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Relaxation study of AlGaAs cladding layers in InGaAs/GaAs (111)B lasers designed for 1.0–1.1 μm operation

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Abstract

InGaAs/GaAs-based lasers require thick AlGaAs cladding layers to provide optical confinement. Although the lattice mismatch between GaAs and AlGaAs is very low, relaxation may occur due to the thickness requirement for an AlGaAs waveguide of the order of microns. We have studied the relaxation of InGaAs/GaAs lasers with AlGaAs waveguides grown on GaAs (111)B substrates. We have observed by transmission electron microscopy (TEM) that certain AlGaAs layers show a high density of threading dislocations (TDs), whilst other AlGaAs layers remain essentially dislocation free. To explain the experimental results a model based on dislocation multiplication has been developed. TDs in the AlGaAs cladding layers are observed when the critical layer thickness (CLT) for dislocation multiplication has been overcome. Consequently, a design rule based on a modified CLT model for AlGaAs/GaAs (111)B is proposed.

Keywords: Laser; Critical layer thickness; Dislocation multiplication; AlGaAs; (111); Transmission electron microscopy; InGaAs; MBE

1. Introduction

There is an industrial requirement for lasers in the 1.0–1.3 μm range in the areas of optical detection of gases, scientific instrumentation, metrology, pumping high power arrays and very recently, in the area of intersatellite communications. At the present time, InGaAs/GaAs (001) laser diodes have been successfully developed for applications in 1.0 μm wavelength. However, efficient and reliable longer wavelength devices have not been successfully developed due to limitations associated with the appearance of dislocations in the strained InGaAs quantum well (QW) and the development of a 3D growth mode for InGaAs with high indium concentrations. The use of InGaAs/GaAs heterostructures grown on GaAs substrates on alternative orientations, such as the (111)B, may offer an extended long wavelength range, for example, low-threshold lasers operating in wavelengths in the range 1.0–1.1 μm have recently been reported [1]. Interest in InGaAs/GaAs (111)B laser heterostructures is due to a larger critical layer thickness (CLT) [2–4] for plastic relaxation, the absence or strong suppression of 3D growth as well as an improved band structure for lasers [5]. In addition, the existence of a large internal piezoelectric field in InGaAs/GaAs (111)B heterostructures [6,7] can have a number of device advantages, including laser tunability.

The growth optimisation of semiconductor structures presents complications due to a large number of parameters, which have to be taken into account for the epitaxial growth of different materials. The laser design involves the growth of AlGaAs, GaAs and InGaAs layers with different thicknesses, alloy compositions and doping levels. Usually, these individual layers are studied separately in order to obtain the optimum growth conditions for each layer. Then the layers are assembled for the continuous growth of the laser structure. In general, although studies of the strained QW layer have monopolised the structural studies of this type of device, other layers in the structure, e.g. the AlGaAs cladding layers may cause problems.

The low lattice mismatch between GaAs and AlGaAs layers tends to induce the belief that dislocation-free AlGaAs cladding layers of *any* thickness can be easily

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Table 1
Structures of the laser samples

Sample	% In	% Al	h (nm)	Substrate misorientation
A	27	40	1.100	M
B	30	40	2.350	S
C	29	45	1.325	S
D	21	60	2.100	M
E	30	66	2.150	S

grown. However, in this paper AlGaAs/InGaAs/GaAs (111)B lasers are observed to show degradation due to dislocations present in the AlGaAs cladding layers. The goal of this work is to present a plastic relaxation model for AlGaAs layers grown in the $[\bar{1}\bar{1}\bar{1}]$ direction in order to explain the dislocation presence in these low lattice mismatch layers.

