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# COHERENT STRAIN AND CLUSTERING IN Ge / Si HETEROEPITAXY

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## Abstract

The Ge / Si heteroepitaxial system grows in the Stranski - Krastanov (layer + island) growth mode with an equilibrium intermediate layer thickness of 3 monolayers. Initially, coherent (dislocation free) islands form above the intermediate layer. These coherent islands relieve part of their lattice mismatch by elastic deformation of the layer + island + substrate system. After cluster growth, these islands may reach a critical radius above which it becomes energetically favorable for strain relief through the introduction of misfit dislocations. Particle size distributions generated from digitally acquired secondary electron images of Ge / Si(100) films grown *in situ* in an ultra - high scanning transmission electron microscope (UHV - STEM) have been used to study particle coarsening processes. By exploiting the large magnification range available it is possible to obtain reliable statistics for islands with radii ranging from ~2nm to over 500nm. These size distributions show that coherent islands significantly impact coarsening processes in this system. In all cases studied, the coherent islands compete less effectively for the diffusing adatoms and consequently grow much more slowly than the dislocated islands. This difference in the growth rate between coherent and dislocated islands is due to the extra energy required for increasing the strain field of growing coherent islands. The reduction of interfacial strain via the introduction of misfit dislocations is shown to relieve a substantial fraction of the elastic energy and lowers the energy cost per additional adatom in the growing cluster. Preliminary results of a continuum elasticity calculation which approximates coherent island growth confirm that elastic deformation of the substrate is a viable means of strain relief for these elastic systems.

Key Words : coherent islands, clustering, particle size distributions, elasticity, strain energy, epitaxy, film growth, Stranski-Krastanov, dislocations, coarsening.

## Introduction

The Ge / Si(100) heteroepitaxial system follows the Stranski - Krastanov or layer + island growth mode with an equilibrium intermediate layer thickness of 3 monolayers (1 ML = 0.136 nm) [3,5,10]. The islands which form above this intermediate layer are observed to be initially dislocation free [5,10]. In these islands, some of the 4% misfit between Ge and Si can be relieved by inducing inhomogeneous elastic strains in the substrate + layer + island system. This type of growth is termed 'coherent' Stranski - Krastanov (S - K) as opposed to 'normal' S - K growth. Surface energetics dictate that the initial growth proceed in a layer - by - layer fashion for both coherent and normal S - K growth. However, for lattice mismatched systems, there is a critical thickness at which layer - by - layer growth becomes energetically unfavorable due to the increasing strain energy. At this critical thickness, there is a strain driven islanding transition [1,11,12]. For Ge / Si(100), the first islands which form (as long as the nucleation density is small enough so that the strain fields from adjacent islands do not interact) are coherent with the substrate and dislocation free. After continued island growth caused either by particle coarsening or continued deposition from the vapor, these islands may reach a critical radius at which it becomes energetically favorable to relieve a substantial fraction of their elastic energy via the introduction of misfit dislocations. Figure 1 schematically depicts the distinction between coherent and normal S-K growth.

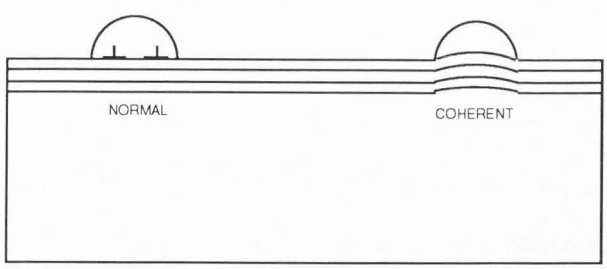


Figure 1. Schematic illustration of coherent Stranski - Krastanov (S -K) growth and normal S - K growth. The coherent island on the right has lowered the interfacial strain by deformation of the island + intermediate layer + substrate system. The normal island on the left has lowered the interfacial strain by introducing misfit dislocations at the substrate - intermediate layer interface.

