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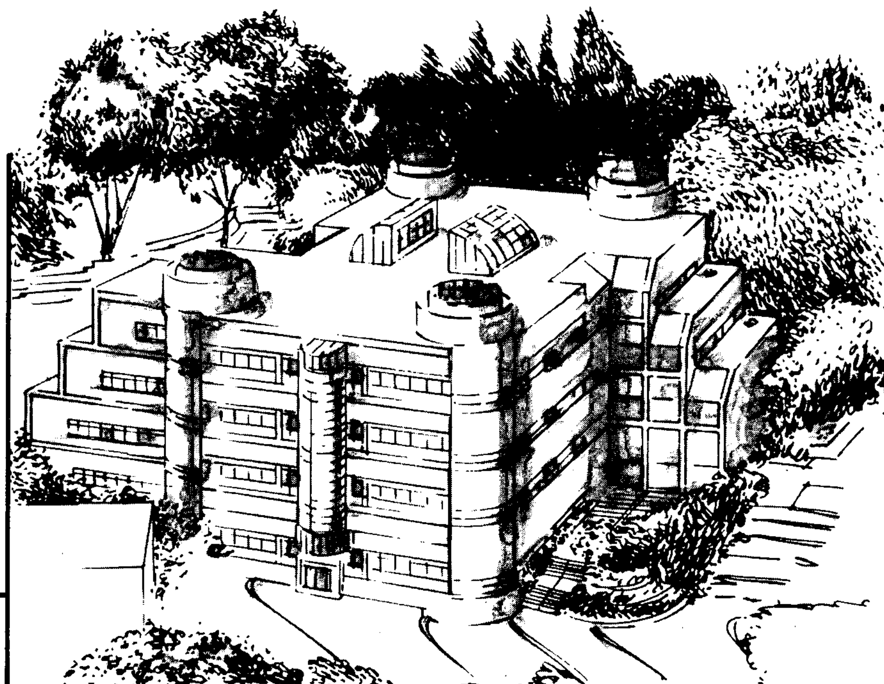

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INVESTIGATION OF MISFIT DISLOCATION CONFIGURATIONS IN MBE-GROWN InGaAs LAYERS ON MISALIGNED GaAs (001) SUBSTRATES

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ABSTRACT

The configurations of misfit dislocations in $In_{0.2}Ga_{0.8}As$/GaAs(001) heterostructures grown on slightly misoriented substrates was investigated by transmission electron microscopy (TEM). Layers 6 nm, 20 nm and 40 nm thick were grown by MBE. The substrate was tilted in [110], $[1\bar{1}0]$, $[1\bar{2}0]$, [210] and [010] directions at angles between 0° and 10°. Only in the 40 nm thick layers networks of 60° and 90° dislocations were formed. Misfit dislocations were found at the interface in <110> directions. In the substrate tilting range between 0° and 4° the changes in dislocation density can be explained by the different character of α and β dislocations. For a substrate tilting above 6° the different dislocation sets show an increased anisotropy. The misfit dislocations at the interface were decorated by In atoms. The influence of three-dimensional crystal growth on increasing surface roughness is discussed .

I. Introduction

In recent years the growth of InGaAs epitaxial thin films has become the topic of numerous investigations due to their application in optoelectronics. However, only in a limited number of cases can a perfect match between the lattices of the film and the substrate be realized. One case is $In_{0.53}Ga_{0.47}As$ layers grown on InP(001) substrate. Since a high crystalline film perfection is needed for good device performance the goal of many investigations is to reduce the density of misfit dislocations or to pin them near or at the interface, where they have almost no influence on the optoelectronical properties of the layers. InGaAs films on GaAs (001) substrates seem to be a useful system for studying the initial stages of the crystal growth of strained layers and the generation of misfit dislocations.

Misfit dislocations found in the InGaAs heterosystem are of the Lomer type (90° edge, Burgers vector **b**=1/2a{110} in the interface)[1], as well as the 60° type[3] (**b**= 1/2a{101} inclined to the interface plane). All these misfit dislocations lie at the interface in the <110> directions. It was recently reported that, in a InGaAs/GaAs(001) structure with high In concentration (x>0.4), edge-type misfit dislocations inclined at 45° to the film surface can be generated by glide in {110} slip planes.[2]

