Misfit dislocation formation in lattice-mismatched III–V heterostructures grown by metal–organic vapour phase epitaxy

J te Nijenhuis, P J van der Wel, E R H van Eck and L J Giling

Department of Experimental Solid State Physics III, RIM, Faculty of Science, University of Nijmegen, Toernooiveld, 6525 ED Nijmegen, The Netherlands

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Abstract. Misfit dislocation formation in lattice-mismatched III–V heterostructures both under tensile and under compressive stress has been studied. Layers of GaAs under tensile stress have been grown by metal–organic vapour phase epitaxy on In$_x$Ga$_{1-x}$As substrates with indium concentrations between 0.1 and 1.1%. Compressively strained In$_x$Ga$_{1-x}$As layers with indium concentrations between 0.5 and 2.5% have been grown on GaAs substrates. For the layers under tensile stress an asymmetrical dislocation pattern has been observed, whereas the compressively strained layers show a symmetrical dislocation pattern. A model describing the relaxation process by the formation of dissociated hexagonally shaped half-loop dislocations is proposed. A difference in the mobilities of the two possible misfit dislocation types is found to be the origin of asymmetrical strain relief at low growth temperatures. In layers under tensile stress the cross slipping of screw dislocations is counteracted by the shear stress, leading to relaxation in only one direction. In layers under compressive stress the nucleated misfit dislocations can undergo cross slipping, resulting in a cross hatched pattern at the surface. At higher growth temperatures the dislocation patterns become more symmetrical due to the higher dislocation mobilities. Growth hillocks are formed on the surfaces of the layers grown under tensile stress, due to local accumulation of dislocations. This hillock growth is prevented at higher growth temperatures by the higher mobility of the dislocations. It is also shown that the (mis)orientation of the substrate is revealed by non-parallel groups of dislocation lines observed at the surface of a relaxed epilayer.

1. Introduction

A typical defect in strained layer epitaxy is the misfit dislocation, which is formed when the thickness of the epitaxial layer exceeds a certain, strain-dependent, critical layer thickness [1]. Several theoretical models for the calculation of this thickness, whereby the transition takes place from pseudomorphic (misfit-dislocation-free) to incoherent growth, have been proposed in the literature [1–9]. In the thermodynamic model of Matthews and Blakeslee [1], known as the mechanical equilibrium model, the critical layer thickness is calculated by balancing the elastic strain energy with the energy of a square grid of misfit dislocations. The strain is released by the movement of pre-existing dislocations along the interface between two layers with different lattice constants. However, in semiconductor films the density of grown-in dislocations is not high enough to relieve the strain completely.

Therefore it is assumed [2–9] that new dislocations have to nucleate at the surface and propagate as half-loop dislocations towards the interface. In order to choose from these models the most realistic one, it is necessary to have a good knowledge of the relaxation processes.

The stress in the epilayer can be either tensile or compressive. In layers under tensile stress the lattice constant of the relaxed epitaxial layer $a_0$ is smaller than that of the substrate $a_s$, whereas in compressively stressed layers the lattice constant of the epilayer is larger than the lattice constant of the substrate. It has been shown [10] that, in the zincblende structure, the surface nucleated misfit dislocations, lying in each of the two (011) directions in the (100) plane, have different Peierls barriers, leading to asymmetrical relaxation patterns [11–16]. In an earlier paper [17] the opto-electronic behaviour of the misfit dislocations in layers under tensile and compressive stress in the GaAs/In$_x$Ga$_{1-x}$As system was examined by spatially resolved photoluminescence.
micrographs are shown of partially relaxed GASSs. In figure 1, interference-contrast microscopy images of layers grown under tensile stress show an asymmetrical:

3.1. The morphology of layers under tensile stress

3. Results

DICM) examined by differential interference-contrast microscopy of the morphologies of the epilayers was vertical direction is the [011] for all the micrographs.

The growth temperature was in the range of 837°C to 910°C. The effects indicate the presence of micromechanical textures. The morphologies of GASSs (110) and (100) epilayers with morphologies of GASSs (100) epilayers with morphologies of surface

Figure 1. Interference contrast micrographs of surface

2. Experimental details
leads to asymmetric strain relief. In-plane shear stress, the absence of the possibility for cross slip due to a lower shear stress. It follows from this model that in layers under the condition of the dislocations and the direction of the dislocations, dislocations are considered as functions of the dislocation mobility and the cross slip possibility of the dislocation network. In this model, the formation and propagation of dislocations are governed by the activation of the edge layer surfaces.