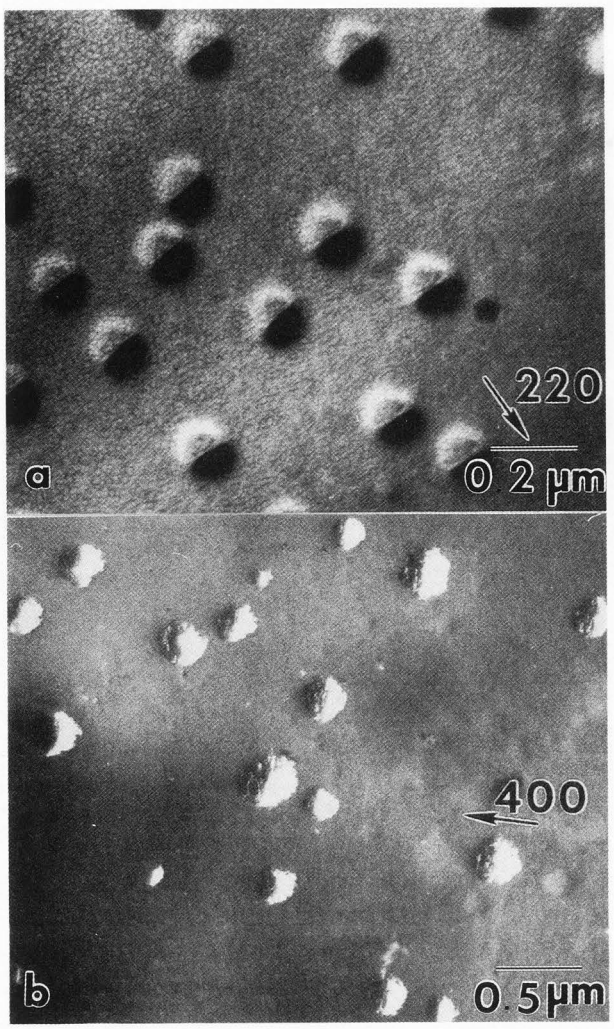


Figure 2. Coherent (2a) and dislocated (2b) islands. The coherent islands display the strong dark - light contrast due to substrate deformation caused by interfacial strain. The dislocated islands show contrast features due to terminated Moire' fringes indicative of highly defective growth.

Jesser and Matthews observed coherent island formation in various metal - metal epitaxial systems [9]. However, these systems follow the Frank - van der Merwe or 3D island growth mode. These observations spawned a number of theoretical investigations into the interfacial and elastic energetics of these systems [2,8]. However, none of these theoretical efforts included substrate deformation as a means of strain energy reduction.

A recent paper [10] showed that these coherent islands significantly impact the coarsening behavior of the Ge islands which form on top of the intermediate layer. The following section reviews the techniques, experimental details and results of the study of island coarsening in the Ge / Si(100) system. This is followed by a discussion where a continuum elasticity calculation confirms that elastic deformation of the substrate is a viable means of strain relief in coherent island growth and equilibrium models of strained epitaxial growth are used to explain some of the experimental results involving coherent and dislocated islands. Finally, a brief summary is given.

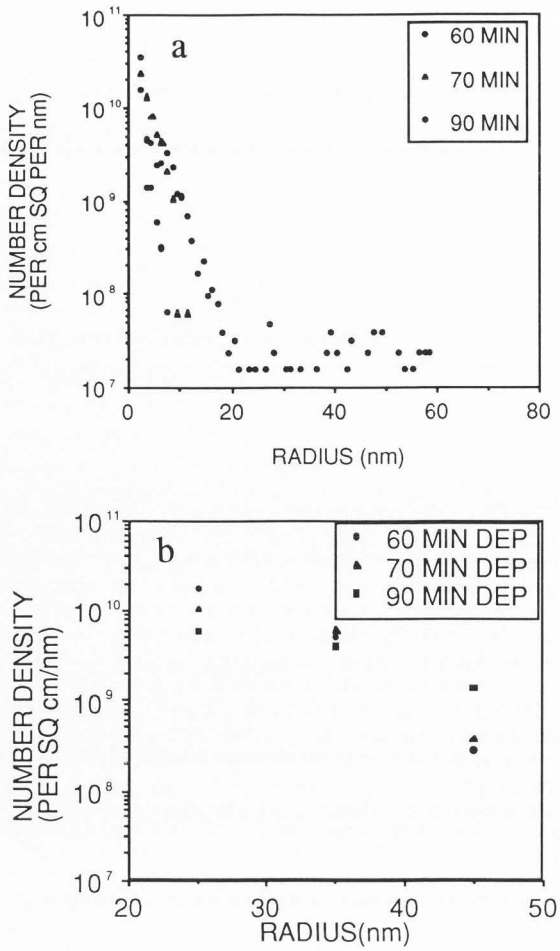


Figure 3. Size distributions generated from secondary electron images obtained in the UHV - STEM. Figure 3a is for the 375°C MBE experiment where Ge was deposited onto Si(100) held at 375°C at 0.1 ML / min. Figure 3b is for MBE deposition as in 3a except the Si was held at 525°C.

## Experimental Techniques

The time evolution of particle size distrubutions is a powerful technique for studying film growth processes such as clustering processes on surfaces [10,15,16]. The majority of these studies were sensitive to particles in the size range of ~0.1 μm and focussed on the late stages of coarsening when the supersaturation was low and Ostwald ripening was the dominant mechanism for film coarsening. In fact, using this technique, it is possible to show that Ostwald ripening concepts are useful for describing coarsening processes in non - conservative systems [15]. A recent paper extended studies of this type to the very early stages of clustering when the supersaturation was still high and was sensitive to islands in the size range of 1.5 nm < radius < 500 nm [10].