2. Experimental

Five separate confinement laser heterostructures, with structure as shown in Table 1, were grown on GaAs (111)B substrates by MBE. The active region of these devices consists of a 10 nm single InGaAs QW sandwiched between two 100 nm GaAs barrier layers. The In content in the QW was kept below 30% to avoid strain relaxation and the formation of a misfit dislocation (MD) network [8]. The AlGaAs cladding layer thickness varied from 1 to 2.5 μm and the Al content from 40 to 70%. The values were chosen so as to provide sufficient optical confinement in the active region. The doping levels in the structures were between 10^{17} and 10^{18} cm^{-3}, with the highest values closest to the respective contacts. The samples were grown on (111)B GaAs substrates misoriented 2° towards the $[2\bar{1}\bar{1}]$ and 1° towards the $[\bar{2}11]$, which we name as type S and M, respectively. These two misorientation directions have been shown to be appropriate for the growth of high quality heteroepitaxial layers on the (111)B substrate and they have no significant differences in the plastic relaxation of the structures [9,10].

3. Results

According to previous studies [8], low mismatch $In_xGa_{1-x}As$/GaAs (111)B layers relax through MDs with a 1/2 ⟨110⟩ Burgers vector out of the growth plane (type-I MD). As the mismatch increases, a different MD network develops with dislocation lines in the $\langle 11\bar{2}\rangle$ directions and Burgers vectors of the 1/2 $\langle 1\bar{1}0\rangle$ type lying on the growth plane (type-II MD). However, in all the lasers samples this type-II MD network was not observed in the InGaAs layer due to the use of In contents less than the threshold for the type-II network, which occurs at about $x = 0.30$ for a 10 nm QW [11].

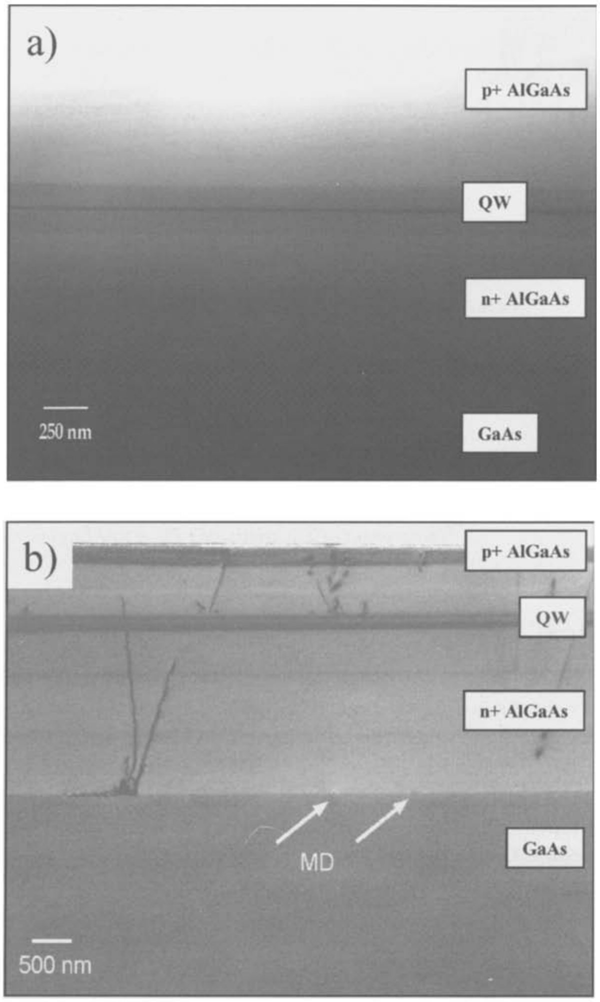

Fig. 1. (a) CS-TEM micrograph of A sample. Note the absence of disloc tions in the $Al_{0.4}Ga_{0.6}As$ cladding layers (b) CS-TEM micrograph of sample showing a high TD density originating at the $Al_{0.66}Ga_{0.34}As$/Ga interface.

