dislocation density is very high (10⁶–10⁹ cm⁻²). This is observed in such high mismatched systems as GaAs/Si¹⁷ and InP/Si¹⁸. The goal in this case is to reduce the threading dislocation density to 10³–10⁴ cm⁻² so that subsequent device layers will possess dislocation densities similar to those in homoepitaxial device layers.

As we will show, growth on small substrate areas shows experimental and theoretical promise for reducing both misfit and threading dislocation densities.

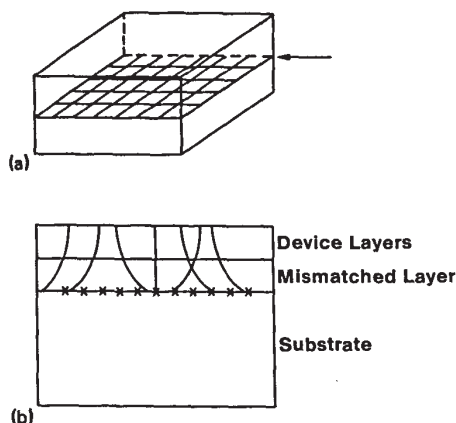


FIG. 1. A diagram showing the two applications for mismatched epitaxy. (a) Utilization of highly strained interface. (b) Employing a mismatched substrate as a template.

III. DISLOCATION SOURCES

To understand the effect of growth area on defect morphology, we first discuss the mechanisms of misfit dislocation nucleation. We categorize the dislocation sources into four groups: (i) fixed sources (substrate dislocations and substrate imperfections); (ii) heterogeneous surface half-loop nucleation; (iii) dislocation interactions and multiplication; (iv) homogeneous surface half-loop nucleation. It should be noted that group (i) and (ii) overlap somewhat, since substrate imperfections typically nucleate dislocation half-loops. However, the second group is categorized separately since epilayer surface imperfections may produce heterogeneous surface loops not related to substrate imperfections. We will briefly discuss each group. Details of the nucleation processes can be found in Ref. 19.

A. Fixed sources

We define fixed sources as those sources which have a constant density per unit area, i.e., the number of fixed sources scales with area. One example of a fixed source is the substrate dislocations, which are inherited as threading dislocations in the overlayer. A certain density of these substrate dislocations exist (dislocations/cm²) and are nucleation sites for misfit dislocations.

Consider the left-hand side of Fig. 2, which illustrates the condition necessary for the formation of a misfit dislocation from a threading dislocation. The strain in the mismatched overlayer produces a force on the dislocation segment in the overlayer (F_e). F_e is the driving force for the glide of the threading segment. However, opposing F_e is (i) the increase in the line tension of the dislocation line due to the creation of the misfit dislocation segment at the interface; (ii) the Peierl's force opposing dislocation motion. Thus, it is possible to obtain an expression which defines the thickness at which the dislocation will glide in the overlayer, producing a misfit dislocation at the interface (i.e., the critical thickness).^{13,20}

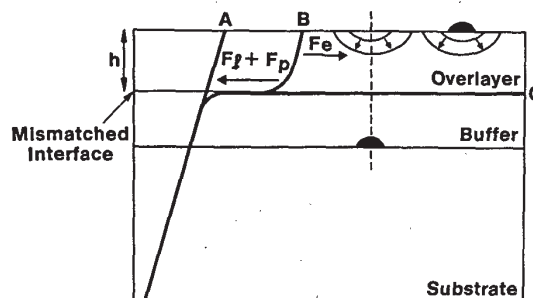


FIG. 2. An illustration depicting the nucleation of misfit dislocations via fixed sources and heterogeneous surface half-loop nucleation: left-hand side, threading dislocations; middle, substrate surface inhomogeneities; right-hand side, epilayer surface inhomogeneities.

In addition to a certain density of substrate dislocations which can create misfit dislocations, there also exists a certain density of inhomogeneities on the substrate surface before growth, such as particulates, impurities, and possibly mechanical damage.¹⁹ This is shown schematically in the center of Fig. 2. A substrate surface inhomogeneity will result in stress concentrations through the buffer layer and into the epilayer. In some cases the particulate or impurity may be larger than the buffer layer itself. The resulting stress concentration lowers the activation energy for nucleation of dislocation half-loops from the surface.