In this study, digitally acquired secondary electron (SE) images of Ge / Si (100) films grown *in situ* in an ultrahigh vacuum scanning transmission electron microscope (UHV - STEM) were used to generate particle size distributions using offline image processing software. The UHV - STEM operates at a base pressure of $P < 10^{-10}$ mbar and is equipped with facilities for thermal treatment to 1500°C and molecular

beam deposition. Hence, the 3mm disk samples used in this study were prepared *in situ* by flash removal of the native oxide followed by Ge deposition at a rate of ~0.1 ML / minute. Both solid phase epitaxy (SPE) and molecular beam epitaxy (MBE) were studied. The SPE experiments were performed by depositing on room temperature substrates and the temporal evolution of the film microstructure was monitored as a function of anneal time at elevated temperature. MBE processes were studied by deposition for various times on substrates held at elevated temperatures. Both the SPE and MBE experiments were performed at 375°C and 525°C. Parallel experiments were performed in an offline UHV chamber to further characterize the growth mode and to prepare samples for observation by ex - situ transmission electron microscopy.

A bright field transmission electron micrograph of Ge islands formed after 6 ML of deposition at 1 ML / minute onto Si(100) held at 600°C is shown in figure 2a. These islands display the strong dark / light contrast characteristic of coherently strained islands. Figure 2b shows islands grown in the same chamber, after 8 ML of deposition, under identical conditions. These islands have grown past the point at which it is energetically favorable for misfit dislocations to form and display contrast features due to terminated Moire' fringes which are indicative of highly defective growth. The islands in figure 2b also display dark / white contrast indicating the presence of residual strain.

A series of digitally acquired secondary electron images obtained of islands grown *in situ* in the UHV - STEM were used to produce island size distributions using offline image processing software. Images were acquired and size distributions were produced for each of the MBE and SPE experiments described above. Figure 3 plots the size distributions for the 525°C MBE and 375°C MBE experiments

Figure 4 plots the size distribution for the 375°C SPE experiment to emphasize the different coarsening behavior of the dislocated and coherent islands. By exploiting the wide magnification range available in the STEM, it is possible to obtain excellent statistics for all islands in the size range 1.5nm < radius < 500nm.

The SE and bright field STEM images of islands grown *in situ* in the UHV - STEM at 375°C shown in figure 5 suggest that the larger (radius > 7nm), facetted islands tend to be dislocated while the smaller islands are coherent with the substrate. In what follows, the assumption that the larger facetted islands are dislocated while the smaller islands are coherent with the substrate, and hence dislocation free, will be made.

## Discussion

Since the results displayed in figures 3 and 4 are described in detail in a previous publication [10], the discussion here will center on the effect that the coherent islands have on the coarsening processes evident in the figures. Only the 525°C MBE experiment (figure 3b) displays 'conventional' coarsening behavior. This figure shows that as the experiment progressed, the larger islands grew at the expense of the smaller islands which is typical of Ostwald ripening. Figures 5 suggest that islands with radii smaller than about 7nm were coherent with the substrate while those larger than this had found it energetically favorable to dislocate, in these experiments. Since all of the islands in figure 3b have radii smaller than 7nm, this figure shows that the coherent islands can grow over time and further confirms that Ostwald ripening concepts can be applied to matter non - conservative systems. The major difference in the data represented by figure 3a, 3b and 4 is that in 3a and 4, the coherent and dislocated islands exist simultaneously. Figures 3a and 4 show that the

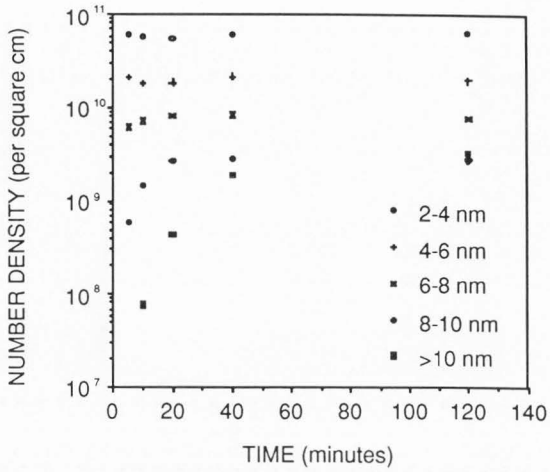


Figure 4. Number density versus time for islands in different size ranges for the 375°C SPE experiment where Ge was deposited on Si(100) held at room temperature at 0.1 ML / min and then annealed at elevated temperature. Note that there is a dramatic difference in the growth rate for the coherent (R < 7nm) and incoherent (R > 7nm) islands.

coarsening behavior is markedly different for coherent and dislocated islands in both MBE and SPE growth.