In AIII-BV crystals, parallel opposite-sign 60° dislocations are not chemically equivalent.[4] For the so-called α-dislocations the extra half-plane terminates with a row of group-III atoms, whereas the core of β-dislocations is characterized by a row of group-V atoms. The α- and β-dislocations also show different behavior. A 60° dislocation can glide on four different {111} planes. For InGaAs/GaAs(001) heterostructures there are four different 60° dislocations, with edge components of their Burgers vector directed in $[11\bar{2}]$, [112] or $[1\bar{1}2]$, $[1\bar{1}\bar{2}]$. The extra half plane extends into the substrate. For half of these 4 types of 60° dislocation (e.g $[11\bar{2}]$, [112]) the extra half planes are terminated by group III elements (α-dislocations), for the other they are terminated by group V elements (β -dislocations). This "chemical" difference at the cores of dislocations may be responsible for an anisotropy of optical and electrical, as well as structural properties.[3] Furthermore, α-dislocations have a higher mobility in comparison to the β-type dislocations.[5,6]

The goal of this work was to study the initial stage of the MBE growth of InGaAs on a (001) GaAs substrate as a function of substrate misalignment. Since different dislocations, as well as surface steps due to the misorientation running in the<110> directions can have a

different chemical nature, as mentioned above, we also investigated the influence of the direction of tilting. In this first study we have chosen an In concentration of x=0.2, for which two-dimensional crystal growth was expected.

II. Experimental

The $In_{0.2}Ga_{0.8}As$ layers were grown by molecular-beam epitaxy at a substrate temperature of 520°C. This composition corresponds to a lattice mismatch between the layer and the substrate of 1.4%. The [001] oriented GaAs substrate was misaligned in the following five directions: [010], [110], $[1\bar{1}0]$, [120], and $[\bar{2}10]$. The tilting angles in each of these directions were 2°, 6°, 8° and 10°. A 0.5 μm-thick GaAs buffer layer was grown to obtain a clean interfaces. Layers were grown with the following thicknesses: 6 nm, 20 nm, and 40 nm. To guarantee identical growth conditions, the growth of all samples of the same thickness was carried in the same growth run.

For TEM investigations, plan-view samples as well as cross-section samples were prepared by standard techniques. Since the interfaces are characterized by different inclinations in different crystallographic directions cross-section samples were prepared allowing observations in the [110] as well as in $[1\bar{1}0]$ directions. The TEM samples were examined in a JEM 200CX and in the Atomic Resolution Microscope at the Lawrence Berkeley Laboratory.

III. Results and Discussion

As expected no misfit dislocations were found in layers with thicknesses of 6 nm and 20 nm. However, the samples did show clear differences in morphology, depending on the substrate misalignement. Island growth occurred on the (001) GaAs surfaces, which were misaligned in [110] and [120] directions at angles >6° (Fig. 1). Island growth was not observed for misaligned substrates tilted in the $[1\bar{1}0]$ and $[\bar{2}10]$ directions. Fig. 1a shows a bright-field TEM image of a 6 nm-thick film grown on a substrate with an inclination of 6° in the [110] direction. Islands of drop-like shape with an average size of about 0.5 μm, are visible due to weak strain contrast. The diffraction contrast has a zero contrast line perpendicular to the diffraction vector **g**, showing the radial strain field. In cross-sections no noticeable surface undulation corresponding to islanding was observed. Therefore island growth only persisted during growth of the first few monolayers. Analysis of image contrast on the plane-view and cross-section samples shows that the observed strain fields could be a result of slight variations of the In concentration. It appears that, at the early stages of film growth, the first deposited material tries to adapt its lattice parameter to that of the substrate by rejection of In. As a result the surface regions becomes enriched in In. Contrast variations visible as fine lines were observed inside the islands (Fig.1b). These lines correspond to regions with a slightly higher electron scattering factor, which was interpreted as slight variations in In concentration with a periodicity of about 20 nm. Suggesting that the first deposited material takes a dendritic shape, such striations could be a result of an elastic energy minimization due to a nonlinear dependence of the lattice parameter on the composition variation.

Island growth was not observed when the substrate was tilted in the $[1\bar{1}0]$ and $[\bar{2}10]$ directions. This is clear evidence that the chemistry of surface steps plays an important role in the layer growth mechanism. Two-dimensional growth as well as island growth of $In_xGa_{1-x}As$ on GaAs (001), can occur, and the kind of growth is determined by the In concentration x. Using transmission electron microscopy (TEM), it was found that a transition from two-dimensional to island growth appear at concentrations x>0.4.[1,2] From the present results it is clear that a type of initial stage island growth takes place at lower In concentration (x=0.2) for some misoriented substrates.