B. Heterogeneous surface half-loop nucleation

The right-hand side of Fig. 2 illustrates the nucleation of a dislocation half-loop due to an epilayer surface imperfection. Although this source has not been verified experimentally, it is possible that particulates or impurities from the sources or vacuum chamber could deposit on the epilayer surface during growth. Also, a drastic change in growth mode (i.e., roughening instead of two-dimensional growth) could lead to heterogeneous surface loop nucleation. Such a nucleation process is difficult to model theoretically, although recent work has shown that the activation energy is indeed lowered by such surface inhomogeneities.²¹

C. Dislocation interactions and multiplication

As growth on small areas indicates,^{16,19} dislocation interactions are a major factor in producing high threading and misfit dislocation densities. If we first consider dislocation interactions, one can imagine several types, two of which are shown schematically in Fig. 3. In a higher misfit system in which many misfit dislocations are forming, dislocations with same-sign Burgers vectors (b_1) may interfere with each other during the misfit dislocation generation process. Thus, the repulsion between threading dislocation segments results in an increased threading dislocation density.

One can also imagine dislocation segments with Burgers vectors such that a very strong attraction exists between the threading segments (b_1 and b_2). In this case, the dislocation attraction will result in a new threading segment with Burgers vector b_3 . In diamond and zinc-blende structures, b_1

and b_2 are 60° dislocations which react to form b_3 , an edge dislocation. Although this second reaction reduces the threading dislocation density by 50%, the resulting sessile edge dislocation has little chance of exiting the system; thus, it is a permanent threading dislocation.

Both interactions above are schematically illustrated in Fig. 3 such that the interacting misfit dislocations lie on perpendicular $\langle 110 \rangle$ directions. We have considered the interaction of parallel misfit dislocations previously, and have shown that groups of parallel misfit dislocations with the same sign Burgers vector can repel dislocations away from the interface, whereas parallel dislocations with complementary Burgers vectors can react at a distance away from the interface to produce an edge dislocation.^{22,23} Such interactions and reactions between parallel misfit dislocations can restrict their threading segments from reaching a free edge.

Recently, real-time analysis of misfit dislocation formation in the transmission electron microscope has shown that a gliding threading segment is impeded when it crosses a perpendicular misfit dislocation.²⁴ This also suggests that dislocation interactions will increase threading dislocation densities.

Dislocation interactions will lead to an increase in threading dislocation densities; however, dislocation multiplication will increase both threading and misfit dislocation densities. The earliest example of dislocation multiplication in strained layers was described by Hagen and Strunk.¹⁰ Their multiplication scheme is illustrated in Refs. 10 and 19. Essentially, intersections between perpendicular misfit dislocations with the same Burgers vector results in the creation of two free-ended threading segments which can glide to create additional misfit dislocation length in the interface. This multiplication scheme requires a thin epilayer; therefore, if dislocation intersections can be minimized when the film is thin, Hagen-Strunk multiplication will not occur. In high misfit systems, one can imagine dislocation multiplication occurring through the intersections of the threading segments in the epilayer.¹⁹

A new form of dislocation multiplication has been observed in the SiGe/Si system.²⁵ The center of the multiplication mechanism is a diamond-shaped defect which is able to punch-out misfit dislocation half-loops along $\langle 110 \rangle$ direc-

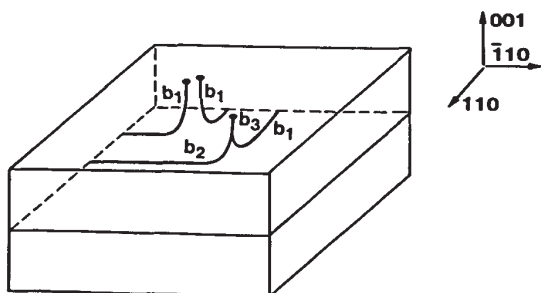


FIG. 3. A repulsive and attractive interaction between threading segments during the formation of misfit dislocations. Although schematically drawn, the threading segments lie on $\{111\}$ planes in $\langle 110 \rangle$ directions.