The low mismatch between AlAs and GaAs layers (f 0.0018%) often gives rise to the misunderstanding that thi AlGaAs claddings layers in laser structures can be grov without threading dislocations (TDs). Samples A, B and appear to support this, as we observe these to be dislocatio free (see Fig. 1(a)). However, for samples D and E, we c observe that relaxation has started in the AlGaAs due to t presence of TDs at a density exceeding 10^8 cm^{-2} (s Fig. 1(b)). A contrast diffraction TEM study has reveal that these dislocations have a 1/2 ⟨110⟩ Burgers vector out the growth plane (i.e. type-I MD). This dislocation ty is also observed for low mismatched $In_xGa_{1-x}As$/Ga layers and can move by means of a glide mechanis contrary to the type-II MD which can only can move climb [12].

Contrary to the earlier observations, all the AlGa cladding layers used in the lasers we have studied a above the CLT according to the Matthews–Blakesl (MB) model [13]. This model explains the first M formation due to the bending of TDs coming from the substrate. The experimental results are therefore not in agreement with this model and it is therefore necessary to develop another CLT model to predict the dislocation presence in AlGaAs/GaAs (111)B structures.

4. CLT for dislocation multiplication in (111) heteroepitaxies

Several authors have demonstrated that the TD density present in commercial GaAs substrates cannot explain the dislocation density observed in relaxed heteroepitaxial layers. Alternative sources of fresh dislocations, operational at larger strained-layer thickness, have been introduced to explain this fact [14,15]. Whitehouse et al. [16] have obtained experimental evidence of this in the form of a second critical thickness of relaxation. This second critical thickness of relaxation, named CLT_R, may be defined as the thickness at which an epilayer suffers a measurable change in its macroscopic elastic strain. This second CLT is more closely related to indirect measurements of strain relaxation, such as changes in the conductivity or photoluminescence efficiency. The essential difference therefore between CLT_{MD} and CLT_R is that the former defines the introduction of the first MDs inducing a located relaxation in the epilayer and the CLT_R represents the thickness at which a MD network develops sufficient density to significantly relax the entire epilayer. For low mismatched systems, models that predict the beginning of the macroscopic second stage of relaxation (CLT_R) are more useful than CLT models that explain the presence of the first isolated MDs. Our interest is therefore focused in the prediction of this second CLT_R, which indicates the beginning of the degradation of optical and electrical properties. CLT_R is 4–7 times higher than CLT_{MD} for (001) heteroepitaxies.

Dislocation multiplication provides a reasonable way to explain the second stage of the relaxation denoted by CLT_R. In general, the dependence of the relaxation rate with the dislocation glide velocity supports this hypothesis [17,18]. A number of diverse possible mechanisms of dislocation multiplication for III–V semiconductor have been described. However, these can be grouped into two general categories: (a) the interaction between perpendicular MD dislocations [19,20,21] and (b) modified Frank-Read (FR) sources. Beanland [22] has reported that FR mechanisms are operative in low mismatch (001) InGaAs/GaAs layers and has proposed a model that explains the relaxation behaviour with an acceptable agreement with the observed CLT_R [23]. We have considered that this mechanism is also operational in (111) strained layered and have adapted the model for this substrate orientation.

The basic scheme of a FR spiral source is presented in Fig. 2(a). A TD is pinned at point B and only the segment BC slips along its glide plane. In the first step, the segment