In figures 3a and 4, the number density of the larger, dislocated islands increases as a function of further deposition (3a) or longer anneal time (4). However, this increase is not at the expense of the smaller, coherent islands. This is especially evident in figure 4 which plots the number density versus time for islands in different size ranges. The number density of islands with radii smaller than about 8nm is essentially constant for the duration of the experiment while the number density of larger islands steadily increases. As discussed in detail previously [10], the source for this increase in number density of the larger islands is the intermediate layer which has grown in excess of its equilibrium thickness, as well as from the vapor in 3a. Prior to a discussion of this difference in the coarsening behavior of dislocated and coherent islands, a continuum elasticity calculation which simulates the elastic deformation of the substrate during coherent island growth (details of which will be published elsewhere) follows.

Using a continuum elasticity model of coherent island growth, it is possible to show that elastic deformation of the substrate is a viable means of energy reduction for these systems. Figure 6 displays the results of a continuum elasticity calculation of the substrate deformation produced by the following boundary conditions:

$$\sigma_{zz} = 0 \qquad 0 < r < \infty\,,\ z = 0 \qquad (1a)$$

$$\sigma_{zr} = 0 \qquad R < r < \infty\,,\ z = 0 \qquad (1b)$$

$$\varepsilon_{rr} = \varepsilon \qquad 0 < r < R\ ,\ z = 0. \qquad (1c)$$

$\sigma_{zz}$, $\sigma_{zr}$ and $\varepsilon_{rr}$ are the normal stress, shear stress and radial strains on the surface z = 0 respectively. R is the island radius. These boundary conditions approximate coherent island growth of a material with a larger unstrained lattice which induces purely radial strain over the island's contact area in the substrate. Figure 6a illustrates the radial displacement, $u_r$, of the substrate surface normalized by the island radius and radial strain as a function of the reduced radial coordinate r / R. Figure 6b illustrates the similarly normalized normal displacement, $u_z$. For an island with a