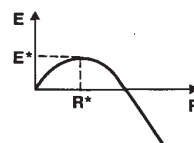
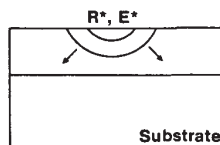


FIG. 4. A diagram illustrating the homogeneous nucleation of surface half-loops. The bottom graph qualitatively shows the relationship between the radius and energy of the loop. The critical radius is R^* and the activation energy is E^* .

tions. However, the origin of such a source is currently unknown.

From the above discussion, it is critical to limit dislocation interactions and multiplication to prevent high threading and misfit dislocation densities in heteroepitaxy. As we will show, growth on small areas is the likely candidate to solve this problem.

D. Homogeneous surface half-loop nucleation

The last category is homogeneous surface half-loop nucleation. This mechanism is shown in Fig. 4, and is essentially the same as heterogeneous surface half-loop nucleation except no inhomogeneity is necessary.

We can calculate the activation energy, E^* , and the critical radius, R^* , of the loop which will spontaneously grow to form a misfit dislocation segment at the interface and two threading segments in the unrelaxed overlayer.^{19,20} The activation energy for the nucleation of the half-loop in the InGaAs/GaAs system at 550°C is plotted in Fig. 5. An increase in In concentration decreases the activation energy due to the corresponding increase in elastic strain. The solid,

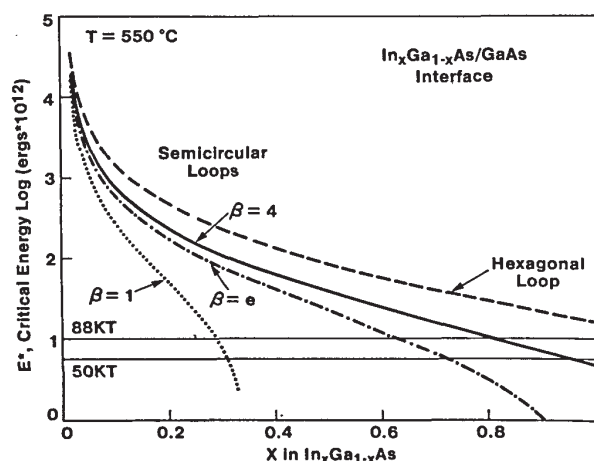


FIG. 5. A graph of the activation energy for homogeneous nucleation vs In composition in a InGaAs/GaAs heterostructure. The available energy in the system for the process is indicated by 50 and 88 kT. The dashed line is the activation energy for a hexagonal half-loop; the other curves are the activation energy for semicircular half-loops with different values for the core energy parameter.

dashed-dot, and dotted activation energy curves are calculated for a semicircular dislocation half-loop using different values for the core energy parameter (β). For the diamond cubic and zinc-blende systems, $\beta = 4$.²⁶ The dashed curve represents the activation energy for the nucleation of a semi-hexagonal loop with $\beta = 4$.

We can determine the In concentration where homogeneous nucleation of loops will occur by locating the point of intersection between the activation energy curves and the available energy curves at 550 °C. The available energy curves are labeled 50 and 88 kT, the average available thermal energy for an average loop size.²⁰ As Fig. 5 shows, a semicircular loop with $\beta = 4$ will nucleate homogeneously at $x = 0.8$, corresponding to approximately 6% strain. The semi-hexagonal loop never intersects the available energy curves, indicating that the homogeneous nucleation of a hexagonal loop in this system is unlikely.

Note the extreme sensitivity of homogeneous surface nucleation on the core energy parameter; if the dislocation core energy is ignored ($\beta = 1$), the nucleation point occurs at $x = 0.2$, approximately 2% strain, instead of the 6% strain for the $\beta = 4$ case. Also note the very high strain necessary for the nucleation of half-loops without the aid of inhomogeneities. Thus, if the strain level is below 2–6% strain, homogeneous nucleation is unlikely.