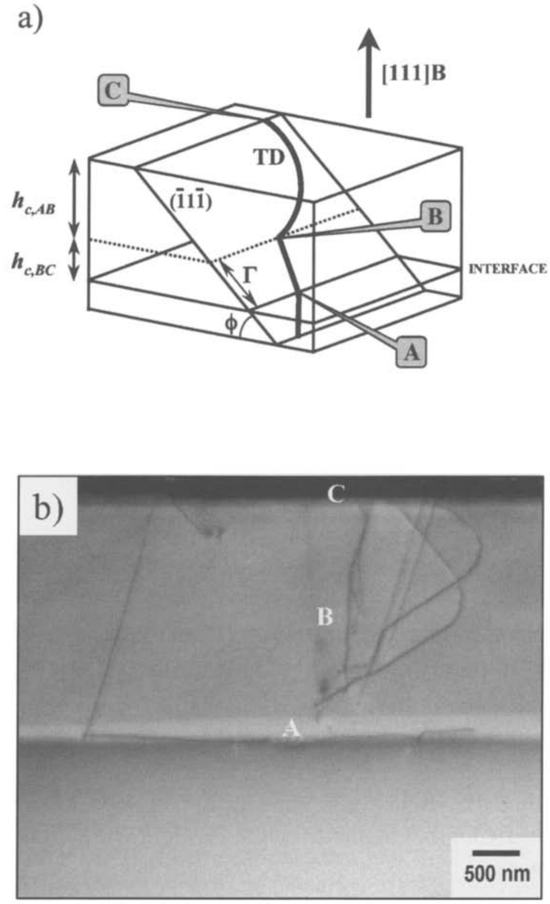

Fig. 2. (a) Schematic representation of a spiral source on a (111) substrate. (b) CS-TEM micrograph of sample E showing a dislocation configuration similar to the proposed model.

BC bends by loop expansion and subsequently a new MD is formed. Finally, the original configuration is reconstructed. We have observed in our samples the experimental evidence of the possible operation of this dislocation source in AlGaAs/GaAs structures. Fig. 2(b) shows a dislocation configuration very similar to the first stage of the spiral mechanism. The segment seems to expand in the form of a large bowing outward, which is very similar to the configuration considered in the model.

This mechanism of dislocation multiplication operates when the glide force due to lattice mismatch overcomes the line tension of each segment. According to Fig. 2, the CLT expression must be the sum of two terms, the first one corresponding to the AB line and the second one to the BC line. The line tension for each segment can be written as follows:

$$F_{\mathrm{line,AB}} = \frac{G_{111}b^2}{4\pi}\frac{1-\nu_{111}\cos^2\theta}{1-\nu_{111}}\ln\left(\alpha\frac{h_c}{b}\right)$$

$$F_{\mathrm{loop,BC}} = \frac{(2+\nu_{111})G_{111}b^2}{2\pi(1-\nu_{111})}\left[\ln\left(\frac{3\alpha h_p}{2\sqrt{2}b}\right)+\frac{\nu_{111}-2}{\nu_{111}+2}\right]$$

where G_{111} is the shear modulus, ν_{111} the Poisson modulus, b the Burgers vector modulus, θ the angle between the Burgers vector and the dislocation line and α the dislocation core.

Acting opposite to the line tension, an elastic strain force due to the lattice mismatch is defined on each segment. This strain is the driving force for dislocation bending in order to form a segment of MD. Caridi and Stark [24] have demonstrated that for (hkl)-oriented fcc pseudomorphic heterostructures, it is always possible to find an appropriate reference system located at the interface plane in order to express the stress tensor as a biaxial stress in function of lattice misfit. Therefore, the glide force for (111) orientation according to Peach–Koehler expression [25] in the new reference system can be reduced by the following expression:

$$F_{\mathrm{strain}} = (\Sigma{\cdot}\mathbf{m})\mathbf{b}\frac{h}{\sin\phi} = \frac{1}{\sqrt{12}}M_{111}fbh$$

where Σ is the biaxial tensor, f is the lattice mismatch, M_{111} is the biaxial module, **m** and **b** are the unit vectors of the glide plane and the Burgers vector in the new basis, respectively, and ϕ is the angle between the glide and the growth plane.