In all 40 nm-thick InGaAs layers, misfit dislocations running in the [110] and $[1\bar{1}0]$ directions and forming rectangular networks were observed (Fig. 2a). Most of the dislocations were of the 60° type. However, some were 90° edge dislocations (**b** = 1/2a[110]). Investigations by high-resolution electron microscopy have shown that the these dislocations sometimes have a complicated core structure, due to decoration by In atoms.[7] For a non-tilted substrate (exact [001] surface normal), the two dislocation sets are exactly

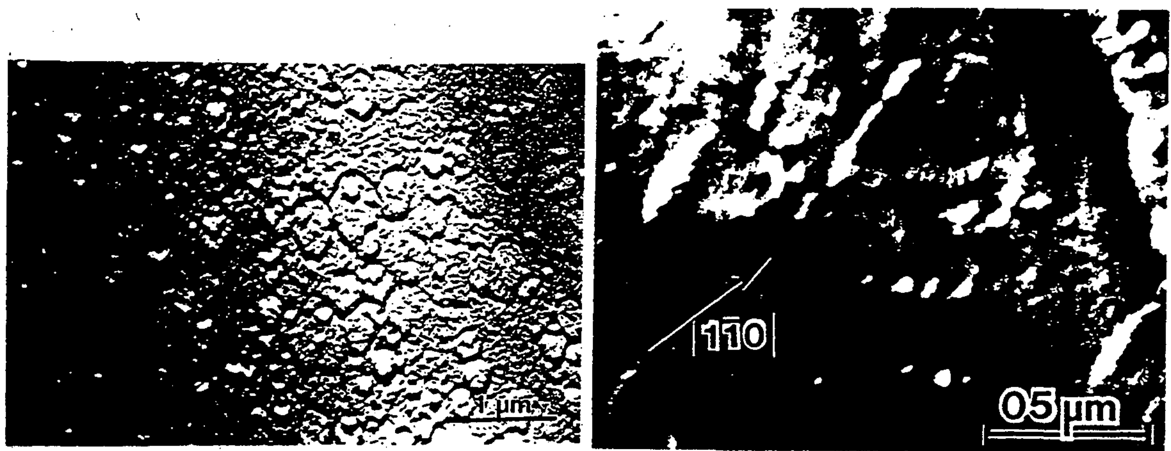

Fig.1. a) TEM bright-field micrograph of a 6 nm thick $In_{0.2}Ga_{0.8}As$ film grown on 6° misaligned substrate in [110] direction. The drop-shaped features correspond to 3-dimensional growth of islands.
b) Dark-field image of an individual island showing an inner structure. The periodisity of observed lines is about 20 nm.

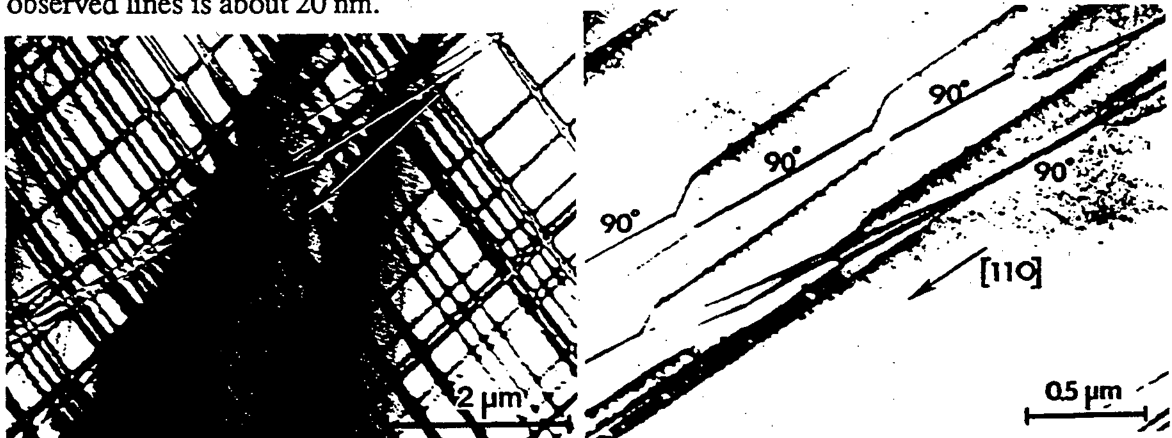

Fig.2. a)Typical dislocation network in 40nm thick $In_{0.2}Ga_{0.8}As$ layers. The defects have mainly the character of 60° dislocation running in [110] and [110] directions. Substrate tilting: 10° in [120] direction. b) TEM bright-field image of a dislocation network resulting from the interaction of 60° dislocations at the interface. 60°- and 90° Lomer type dislocations appear as A and B lines correspondingly.