IV. HIERARCHY OF NUCLEATION PHENOMENA

As mentioned in the Introduction, there is a controversy about the critical layer thickness due to the varying ability of different instruments to detect the onset of misfit dislocation formation, and the metastability of epilayers from different systems. Also, it is quite difficult to correctly model the degree of relaxation in the film after the critical thickness has been exceeded. This is not surprising considering the multitude of nucleation sources discussed in the above sections. An additional point is that each source has a different activation energy for the nucleation of a misfit dislocation.

Consider Fig. 6, which is a diagram of the activation energy hierarchy of misfit dislocation sources. The vertical axis qualitatively represents the activation energy of the source, which depends on growth conditions (T), as well as the strain state of the film (ϵ). At low growth temperatures and strain levels, only the lowest activation energy sources will be active: fixed sources, at a minimum density of 10^3 – 10^4 cm^{-2} . When these sources are active, some dislocation interactions and multiplication are possible, resulting in increased threading and misfit dislocation densities. However, at higher strain levels and higher temperatures, heterogeneous and eventually homogeneous surface half-loops become the major sources. With a multitude of dislocations forming at the surface, and with the enormous increase in dislocation interactions and multiplication, the epilayer develops a high threading and interface dislocation density.

Figure 6 shows that if growth must occur at higher temperatures, the strain level must be kept low to avoid the higher activation energy processes. Also, growth on small areas would limit the number of dislocation interactions and multiplication events.

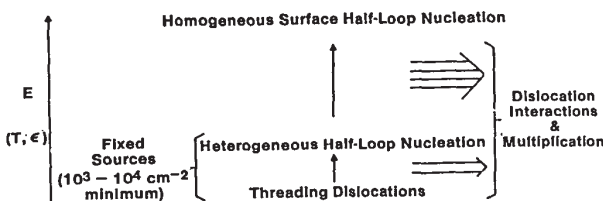


FIG. 6. An illustration showing the activation energy relationship between the nucleation sources of misfit dislocations.

If growth can occur at lower temperatures on substrates with a low density of fixed sources, one expects slow relaxation since the higher activation energy processes will be inactive. This is the case for GeSi/Si,^{12,15} and more recently, GeSn/InSb,²⁷ where metastable, high-strain films are observed.

V. THE EFFECT OF GROWTH AREA ON DEFECT MORPHOLOGY

A. Misfit dislocations

To date, the only methods (besides the current study) of reducing misfit dislocation densities for application I in Fig. 1 is lattice-matching or keeping the epilayer thickness below that of the critical layer thickness. Due to the limited number of lattice-matched systems, and because of the limited thickness and strain required for coherently strained growth, the design options for heterostructures are restricted.

Growth on small areas relaxes these restrictions in the following manner. Figure 6 shows that the homogeneous nucleation of dislocation half-loops must be inoperative to prevent high dislocation densities. Also, dislocation multiplication and interactions must be limited. Consider a case where the strain in the growing epilayer is low enough ($< 2\%$ – 6%) so that homogeneous surface nucleation is not an active source. In this case, the only operative sources for misfit dislocations are the fixed sources (substrate dislocations and inhomogeneities).

Figure 7 illustrates the influence of growth area on misfit dislocation densities when only fixed sources are active. The black dots on the surface of the growth areas represent fixed sources. For large area growth, as the mismatched overlayer is deposited, numerous fixed sources produce many misfit

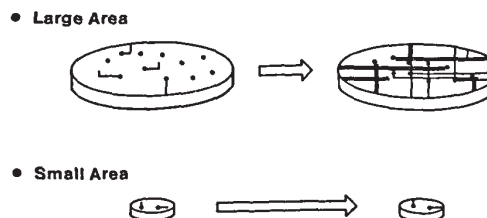


FIG. 7. A diagram of the advantages of growth on small areas when only fixed sources (black dots) are operative.

dislocations, and each fixed source is able to create long lengths of misfit dislocation. The long lengths of misfit dislocations produce many dislocation interactions and multiplication events, further increasing the misfit dislocation density (as well as threading dislocation densities).