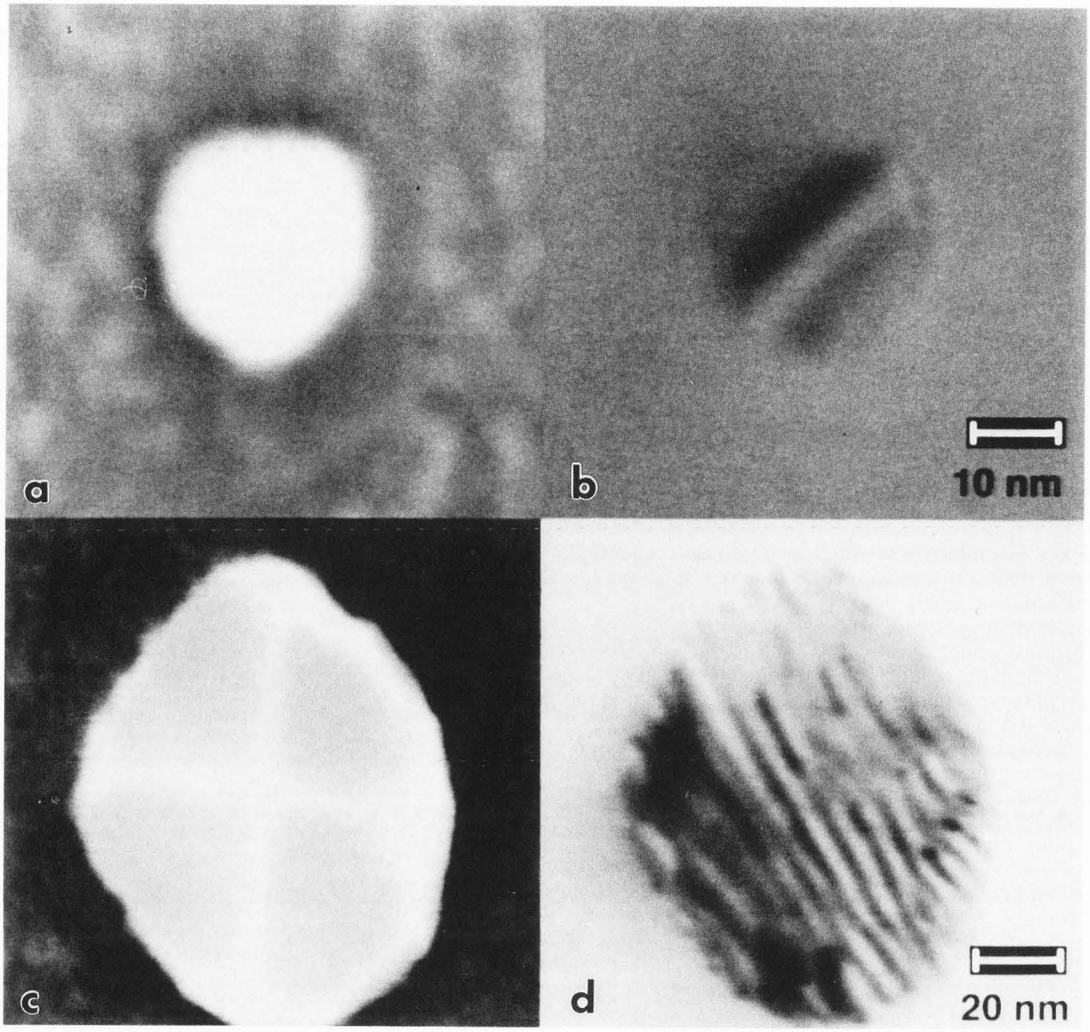


Figure 5: Simultaneously acquired bright field STEM and secondary electron images of coherent (5a and 5b) and dislocated (5c and 5d) Ge islands. In all cases studied in the UHV - STEM, islands with radii smaller than about 7 nm appeared to be coherent while larger islands appeared to be dislocated.

larger lattice constant (positive strain induced in the substrate) than the substrate, the normal displacement is outward. For an island with a smaller lattice constant than the substate, the normal displacement will be inward.

The magnitude of these displacements decays with distance into the substrate so that the maximum displacements are on the surface. From figure 6a, it is obvious that two island's strain fields will not significantly overlap if they have nucleated such that their center to center separation is > 5 R. If the islands nucleate so that this condition is not satisfied then they will probably dislocate at a smaller radius since the overlapping strain fields will increase the total strain energy density. The induced elastic strain (equation 1c) brings the lattice constant of the substrate closer to that of the island and therefore lowers the elastic energy of the system. This result suggests that elastic deformation of the substrate is a viable means of strain relief in coherent island systems.

It was previously suggested [10] that the difference in the growth rate of the dislocated and coherent islands was due to an energy cost associated with increasing the elastic energy of the substrate + intermediate layer + coherent island system. Since dislocation formation releases some fraction of the elastic energy, the energy cost per each additional adatom incorporated into a growing island will be less for a dislocated island than for a coherent island. The energy released during dislocation formation can be estimated with the help of theoretical models developed to describe lattice mismatched epitaxial growth. Equilibrium models developed by Cabrera [2] and Jesser and Kuhlmann - Wilsdorff [8] which relied on van der Merwe's [13,14] expressions for the interfacial energy for lattice mismatched epitaxial growth are useful in this context.

These equilibrium models have recently been supplanted by kinetic models which take into account the energies required for dislocation propagation processes necessary for elastic relaxation of strained epitaxial growth [4,7]. These kinetic models rely on measurements of dislocation propagation velocities to extract activation energies and prefactors for the driving forces necessary for elastic relaxation of strained growth in the Ge - Si alloy system. Although these quantities are becoming well known for certain alloy concentrations for layer growth in heterostructures, they have not been measured

for coherent islands. In fact, the precise dislocation mechanisms for these strained islands are as yet unknown. Therefore, the equilibrium models will be used in what follows.

These equilibrium models were used by Jesser and Kuhlmann - Wilsdorff to predict both the critical thickness for dislocation introduction for layer growth and the critical radius for dislocation introduction in coherent island growth. In their calculations and in what follows, agreement with the experimentally measured critical radii and those calculated must be fortuitous due to the approximations in these simple theories. Their calculations apply to the case of a purely radial strain induced in both the substrate and a hemispherical island of radius R. Their model predicts that the lowest energy state of clusters smaller than a critical radius is for the cluster to be coherent with the substrate. Above this critical radius, it becomes energetically favorable for the cluster to reduce its elastic energy via the introduction of misfit dislocations. This is analogous to the critical thickness at which dislocation free layer growth becomes energetically unfavorable. It has been found for layer growth that kinetics can be exploited to grow the layers far in excess of the critical thickness predicted for these equilibrium models. As will be discussed later, figure 2 shows that this is also possible for coherent island growth.

Jesser and Kuhlmann - Wilsdorff have shown that minimization of the sum of the island and substrate strain energies along with the energy of the bicrystalline interface with respect to the island and substrate strains results in the equilibrium elastic strains for the island and substrate respectively. Their results show that the strain induced in the substrate is proportional to the island strain

$$\frac{\varepsilon_{substrate}}{\varepsilon_{island}} = \frac{1}{4}\frac{\mu_{island}}{\mu_{substrate}}\frac{a'_{island}a_{substrate}}{a'_{substrate}a_{island}} \qquad (2)$$

so that the system can be characterized according to the island strain. The $\varepsilon$'s are the radial strains, the $\mu$'s are the shear moduli and the a's are the in - plane lattice constants. The primes indicate the strained values of a quantity.