The evaluation of the dislocations are no longer discernible, which hinders the detectability of the individual misfit dislocations. After partially relaxed layers, the low dislocation density is essential to reveal the individual misfit dislocations after low-angle layers as studied in the present paper. The mismatch in the lattice-mismatched single heterostructures in the GaAs/InGaAs system is studied. The mismatch in the lattice-mismatched single heterostructures, the mismatch in the GaAs/InP and InGaAs/GaInP heterostructures. In these experiments, two different dislocation core types were observed. In the present paper, the mismatch dislocation formation led to the asymmetric strain relief.
4. Discussion

The figures under tensile stress (Fig. 2), which is applied to the fractures of the tensile fracture plane, illustrate the formation of the fractures, and in the diagram [110] it can be seen that the fracture occurs in the direction of the crystallographic axes of the sample. The figures show that under the condition of the samples, the fracture occurs in the direction of the crystallographic axes of the sample, and it can be seen that the fracture occurs in the direction of the crystallographic axes of the sample.

4.1. Introduction and application of misorientations

As shown in Figure 2, the interference contrast microscopy of the fracture surface shows clearly that the fracture surface is formed at the crystallographic axes of the sample. The fracture occurs in the direction of the crystallographic axes of the sample, and it can be seen that the fracture occurs in the direction of the crystallographic axes of the sample.

4.2. The morphology of layers under compressive stress

(4.2.1) A cross-hatched pattern with different thicknesses is observed. In Figure 2(c), the cross-hatched pattern with different thicknesses is observed. In Figure 2(c), the cross-hatched pattern with different thicknesses is observed. The lines of different thicknesses are shown clearly in Figure 2(c).

The effect of the growth process on the fracture surface shows clearly in Figure 2. The effect of the growth process on the fracture surface shows clearly in Figure 2. The effect of the growth process on the fracture surface shows clearly in Figure 2. The effect of the growth process on the fracture surface shows clearly in Figure 2.
other by dislocation climbing. The atoms in the dislocation core in III(A)–V(B) compounds can be of element A or B. The four types are indicated by the terms A(g) and B(g) for the glide set dislocations and A(s) and B(s) for the shuffle set dislocations. Screw dislocations have no net A or B characters; they are made of partials with A and B characters as stated below.

It is well known that a 60° dislocation is dissociated into a 30° and a 90° Shockley partial, containing a stacking fault area in between them. Since the movement of extended dislocations in the shuffle set is difficult, due to the complex core of the partials [28], it is generally believed that the mobile 60° dislocations in GaAs are of the glide set. This assumption is in agreement with the calculated [29] and the observed [17] electric behaviour of the dislocation cores. Hence the discussion is limited to the glide set dislocations.

When the density of the threading dislocations is too low to release the strain in the epitaxial layers, the misfit dislocations nucleate at the surface and propagate as half-loop dislocations towards the interface [2]. As will be clarified in more detail in the next section, the shape of the half loops is considered to be hexagonal. The configurations of dissociated hexagonal half-loop dislocations [30] are shown in a projection on the {111} planes in figure 4. A hexagonal half loop, that relieves the strain, has a Burgers vector inclined on the (100) surface and consists of two 60° dislocations and one screw dislocation, all dissociated into Shockley partials. The dissociated screw dislocation differs from the 60° (either glide or shuffle) dislocation since it consists of two 30° Shockley partials, instead of one 30° and one 90° partial. The individual parts of the whole dislocation half loop lie in the Peierls valleys along the (011) directions. Insofar as the 60° glide dislocations are concerned, it can be seen that the two 60° parts of one dislocation half loop are of the same dislocation type, either As(g) or Ga(g), only the orientations of the Burgers vectors of the leading and the trailing Shockley partials with respect to the orientation of the line are reversed. Four of eight possible half loops are shown, two in layers under tensile stress and two in layers under compressive stress. In each of the two cases the horizontal (bottom) lines are in the mutually perpendicular [011] and [011] directions. An overview of the relations among stress type, orientation of the dislocation (bottom) line at the interface and dislocation type is given in table 1. The four non-shown half loops lying in equivalent (111) planes have the same characteristics insofar as the leading/trailing partials are concerned.