Now consider growth on the small areas. It can be shown that the misfit dislocation density, for a sample in which only fixed sources are operative, is proportional to: (i) the lateral dimension of the sample; and (ii) the areal density of fixed sources.^{8,19} This can be viewed qualitatively in the following way. On a small growth area, the number of fixed sources present which can relax the film is small. Thus, by reducing the size of the growth area, one can continue to decrease the available nucleation sources. If the growth area is held constant, then a reduction in the areal density of fixed sources will produce an equivalent effect.

However, another important aspect of growth on small areas is that each fixed source within a small area can only produce a short length of misfit dislocation in the interface. This short length of misfit dislocation from each source virtually eliminates dislocation interaction and multiplication. Thus, the final misfit dislocation density is much lower than that in the large growth area.

B. Threading dislocations

A popular method to reduce threading dislocation densities is the strained-layer superlattice (SLS). However, for large area growth, the ability to reduce threading dislocation densities is extremely limited for the following reasons.

Firstly, in order to escape at the edge of a large wafer, the threading segment must travel long distances (cm) and create equally long misfit dislocations. This encourages dislocation interactions and multiplication, and it is likely that the interface dislocation density is increased as well as the threading dislocation density.

Secondly, the creation of long lengths of misfit dislocations implies that a single fixed source can relieve a relatively large amount of elastic strain; therefore, only a small number of threading dislocations will reach the edge of the growth area before all the elastic strain is zero.

To demonstrate the ideas above, consider a square mesa or substrate of edge length L with a mismatched overlayer of misfit f . The linear interface dislocation density will be⁸

$$\rho_{l(110)} = \rho_d \cdot \left(\frac{1}{2}\right) \cdot \left(\frac{L}{2}\right) = \frac{\rho_d \cdot L}{4}, \quad (1)$$

where ρ_l is the interface dislocation density along one $\langle 110 \rangle$, and ρ_d is the threading dislocation density inherited from the substrate. The factor of $\frac{1}{2}$ is necessary since the threading segment can glide in either in-plane $\langle 110 \rangle$ direction, and the average length of a misfit dislocation is $L/2$. Equation (1) is only valid when $\rho_l < b_{\text{eff}}/f$ and dislocation interactions are limited. The interface dislocation density is the inverse of the dislocation spacing, which is related to the elastic strain in the overlayer by

$$\frac{1}{\rho_l} = S = \frac{b_{\text{eff}}}{\delta} = \frac{(b/2)_{\text{eff}}}{f - \epsilon}, \quad (2)$$

where b_{eff} is the in-plane effective Burgers vector ($b/2$ for 60° dislocations), δ is the plastic deformation, and ϵ is the elastic strain. Solving for the relationship between strain and threading dislocation density using Eqs. (1) and (2) gives

$$\epsilon = f - \frac{\rho_d L b}{8}. \quad (3)$$

To avoid the nucleation of new dislocations from higher activation energy sources, we will consider interfaces with a misfit f of less than 1%. We can solve for the maximum number of dislocations to be "filtered-out" by setting $\epsilon = 0$, i.e., the point where no elastic strain is left in the overlayer to force the glide of the threading segment. Using $b = 4 \text{ \AA}$, we obtain

$$\rho_{d_{\text{max}}} = \frac{2 \times 10^6}{L(\text{cm})}. \quad (4)$$

Thus, let us consider a single strained layer on a 1 cm^2 sample. The maximum density of threading dislocations which can reach a free edge is approximately 10^6 cm^{-2} . This is a rather optimistic value since many dislocation interactions will occur in a 1 cm^2 sample. Now consider a smaller growth area of $10 \mu\text{m}$. The maximum density that can be filtered is 10^9 cm^{-2} . Also, because of the short length the dislocation must travel to reach the growth area edge, little or no dislocation interactions will occur. Thus, a small amount of strain in a small growth area can filter large numbers of threading dislocations.