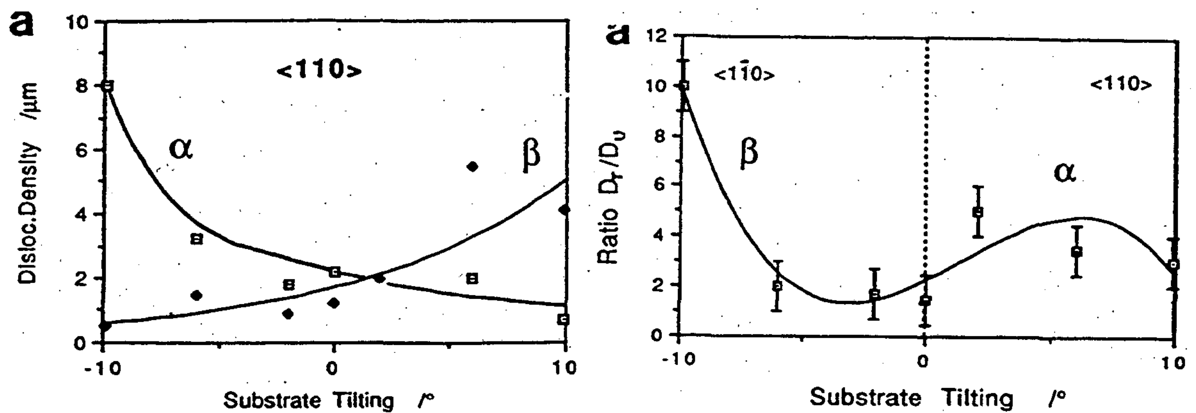

Fig.3. a) Dislocation line densities measured in 40nm thick $In_{0.2}Ga_{0.8}As$ layers grown on misaligned GaAs substrate. Misalignement in <110> directions b) Ratio D_α/D_β of the dislocation densities lying perpendicular (D_α) and parallel (D_β).to the tilting direction.

dislocations sometimes have a complicated core structure, due to decoration by In atoms.[7] For a non-tilted substrate (exact [001] surface normal), the two dislocation sets are exactly perpendicular to each other. In the case of tilted substrates, each dislocation set is split into two new sets of dislocation lines, still lying close to the interface, but slightly inclined to each other (Fig. 2b). The angle between them (about 11°) is related to the substrate tilting and corresponds to the intersection of the {111} slip planes with the crystal surface.

Since the sets of dislocations run at the interface in approximately the [110] and $[1\bar{1}0]$ directions, it is useful to determine dislocation line densities D for both directions separately. We define D_α, D_β as a number of dislocations on slip planes [111], $[11\bar{1}]$ and $[\bar{1}11]$, $[1\bar{1}1]$ per unit length, respectively. Fig. 3a shows the dependence of dislocation line density on tilting angle. The two curves, marked by α and β, correspond to the two perpendicular dislocation sets. For a substrate tilted at -10° in [110], the dislocation line densities differ by about one order of magnitude. At an exact [001] substrate orientation, the dislocation densities are lower, but still differ by a factor of two. The same behavior of D was found for <120> misaligned substrates. For a [010] inclination the dislocation densities were not so strongly affected by substrate tilting. In all cases the dislocation densities in the two sets marked by α and β were different. This different density of dislocations running in perpendicular directions can be more clearly illustrated by their ratio $R = D_\alpha/D_\beta$ (Fig. 3b).

For the substrate misalignment up to about 4°, R increases from left to right (Fig. 3b). Substrate inclination in the [110] or $[1\bar{1}0]$ directions gives rise to preferential formation of α- or β-type dislocations. The density of dislocations observed was rather low corresponding to release of only 10% of the misfit stress. The dislocations were very long straight lines laying at the interface. This suggests that the dislocation density was controlled by nucleation rather than by dislocation mobility. Thus the nucleation appears to depend on substrate orientation. Such an effect could be due to differences in the density, polarity and bunching of steps on the growing surface and surface reconstruction.
For larger substrate inclinations (>4°), the three dimensional growth has a strong influence on dislocation structure asymmetry.

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