Using this result, it is possible to calculate the energy released by the introduction of a single misfit dislocation in a coherent island. For a coherent island, the island strain will be approximately equal to the misfit, f, between the film and substrate

$$f = \frac{a_{island} - a_{substrate}}{a_{substrate}} \qquad (3)$$

For Ge / Si(100) f = 4%. So, a coherent island will be strained so that its in plane lattice constant is approximately

$$a'_{island} = a_{island}(1-f). \qquad (4)$$

For a purely radial strain, the strain relieved by a single misfit dislocation in a coherent island of radius R will be

$$\Delta\varepsilon = \frac{b}{R} \qquad (5)$$

where b is the Burgers vector of the dislocation. The energy change, $\Delta E$, when an island dislocates is defined as

$$\Delta E = E(\text{coherent}) - E(1\ \text{dislocation}) \qquad (6)$$

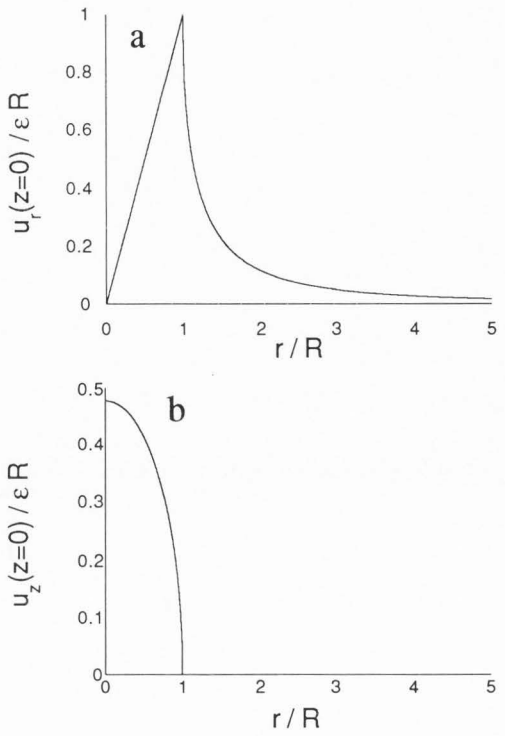


Figure 6. Elastic deformation of the substrate surface produced by the boundary conditions of equations 1. Figure 6a illustrates the radial displacement of the substrate surface as a function of the reduced radial coordinate r / R where R = island radius. $\varepsilon$ is the induced radial strain. The magnitude of the radial displacement decays with distance into the substrate so that the maximum displacements are on the surface. From figure 6a, it is obvious that two island's strain fields will not significantly overlap if they have nucleated such that their center to center separation is > 5 R. Figure 6b illustrates the displacement normal to the surface for the boundary conditions of equation 1. For an island with a larger lattice constant (positive strain induced in the substrate) than the substrate, the 'puckering' is outward. For an island with a smaller lattice constant than the substate, the 'puckering' will be inward.

where E(coherent) and E(dislocation) are the sums of the elastic energies of the island and substrate and the energy of the bicrystalline interface for coherent and dislocated islands respectively.

Figure 7 shows the energy released by misfit dislocation formation, $\Delta E$, for b = 0.543 nm and b = 0.39 nm. It is possible to read the critical radius for misfit dislocation introduction, when $\Delta E$ = 0, directly off of figure 7. The critical radius for b = 0.543 nm is ~7 nm, in excellent agreement with the value suggested by figures 3, 4 and 5. However, Hull and Bean [6] have shown that the most probable dislocations to be formed in a strained epilayer of pure Ge on Si(100) is a 90° dislocation with b = 0.39 nm producing a critical radius of ~5 nm. In the future, further work on these coherent islands may help to characterize the dislocations as well as the exact mechanism for misfit dislocation introduction. It is important to note that the exact 'critical' radius predicted by these equilibrium models assumes that kinetics do not play an important role in the dislocation introduction process. Figure 2 shows that this is not the case.

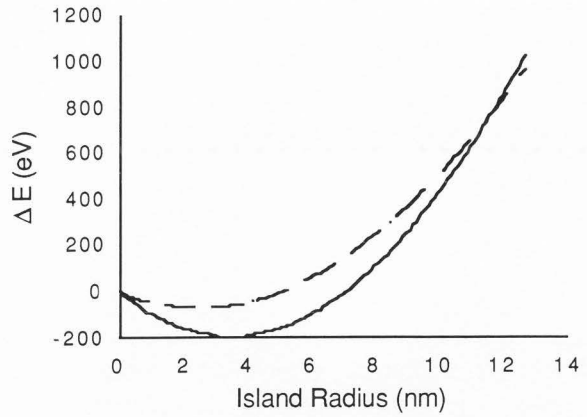


Figure 7: The energy released during the introduction of a single misfit dislocation into a coherent Ge island. The solid line is for an edge dislocation with Burgers vector = 0.543 nm, the Si(100) surface repeat distance. The dashed line is for an edge dislocation with b = 0.39 nm. The dislocations with b = 0.39 nm were shown by Hull and Bean to be the most likely to form in pure Ge layers grown on Si(100).