VI. EXPERIMENTAL RESULTS

We briefly describe two examples of employing small growth areas to reduce misfit and threading dislocation densities. Details of the experiments can be found in Refs. 16, 19, 28, and 29.

Figure 8 illustrates the sample configuration for the patterned mesa experiments. Before molecular-beam epitaxy (MBE) growth, the substrate (Si or GaAs) is vertically patterned to form a multitude of mesas, varying in lateral dimension and shape. The trenches in between the mesas were $2 \mu\text{m}$ in depth, so that the material on top of the mesas would be isolated from the deposited material in the trenches. The lateral dimension of the mesa varied from 2 to $400 \mu\text{m}$, and the mesa shapes were circular, square, and rectangular. In the InGaAs/GaAs case, we also patterned substrates with different substrate dislocation densities (10^2 – 10^5 cm^{-2}). With MBE, a buffer layer of GaAs was grown on the patterned GaAs, followed by a $3500 \text{ \AA}$ layer of $\text{In}_{0.05}\text{Ga}_{0.95}\text{As}$

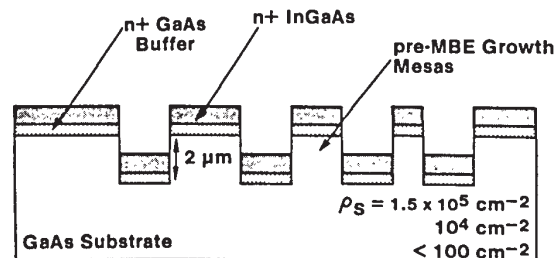


FIG. 8. The sample configuration for the mesa experiments.

(0.4% mismatch). In the GeSi/Si experiments, MBE or chemical vapor deposition was used to deposit a Si buffer layer, followed by a 2100 Å thick $\text{Ge}_{0.15}\text{Si}_{0.85}$ layer (0.6% mismatch). The compositions for the InGaAs/GaAs and GeSi/Si structures were chosen so that the strain in the epilayer will be much less than the 2%–6% strain necessary for homogeneous surface half-loop nucleation. Thus, only fixed sources should be operative.

A. InGaAs/GaAs

Figure 9 is a plot of the linear interface dislocation density versus the mesa diameter of the circular mesas fabricated on a GaAs wafer with a substrate dislocation density of $1.5 \times 10^5 \text{ cm}^{-2}$. The linear dislocation density is simply the inverse of the average misfit dislocation spacing. Cathodoluminescence was used to image the low density of misfit dislocations, which appear as dark lines in the cathodoluminescence image.^{16,19} By counting the number of dislocations in a large number of mesas, the average linear interface dislocation density can be obtained. The error in the dislocation densities is approximately 100 dislocations/cm for the 400 μm mesa, and increases to 200 dislocations/cm for the 25 μm mesa. The linear interface dislocation density was much greater than 5000 dislocations/cm for 1 cm^2 samples.

Examination of Fig. 9 shows that the interface dislocation density, even on mesas as large as 400 μm , is significantly smaller than the large area (1 cm^2) reference sample described above. Figure 9 shows that the dislocation density along the two $\langle 110 \rangle$ directions in the interface is different. This can be attributed to the different crystallographic nature of dislocations in zinc-blende structures which lie along perpendicular $\langle 110 \rangle$ directions.³⁰ Also note that the linear interface dislocation density decreases in a linear fashion as the mesa width decreases, and that the linear fits nearly pass through the origin of the plot. The linear decrease in interface dislocation densities suggests that dislocation multipli-

cation is not occurring, since a superlinear dependence of interface dislocation density on mesa width would be expected if multiplication were active. The near-zero y -intercept also suggests that only fixed sources are active; i.e., those sources which scale with area as previously discussed.

Using the linear interface dislocation density versus mesa diameter shown in Fig. 9, we can calculate the density of fixed misfit dislocation nucleation sources. We arrive at a value of $2 \times 10^5 \text{ cm}^{-2}$, remarkably close to that of the substrate dislocation density. This is further evidence that only fixed nucleation sources are active, and that substrate dislocations are the dominant fixed source in this sample.