Equating the strain force and the line tension for each segment, a CLT is obtained for the AB and BC segments as follows:

$$\frac{h_{\mathrm{c,AB}}}{b} = \frac{G_{111}}{4\pi}\,\frac{1-\nu_{111}\cos^2\theta}{1-\nu_{111}}\,\frac{\sqrt{12}}{M_{111}f}\ln\left(\alpha\frac{h_{\mathrm{c,AB}}}{b}\right)$$

$$\frac{h_{\mathrm{c,BC}}}{b} = \frac{\sqrt{12}(2+\nu_{111})G_{111}}{2\pi f M_{111}(1-\nu_{111})}\left[\ln\left(\frac{3\alpha h_{\mathrm{c,BC}}}{\sqrt{2}b}\right) + \frac{\nu_{111}-2}{\nu_{111}+2}\right]$$

Both equations provide the conditions for the operation of a spiral FR source. The dislocation multiplication mechanism will occur when the AlGaAs layer thickness surpasses the sum of both CLTs

$$h_{\mathrm{c}} = h_{\mathrm{c,AB}} + h_{\mathrm{c,BC}}$$

Fig. 3 shows the MB CLT and the spiral CLT for the AlGaAs/GaAs system grown in the [111] direction. We can note that the spiral CLT is between 5 and 7 times higher than the MB CLT. Experimental results from laser samples containing AlGaAs cladding layers are presented on the same figure. We observe that only samples D and E overcome the dislocation multiplication CLT and this fact explains well the presence of TDs in the AlGaAs cladding layers of the (111) lasers.

The obtained expression defines the limitations on Al content and thickness in AlGaAs cladding layers for the operation of laser devices in (111) orientation. To avoid dislocation multiplication, any increase in the Al content must be taken into account by a decrease in the layer thickness according to the obtained CLT expression. Probably, the doping level has a considerable influence in the density of pinning points, which is seen as a key aspect for the multiplication model to work. Nevertheless, the formation of fresh dislocations might be expected to occur even in samples without doping. The presence of high doping levels in the laser device possibly implies a predominance of this dislocation source over others. Additional studies of new laser structures with different AlGaAs parameters would be useful in order to contrast and to revise the proposed model.

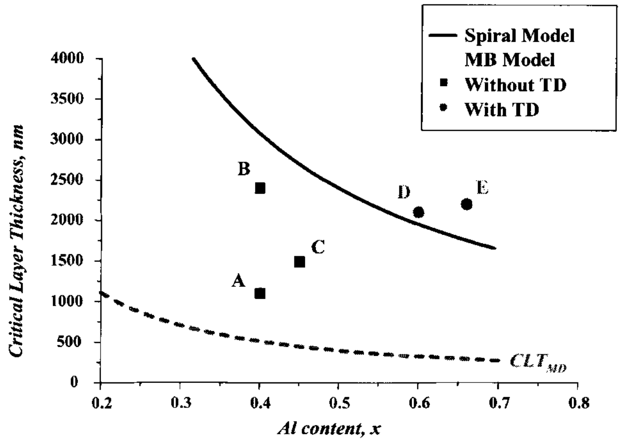

Fig. 3. A representation of the CLT for MB and spiral models in AlGaAs/GaAs layers adapted to the (111) orientation. The spiral CLT curve explains well the experimental results obtained for AlGaAs cladding layers.

5. Conclusions

In the design of InGaAs/GaAs (111)B laser devices, the possibility of plastic relaxation occurring in low mismatched AlGaAs cladding layers is not usually taken into account. In a TEM study of different laser devices, we have observed in AlGaAs layers of high Al content and thickness that the AlGaAs/GaAs misfit can generate TDs. The presence of TDs is expected to seriously degrade the laser performance. For low mismatched systems, models that predict the beginning of a macroscopic second stage of relaxation are more useful than CLT models that explain the presence of the first isolated MDs. We have presented a formulation, based on a dislocation multiplication model adapted for the (111) orientation, that explain with a good agreement the presence of TDs in the AlGaAs layers. In the design of laser structures, the Al content and thickness of AlGaAs cladding layers needs to be chosen carefully not only to optimise the refractive index in the guide, but to avoid defect formation due to lattice mismatch.

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