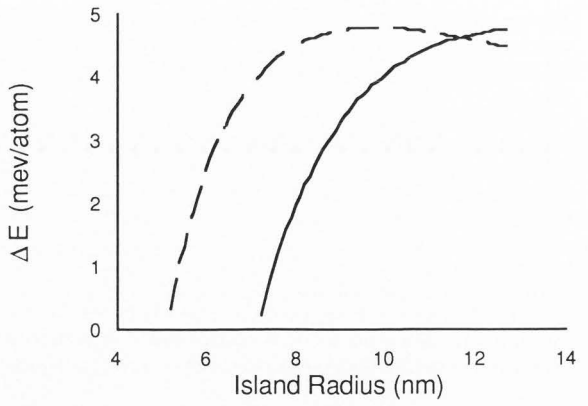


Figure 8. The energy released per atom during misfit dislocation introduction into a coherent Ge island. The dashed line is for an edge dislocation with Burger's vector, b = 0.390 nm. The solid line is for an edge dislocation with b = 0.543 nm.

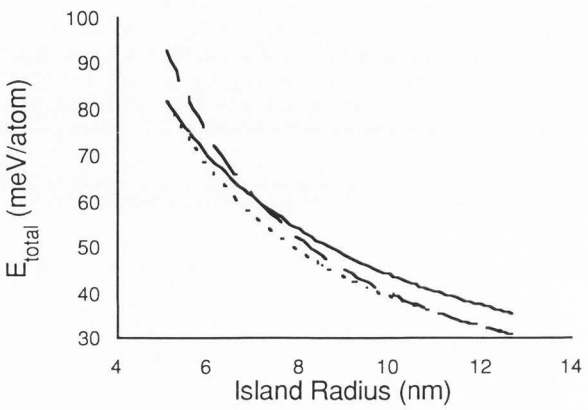


Figure 9. The total elastic energy per atom of Ge islands. The solid line is for a coherent Ge island. The coarse dashed line is for a Ge island with 1 edge dislocation with b = 0.543 nm. The fine dashed line is for a Ge island with 1 edge dislocation with b = 0.39 nm. These curves show that for either type of dislocation, the energy cost per adatom in a growing Ge island is less for a dislocated island than for a coherent island. This result explains the disparity in the growth rates of dislocated and coherent islands.

For islands grown at ten times the deposition rate (figure 2a) of those for which the size distributions of figures 3 and 4 were collected, the radii are 2 orders of magnitude larger. In fact, the equilibrium models do predict that there is a finite energy barrier for misfit dislocation introduction which may partially explain this large discrepancy [8]. This result suggests that there may be a kinetic barrier for dislocation nucleation in coherent island systems which may be exploited to grow these coherent islands to radii far in excess of their critical radius.

The energy released per atom in a coherent island of radius R and the total elastic energy per atom in an island of radius R are plotted versus island radius in figures 8 and 9 respectively. From figures 7, 8 and 9, it is apparent that coherent islands which have grown metastably above the critical radius can relieve a substantial fraction of their elastic energy via misfit dislocation introduction. Approximately 10% of the elastic energy is released during misfit dislocation formation in a 10 nm radius island. This energy reduction can explain the disparity in the growth rates of the coherent and incoherent islands. An adatom incorporated into a coherent island will increase the energy of the strain field of the island. An adatom incorporated into a dislocated island will also increase the elastic energy of the island but by an amount less the energy released during dislocation formation. That is, the energy cost per additional adatom is less for a dislocated island than for a coherent island. If a coherent island were to grow to 10 nm radius, it would cost 44 meV to incorporate an additional adatom. A dislocated island of the same radius would only gain 40 meV by growing by an additional adatom. This energy cost per additional adatom decreases monotonically with island radius since it is the ratio of the energy of incorporation (which rises quadratically with radius) to the island volume.

At this point, it must be stressed that the elastic energy of these coherent islands is only one of the energies which will influence the film growth in these systems. The energy of formation of the clusters, surface energies etc. must also be considered. The concepts presented here serve only to illustrate that elastic energy should also be considered important in any discussion of the energies involved in clustering phenomena. In fact, this elastic energy is central in the explanation of the difference in the growth rates of dislocated and coherent islands apparent in figures 3 & 4. In combination with the other energies important in clustering processes, the energy released during misfit dislocation introduction is apparently enough to swing the energy balance so that island growth becomes easier for dislocated islands.

## Conclusions

Coherent islands dramatically affect cluster coarsening in the Ge - Si (100) heteroepitaxial system. In all cases studied, coherent islands grow more slowly than those islands which have found it energetically favorable to reduce their elastic energy via misfit dislocations. A continuum elasticity

calculation which simulates the substrate deformation in coherent island growth suggests that elastic deformation of the substrate is a viable means of energy reduction in coherent islands. Within the context of a simple equilibrium model of strained epitaxial growth a number of quantities important in the microstructural evolution of islanded systems can be estimated. This model shows that the introduction of a single misfit dislocation can lower the elastic energy of a Ge cluster by a significant amount, about 10% for a island with a 10 nm radius. The approximations in this simple model preclude any absolute quantitative predictions of critical radii for coherent islands or energy differences between coherent and dislocated islands. However, it is adequate to help explain some of the unconventional coarsening behavior observed in coherently islanded systems. The reduction in energy during the coherent to incoherent traisition is apparently enough to swing the energy balance in the favor of accelerated growth. The energy cost per additional adatom in a growing dislocated island is less than that for a growing coherent island. These concepts explain the difference in the growth rates of coherent islands and dislocated islands.