If the substrate dislocations are the dominant nucleation source, then by decreasing the substrate dislocation density, the slope of the linear fits in Fig. 9 should decrease. This is indeed observed in heterostructures grown on the GaAs substrates with a dislocation density of 10^4 cm^{-2} .

By keeping the strain level below the value for homogeneous surface loop nucleation, and by virtually eliminating dislocation interaction and multiplication, we have drastically reduced the interface dislocation density with a reduction in growth area.

B. Ge/Si/Si

Our preliminary results from the GeSi/Si experiments indicate that small growth areas are also effective at reducing threading dislocation densities, although transmission electron microscopy (TEM) results are needed to confirm the wet chemical etching results described below.

The threading dislocations in the mesa structures were revealed using a 4:5 volume ratio of 48% HF:0.3M CrO_3 .³¹ Figure 10 shows the percentage of threading dislocation density reduction versus mesa size. The threading dislocation density for unpatterned areas averaged approximately 1000 dislocations/ cm^2 . As Fig. 10 shows, the threading dislocation density decreases with a decrease in mesa width as ex-

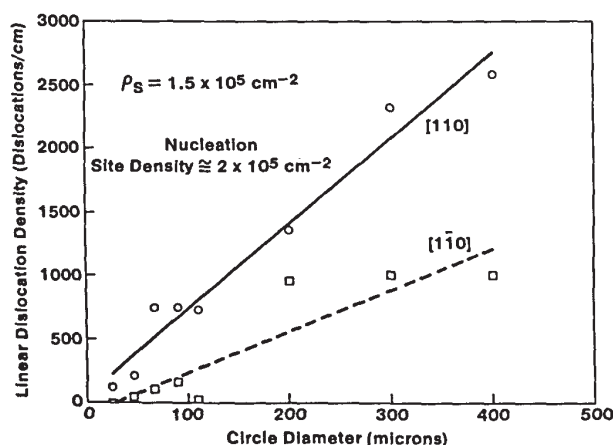


FIG. 9. A graph of linear interface dislocation density vs circular mesa diameter along both $\langle 110 \rangle$ directions in the $\{100\}$ interface between $\text{In}_{0.05}\text{Ga}_{0.95}\text{As}$ and GaAs.

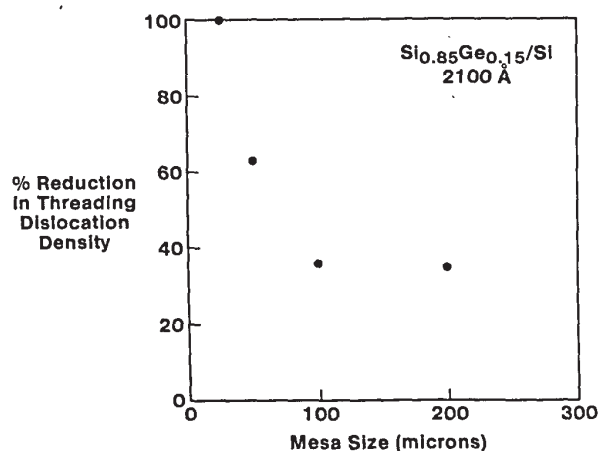


FIG. 10. A graph of the percent reduction in hillock density (threading dislocation density) as a function of mesa size for a $\text{Ge}_{0.15}\text{Si}_{0.85}/\text{Si}$ heterostructure. The unpatterned heterostructure area contained 1000 dislocations/ cm^2 .

pected, and no threading dislocations were observed with this chemical etching technique on the 25 μm mesas.

This reduction in threading dislocation densities must occur if only fixed sources are operative, since all threading segments will glide to form misfit dislocations until the threading segment escapes at the mesa edge.

VII. CONCLUSIONS

We have shown that misfit and threading dislocation densities can be reduced in mismatched heterostructures by growing on small areas. The tremendous advantage of growth on small areas is derived from the short misfit dislocation lengths generated by each dislocation source, and the elimination of dislocation interactions. Extension of this technique to higher misfit systems may lead to higher quality, high misfit, epitaxial layers.

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