For the origin of the observed asymmetry in the formation of misfit dislocations, there are three possibilities:
Misfit dislocation formation in III–V heterostructures

Projections of hexagonal dissociated dislocation half loops on the (111) planes, propagating from the surface to the interface, for layers under tensile stress (a) and (b) and under compressive stress (c) and (d). The dislocation types are (a) As(g), (b) Ga(g), (c) As(g) and (d) Ga(g). The shaded areas between the partials contain a stacking fault. The width of these areas gives an indication of the dissociation widths of the partials. The line sense of the dislocation loops is counter-clockwise.

Table 1. Overview of the relation between line direction parallel to the (100) surface and the dislocation core as a function of the stress type.

<table>
<thead>
<tr>
<th>Stress type</th>
<th>Line direction</th>
<th>Tensile</th>
<th>Compressive</th>
</tr>
</thead>
<tbody>
<tr>
<td>(011)</td>
<td>Ga(g)</td>
<td>As(g)</td>
<td></td>
</tr>
<tr>
<td>(011)</td>
<td>As(g)</td>
<td>Ga(g)</td>
<td></td>
</tr>
</tbody>
</table>

An asymmetry in the nucleation process itself; an asymmetry in glide behaviour and an asymmetry in the possibility of producing cross slip. Experimentally, no evidence has been found for asymmetry caused by a difference in the energy barrier for homogeneous nucleation of misfit dislocations at the surface for both glide set types [10]. Moreover, asymmetrical dislocation patterns have also been observed in epitaxial layers with a high density of threading dislocations, for which no nucleation of misfit dislocations at the surface is required [13]. It has been stated by Fox and Jesser [10] that the asymmetrical relaxation, as seen in figure 1(a), is caused by the difference in the Peierls barrier for dislocation motion of the glide sets of type As and Ga. In order to check whether this can give an explanation for the observed asymmetry, a closer analysis is given of the influence of the difference in the mobilities between the As(g) and Ga(g) dislocation types on the relaxation process.

The As(g) perfect dislocation in GaAs has a much higher mobility than does the Ga(g) perfect dislocation [31]. For dislocations in In₀.₅Ga₁₋₅As no exact data on mobility are known. It has been reported by Burle-Durbec et al [30] that the mobility of the As(g) dislocation is reduced with indium concentration in GaAs, whereas the mobility of the Ga(g) dislocation remains constant. However, the As(g) dislocation still has a higher mobility in In₀.₅Ga₁₋₅As than does the Ga(g) dislocation. Hence the relaxation will start in the [011] direction in layers under tensile stress and along [011] in layers under compressive stress. In figure 1(a) it can be seen that, in the layers under tensile stress, the relaxation has indeed started in the [011] direction. The dislocations are more prominently present in this direction, which is in agreement with the model. However, at the higher growth temperatures the dislocations in this tensile system are observed in both the (110) directions (figure 2(c)). Apparently, the mobility of the Ga(g) dislocation has become sufficiently high for it to move through the epitaxial layer to release the strain. In contrast to this, it is experimentally observed that, in layers under compressive stress, both the As(g) dislocation in the [011] direction and the Ga(g) dislocation in the perpendicular [011] direction are already present, even at the lower growth temperatures (figure 3). So it is concluded that the above description of the relaxation process, based on differences in glide motion, certainly is not complete. It still has to be explained why the formation and propagation of the Ga(g) dislocations are easier in the layers under compressive stress than they are in those under tensile stress. In order to do this the motion of dissociated dislocations, especially their ability to produce cross slipping, has to be considered.

4.2. Relaxation by cross slipping of misfit dislocations

Maree et al [3] have proposed a model for relaxation in III–V compounds for layers both under tensile and under compressive stress in terms of semicircular loops.
This model is summarized below; thereafter some remarks regarding this model are made. It is assumed [3] that the relaxation is initiated by the nucleation of Shockley partial dislocations, which expand to form a semicircular dissociated dislocation loop. In layers under tensile stress the leading partial, namely that at the bottom of the loop, is the 90° As(g). In compressively strained layers the 30° As(g) is the first partial to move. This is shown in figure 5 and summarized in the first column of table 2. The width of the stacking fault area depends on the stress direction. Since the shear stress, exerted on the 90° partial, is twice that on the 30° partial, the dissociation distance is large when the 90° Shockley partial is the leading one. This is the case in the layers under tensile stress for which the dislocation loop is widely extended, because here the 90° partial is the first to nucleate. The dissociation distance is small when the 90° Shockley partial is the trailing one, as is the case in layers under compressive stress. From this Marée et al [3] concluded that cross slipping is easier in layers under compressive stress, for which the 60° dislocations are (almost) perfect, whereas the large distance between the loops in the layers under tensile stress explains why cross slipping does not occur.