## Acknowledgements

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## Discussion With Reviewers

D.H. Vanderbilt : The normal displacement, $u_z$, of the surface shown in Fig. 6(b) appears to vanish for $r > R$. It is not clear to me that this follows from the choice of boundary conditions in Eqs. (1 a - c). Is there a simple explanation for it ?

Author: No, there is no simple explanation for the vanishing of $u_z$ for $r > R$. It is a consequence of the continuum elasticity calculation with the boundary conditions specified in Eqs. (1 a - c), results of which will be published elsewhere.

D.H. Vanderbilt : At the end of section 3, there is a discussion of the "energy cost per additional adatom" for coherent and dislocated islands. Isn't this quantity really just the chemical potential ? And can't we then interpret the results as indicating that the chemical potential of a 10 nm dislocated island is 4 meV lower than a corresponding coherent island ? If so, this may be a useful way of thinking about the driving force for the ripening kinetics which favor large dislocated islands.

Author: Yes, the change in the chemical potential of a cluster with N atoms is $\Delta\mu = dE/dN$. This change in chemical potential is related to the equilibrium solubility of the cluster through the Gibbs - Thomson equation. The consequences are that a coherent island will 'dissolve' into a dislocated island of the same radius.

R. Hull : No details are given of the efficiency of the substrate cleaning, which could of course play a major role in nucleation and coarsening. Is the 2x1 reconstruction seen ? Also, the substrate used for the SPE is likely to have considerably more contamination than the MBE substrates as the sticking coefficients for O etc. will be much greater at room temperature.

Author: This is an important issue. For the experiments performed in the 'offline' chamber, the 2x1 reconstruction was observed in every instance. Furthermore, all contaminants (O, C, etc.) were below the detectivity limits of Auger electron spectroscopy for both MBE and SPE

experiments. Samples prepared in the UHV - STEM were prepared in an identical matter. However, the UHV - STEM samples were 3mm diameter discs versus 20mm x 3mm strips for the offline experiments. The strip samples are far easier to clean. Further details of the substrate cleaning (and other experimental details are given in reference 10.

R. Hull : I think the neglect of kinetic effects for dislocation introduction is probably less of a problem than the authors believe. Certainly dislocation propagation velocities for coherent Ge/Si will be enormous (compared to the island dimensions) even at 375°C. The nucleation kinetics remain more of a question, but recent calculations of homogeneous nucleation barriers (e.g. those by Hirth) in this system, which use lower dislocation core energies based upon predictions of core reconstructions, would suggest that nucleation in pure Ge/Si is quite accessible for the temperatures considered here.
Author: I am not aware of this work, thank you for pointing it out.

R. Hull : The dislocation Burgers vector magnitude in eqn. 5. should presumably be only that component of the Burgers vector lying withih the interfacial plane (which is 100% for a/2<110> edge dislocations but only 50% for a/2<101> glide dislocations. The concept of a 0.543 nm burgers vector is interesting: this would correspond to b = a<010> which to my knowledge has never been observed in bcc semiconductors (but, possibly at these enormous stresses.....). In the absence of any supporting evidence, such a dislocation must be deemed unlikely, but "simple" TEM diffraction contrast should determine its existence, and identification of such a defect would be very important.
Author: As regards the component of the Burgers vector in the interfacial plane, you're absolutely right. The identification of the type of defect as well as the nucleation process and kinetics are topics for further research in our laboratory.

R. Hull : I do not understand why the data of Figure 4 necessarily implies that small islands are not growing. Assuming the increase in matter in the island population is supplied from the Stranski-Krastanov layer (is it known that such a layer exists for SPE?), surely all that is known is that dN/dt is constant for the small islands. This could arise from an equilibrium between a loss to the population from small islands growing bigger and an increase in the population due to small clusters forming from atoms in the S - K layer.
Author: You're correct that all that can be inferred from our data of Figure 4 is that dN/dt is a constant for the smaller, coherent islands. Regardless of whether the 'increased supersaturation' due to the decay of the metastable S - K layer feeds the larger islands or nucleates additional small islands, the data of Figure 4 does not display 'conventional' coarsening behavior. Although it is not known that a S - K layer exists for SPE, recent work by Dr. Eaglesham on the critical epitaxial thickness as a function of temperature would imply the existance of such a layer.