On the above-described model of Marée et al two serious remarks can be made. Firstly, since the dislocation is not a straight line and the orientation of the Burgers vector is constant along the loop, the character of the dislocation depends on the position on the line. Consequently, the dissociation distance of the Shockley partials changes too and so the proposed cross slipping mechanism should be dependent on the position along the loop. In the second place, only screw (0°) dislocations, for which the Burgers vector is parallel to the dislocation line, are known to possess the cross slipping possibility [26]. These objections make the described mechanism of a complete cross slip along the whole dislocation line less likely. Therefore here a model is presented, which deals with the changing dislocation character along the half loop and in which only the screw dislocation part is supposed to be able to undergo cross slipping. However, first the 60° dislocations are considered, since these dislocations are able to release the misfit strain.

Observations of misfit dislocations, nucleating at the surface and propagating towards the interface, using transmission electron microscopy (TEM) in the diamond cubic and in the zincblende structure with mismatches of about 10^{-2} [3,8,32-34] have shown that these dislocations are V-shaped. No evidence for semicircular dislocations as assumed in theory [2,4], has been found experimentally. Based on these TEM observations, Fukuda et al [34] proposed a model which assumes that misfit dislocations nucleate and propagate as perfect V-shaped dislocations instead of semicircular loops. These dislocations can be divided into two parts, as is shown in figure 6. One is a screw dislocation, whereas the other is a 60° dislocation. When this V-shaped dislocation has reached the interface between the epilayer and the substrate, it will expand along
Figure 7. Misfit forces $F$ and frictional forces $R$ acting on leading (l) and trailing (t) Shockley partials of screw dislocations in layers under (a) tensile stress and (b) compressive stress.

The interface forming a hexagonal half loop (figure 4).

It is reasonable to assume that the observed V-shaped misfit dislocations also nucleate in low-mismatch structures. In this case the thicknesses at which the dislocations nucleate are much higher and the branches of the dislocations inclined to the surface are long and tend to form straight lines along the Peierls valleys in the $\{011\}$ directions. Here the model of Fukuda et al. [34] is extended by assuming that, in this case, the dislocations are also moving in a dissociated state. The leading partial of the 60° $\text{As}(g)$ part of the V-shaped loop, which is the first partial to nucleate, is the 30° $\text{As}(g)$ in layers under tensile strain (see figure 6 and the second column of table 2). As will be discussed more quantitatively later, according to this model the dissociation distance of the dislocation is smaller in layers under tensile stress than it is in the stress-free situation. In compressively strained layers the 90° $\text{As}(g)$ now is the leading partial and the dislocation is more widely dissociated.

For the screw dislocation branch of the V-shaped dislocation the situation is different. In the case of tensile stress the Burgers vectors along the dislocation half loop, parallel to the $\{111\}$ planes, are shown (figure 6(a)). The line sense of the dislocation loops in figure 6 is taken to be counter-clockwise. The arrows of the screw dislocation point outwards from the stacking fault area. In this configuration the shear stress along the $\{111\}$ plane increases the dissociation distance. This means that, under tensile stress, the stacking fault area has become larger, so that the screw dislocations cannot re-join and consequently no cross slipping is possible. Relaxation in layers under tensile stress is therefore asymmetrical. In the case of compressive stress the arrows of the screw dislocation are pointing inwards to each other (figure 6(b)). The shear stress, generated by the mismatch, now pushes the Shockley partials together. Cross slipping in order to relieve the strain in the other direction, thereby changing from a mobile $\text{As}(g)$ dislocation to a slower $\text{Ga}(g)$ dislocation, is possible. The configuration of the cross slipped $\text{Ga}(g)$ dislocation is equivalent to the hexagonal half-loop dislocation shown in figure 4(d), in which the bottom part is in the $\{011\}$ direction. When this process is active a symmetrical cross hatched pattern is observed (figure 3). This latter model seems to be consistent with all observations and theoretical considerations.

These considerations about contraction and dissociation of misfit dislocations can be made more quantitatively by the following relations. In a stress-free crystal the equilibrium dissociation width of an extended dislocation $d_0$ is determined by (i) the repulsive interaction of the partials [26] $A/d$, which is inversely proportional to the dissociation distance $d$; and (ii) the attractive force $\gamma$, caused by the stacking fault area, bounded by the Shockley partials. The equilibrium dissociation width is found from the expression

$$\gamma = \frac{A}{d_0}.$$  (1)

In strained layers also misfit stresses $F_l$ act on the partials, causing dislocation motion. This motion is impeded by the presence of frictional forces $R_l$. In figure 7 these forces are indicated. For 60° dislocations in strained crystals the dissociation distance is given by [35]

$$d = \frac{d_0}{1 + (g - \rho)bt/2\gamma}.$$  (2)

in which $g$ is a geometrical factor, determined by the relation $F_l - F_t = gbt$ (subscripts l and t refer to the leading and the trailing partials), $b$ is the Burgers vector of the perfect dislocation, $t$ is the shear stress on the perfect dislocation parallel to the Burgers vector and $\rho$ is given by $\rho = (R_l - R_t)/(R_l + R_t)$.

In the case of screw dislocations the Burgers vectors of both the Shockley partials point either outwards (tensile stress) or inwards (compressive stress), as is shown in figure 6. For these cases the above relation needs to be modified. On each partial the force balance is given by (figure 7)

$$F_l = \frac{\gamma d_0}{d} \pm \gamma + R_l$$  (3)

$$F_t = \frac{\gamma d_0}{d} \pm \gamma + R_t$$  (4)

in which for the repulsive force (1) is used. The upper sign represents the system under tensile stress (figure 7(a)), whereas the lower sign indicates the compressive system (figure 7(b)). Following the procedure of Wessel and Alexander [35], thereby substituting $F_l + F_t = bt$ for the total force on the perfect dislocation in the stationary state of steady motion and $R_l + R_t = (R_l - R_t)/\rho = gbt/\rho$, we can derive from equations (3) and (4) for the dissociation width of the screw dislocations

$$d = \frac{d_0}{1 \mp (1 - g/\rho)bt/2\gamma}.$$  (5)

No exact data are known for the frictional forces $R_l$ in GaAs; for semi-insulating GaAs they are only classified [36]

$$R_{30^\circ \text{Ga}(g)} > R_{30^\circ \text{As}(g)} > R_{90^\circ}.$$  (6)

A crude estimate for $R_{90^\circ}/R_{30^\circ \text{As}(g)}$ is 0.4 [36]. Using this value and data of the dislocation velocities for 60° $\text{As}(g)$ and $\text{Ga}(g)$ dislocations [37], the value of $R_{90^\circ}/R_{30^\circ \text{Ga}(g)}$ is estimated to be $14 \times 10^{-3}$. The results of the calculations of the dissociation distances are given in table 3. From table 3 it appears that, under tensile stress, the 30°$\text{As}(g)$ and the 30°$\text{Ga}(g)$ partials of the screw dislocation part of the hexagonal half loops (figures 4(a) and 4(b)) indeed are pushed away from each other; the dissociation widths
Table 3. Calculated dissociation widths of extended dislocations in (111) planes, as shown in figure 4, according to equations (2) and (5). The values of \( b \), \( \gamma \) and \( \tau \) are taken to be respectively 0.4 nm, 45 mJ m\(^{-2}\) [31] and 42 MPa. The equilibrium distances \( d_0 \) are 3.1 nm for 0° (screw) and 5.4 nm for 60° dislocations.

<table>
<thead>
<tr>
<th>Dislocation type</th>
<th>Leading/trailing partial</th>
<th>( g )</th>
<th>( \rho )</th>
<th>( d/d_0 )</th>
<th>( d ) (nm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tensile</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0°</td>
<td>30 As(g)/30 Ga(g)</td>
<td>−1/3</td>
<td>0.93</td>
<td>1.34</td>
<td>4.1</td>
</tr>
<tr>
<td>60°</td>
<td>90 As(g)/30 As(g)</td>
<td>−1/3</td>
<td>0.43</td>
<td>1.17</td>
<td>6.3</td>
</tr>
<tr>
<td>60°</td>
<td>30 As(g)/90 As(g)</td>
<td>−1/3</td>
<td>−0.43</td>
<td>0.98</td>
<td>5.3</td>
</tr>
<tr>
<td>0°</td>
<td>30 Ga(g)/30 As(g)</td>
<td>−1/3</td>
<td>−0.93</td>
<td>1.14</td>
<td>3.5</td>
</tr>
<tr>
<td>60°</td>
<td>90 Ga(g)/30 Ga(g)</td>
<td>−1/3</td>
<td>0.97</td>
<td>1.33</td>
<td>7.2</td>
</tr>
<tr>
<td>60°</td>
<td>30 Ga(g)/90 Ga(g)</td>
<td>−1/3</td>
<td>−0.97</td>
<td>0.89</td>
<td>4.8</td>
</tr>
<tr>
<td>Compressive</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0°</td>
<td>30 Ga(g)/30 As(g)</td>
<td>1/3</td>
<td>−0.93</td>
<td>0.80</td>
<td>2.4</td>
</tr>
<tr>
<td>60°</td>
<td>30 As(g)/90 As(g)</td>
<td>1/3</td>
<td>−0.43</td>
<td>0.87</td>
<td>4.7</td>
</tr>
<tr>
<td>60°</td>
<td>90 As(g)/30 As(g)</td>
<td>1/3</td>
<td>0.43</td>
<td>1.02</td>
<td>5.5</td>
</tr>
<tr>
<td>0°</td>
<td>30 As(g)/30 Ga(g)</td>
<td>1/3</td>
<td>0.93</td>
<td>0.89</td>
<td>2.7</td>
</tr>
<tr>
<td>60°</td>
<td>30 Ga(g)/90 Ga(g)</td>
<td>1/3</td>
<td>−0.97</td>
<td>0.80</td>
<td>4.3</td>
</tr>
<tr>
<td>60°</td>
<td>90 Ga(g)/30 Ga(g)</td>
<td>1/3</td>
<td>0.97</td>
<td>1.14</td>
<td>6.2</td>
</tr>
</tbody>
</table>

Table 4. A comparison between the calculated and the observed angle \( \phi \) on substrates with different orientations.

<table>
<thead>
<tr>
<th>Substrate orientation</th>
<th>Line direction</th>
<th>( \theta )</th>
<th>Calculated angle ( \phi ) (±0.7°)</th>
<th>Observed angle ( \phi ) (±0.2°)</th>
</tr>
</thead>
<tbody>
<tr>
<td>(100)2°(011)</td>
<td>[011]</td>
<td>2°</td>
<td>2.8°</td>
<td>3.1°</td>
</tr>
<tr>
<td></td>
<td>[011]</td>
<td>0°</td>
<td>0°</td>
<td>0°</td>
</tr>
<tr>
<td>(100)2°(011)</td>
<td>[011]</td>
<td>0°</td>
<td>0°</td>
<td>0°</td>
</tr>
<tr>
<td></td>
<td>[011]</td>
<td>2°</td>
<td>2.8°</td>
<td>3.0°</td>
</tr>
<tr>
<td>(100)2°(110)</td>
<td>[011]</td>
<td>1.4°</td>
<td>2.0°</td>
<td>2.2°</td>
</tr>
<tr>
<td></td>
<td>[011]</td>
<td>1.4°</td>
<td>2.0°</td>
<td>2.1°</td>
</tr>
</tbody>
</table>

Figure 8. The intersection of the (100) surface, misoriented in the [011] direction (cross hatched plane), with the four (111) planes. In the [011] direction the intersecting lines of the two [111] planes with the surface are parallel. In the [011] direction the intersecting lines make a slight angle with each other.

(4.1 and 3.5 nm respectively) are increased compared to the stress-free situation (3.1 nm). Therefore cross slipping of screw dislocations in the tensile system is prevented. In the case of compressive stress (figures 4(c) and 4(d)) the distances between the partials of the screw dislocations (2.4 and 2.7 nm respectively) are decreased. This makes cross slipping of the screw dislocation possible, resulting in relaxation in both the (011) directions. A
The above-described cross slipping process occurs at relatively low growth temperatures. The Ga(g) dislocations appear not to be mobile enough to nucleate and to expand towards the interface. At higher temperatures, above about 950 K, the mobility of the Ga(g) dislocations becomes higher and the dislocations move through the epitaxial layers. The cross slipping process is no longer necessary for strain relaxation. The observed cross hatched patterns on the surfaces of samples grown in the tensile system at 970 K support these considerations (figure 2(c)).

The misfit dislocation half loops introduce new steps at the epi layer surface. These steps can be the origin of a local increase in the growth rate [38,39]. When the local accumulation of misfit dislocations is sufficiently high, and hence the concentration of newly formed surface steps, growth hillocks are formed [40]. These hillocks are observed on surfaces of GaAs samples grown under tensile stress at relatively low temperatures (figure 1). Since the misfit dislocation density increases with layer thickness [19], it follows that the hillock growth is enhanced with increasing layer thickness, as is shown in figure 1(c). In the layers grown under compressive stress the hillocks are absent (figure 3), since in this situation cross slipping prevents local accumulation (piling up) of dislocations. The cross slipping process is not possible for the dislocations in the tensile system. Therefore these hillocks are observed only in the tensile system. At higher growth temperatures (figures 2(b) and 2(c)) the mobilities of the Ga(g) dislocations appear to be sufficiently high to prevent the formation of hillocks at the surface.

4.3. The non-parallel dislocation lines

As was shown earlier, the misfit dislocations in mismatched (100) oriented epi layers form a cross hatched pattern, such that the dislocations run along the (011) directions and cross each other at right angles. However, on vicinal surfaces this is no longer true, as can be seen from figure 3(b), in which an epi layer is shown with a misorientation in the [011] direction. In this direction non-parallel groups of dislocations are observed. This can be explained as follows. The lines of intersection of the (100) surface and the four possible (111) slip planes are the two (011) directions. The intersecting lines of the (111) planes coincide two by two. By a misorientation of the (100) substrate surface, the orientation of the intersecting lines changes too, breaking the coincidence of the intersecting lines of two different (111) planes. This is illustrated in figure 8. It appears from figure 8 that, if the substrate surface is misoriented in one of the dislocation directions, then only in this direction are the dislocation traces on the surface not parallel. If the surface is misoriented in any other than one of these two directions, like the [110] in figure 3(d), then non-parallel groups can be observed in both the dislocation directions.

The above can be verified by comparing the measured and calculated angle \( \phi \) between the non-parallel lines. With the aid of figure 8 one obtains the following relation for \( \phi \):

\[
\tan \left( \frac{\phi}{2} \right) = \sin \theta \frac{1}{\sqrt{2}}
\]

where \( \theta \) is the misorientation of the substrate surface in the direction of the non-parallel dislocation groups, between which the angle \( \phi \) is determined. In table 4 the results of the calculated and observed angles are given. The error in the calculated values of \( \phi \) is the consequence of the inaccuracy of substrate orientation given by the manufacturer. It appears that the observed angles are in good agreement with the calculated ones. This shows that the presence of non-parallel dislocation groups can be used to determine and to check the direction and the magnitude of the misorientation.

5. Summary and conclusions

The surface morphology of partially relaxed lattice-mismatched III-V heterostructures has been studied. The epitaxial layers under tensile stress show an asymmetrical dislocation pattern, whereas at the surface of the layers under compressive stress a (symmetrical) cross hatched pattern is observed. A model has been proposed describing the relaxation in strained layers by the nucleation and propagation of dissociated hexagonal half-loop dislocations. A difference in mobilities between the As(g) and the Ga(g) dislocations in GaAs-based heterostructures is the origin of asymmetrical strain relief in (100)-oriented epitaxial layers under tensile stress at low growth temperatures. According to the model the V-shaped As(g) dislocations, nucleated in layers under tensile stress, are not able to undergo cross slipping in order to relieve the strain in the orthogonal direction, since the dissociation width of the screw dislocation is increased by the shear stress. In compressively strained layers the dissociation width is decreased by the shear stress, enabling the dislocation to undergo cross slipping. In this situation a symmetrical cross hatched pattern is observed at the surface.

Growth hillocks are formed on the surfaces of the layers grown under tensile stress, due to pinning of dislocations. This hillock growth is prevented at higher growth temperatures by the higher mobility of the dislocations. In layers grown under compressive stress, the hillocks are absent because in this situation the dislocations can pass an obstacle by means of cross slipping.

It has been shown that the (mis)orientation of the substrate is revealed by the non-parallel groups of dislocation lines observed in the cross hatched pattern at the surface.

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