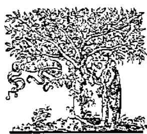

ELSEVIER

Materials Science and Engineering B44 (1997) 106–109

MATERIALS SCIENCE & ENGINEERING B

Structural study of AlGaAs/InGaAs superlattices grown by MBE on (111)B GaAs substrates

T.C. Rojas [a,*], S.I. Molina [a], M.J. Romero [a], A. Sacedón [b], E. Calleja [b], R. García [a]

[a] *Departamento de Ciencia de los Materiales e I.M. y Q.I. Universidad de Cádiz, Apdo. 40, 11510 Puerto Real (Cádiz), Spain*

[b] *Departamento de Ingeniería Electrónica, E.T.S.I. Telecomunicación, Universidad Politécnica de Madrid, Ciudad Universitaria s/n, 28040 Madrid, Spain*

Abstract

AlGaAs/InGaAs strained layers superlattices (SLS) have been grown by molecular beam epitaxy on (111)B and (100) GaAs substrates at different temperatures (520, 540 and 560°C). The heterostructures are studied by transmission electron microscopy, scanning electron microscopy and low-temperature photoluminescence. The controversy concerning the growth temperature for In-based/Al-based III–V alloys on (111)B GaAs is discussed. In this context, the (111)B and their corresponding (100) structures are compared. The best growth temperature is 560°C, as a smooth surface and the lowest defect (dislocations, planar defects and stacking faults tetrahedra) density is obtained at this temperature. © 1997 Elsevier Science S.A.

Keywords: Alloys; Superlattices; Molecular beam epitaxy

1. Introduction

The AlGaAs/InGaAs heterostructures grown on (111)B GaAs substrates exhibit advantages with regard to (100) GaAs for new applications in optoelectronic devices. Such improvements are due to the large internal electric fields generated by the piezoelectric effect in coherently strained (111) layers [1,2]. Quantum wells (QWs) grown on the (111) orientation possess enhanced optical transitions and smaller in-plane electron effective mass, allowing the design of low-threshold current density QWs lasers [4]. Theoretical calculations and experimental investigations, recently confirmed [3–6], predict coherent blue emission for the (111) structures as a consequence of their high second harmonic conversion efficiency. Therefore, interesting optical nonlinearities properties are obtained [2,7]. Due to strain-induced electric fields, the depletion or accumulation of a two-dimensional (2D) carrier gas has been reported to produce large carrier densities without modulation doping [8].

Unfortunately, the growth of heterostructures on (111)B GaAs substrates presents several problems:

* Corresponding author.

0921-5107/97/$17.00 © 1997 Elsevier Science S.A. All rights reserved.
PII S0921-5107(96)01737-0

First, when growing InGaAs and AlGaAs layers simultaneously, different growth conditions for each are ideally required [9–11]. Indeed, AlGaAs films with specular surfaces need high growth temperatures with low Ar gas flux, whereas InGaAs films require low growth temperatures, due to the low In sublimation temperature. This discrepancy between both temperatures makes the growth of InGaAs/AlGaAs heterostructures with high crystalline quality difficult.

Second, tilted substrates towards the [211] direction are necessary to obtain low defect (stacking faults and triangular pyramids) density structures [12]. The range of tilt angle is between 1° and 4°. To improve the growth temperature in such heterostructures, we study AlGaAs/InGaAs strained layers here, grown by molecular beam epitaxy (MBE) on (100) and (111)B GaAs substrates 1° misoriented towards the [211] direction. Samples grown on (100) GaAs are used as references.

2. Experimental procedure

The grown sample structure is schematically described in Fig. 1. The AlGaAs/InGaAs superlattices were grown at 520, 540 and 560°C, labelled as A, B and C for the (111)B samples and A*, B* and C* for the

(100) ones, respectively. The growth rate used was 0.36 μm h^{-1} and the V–III ratio was 3. Specimens were prepared for study by transmission electron microscopy (TEM) by mechanical thinning, followed either by Ar^+ ion milling for cross-sectional observations and by chemical etching ($H_2SO_4 + H_2O_2 + H_2O$ and Br_2–methanol) for the planar view orientation. TEM and scanning electron microscopy (SEM) observations were performed in a JEOL 1200EX transmission electron microscope and a JEOL 820SM scanning electron microscope, respectively.

3. Experimental results and discussion

Cross-sectional transmission electron microscopy (XTEM) and planar-view transmission electron microscopy (PVTEM) observations of samples A and B are displayed in Fig. 2(a) and Fig. 3(a), respectively. Both samples manifest a high density of defects (dislocations, planar defects and microtwins), most of which are planar defects, commonly associated with the growth on (111) substrates. The defect distribution is less homogeneous in sample B, with vicinal regions of

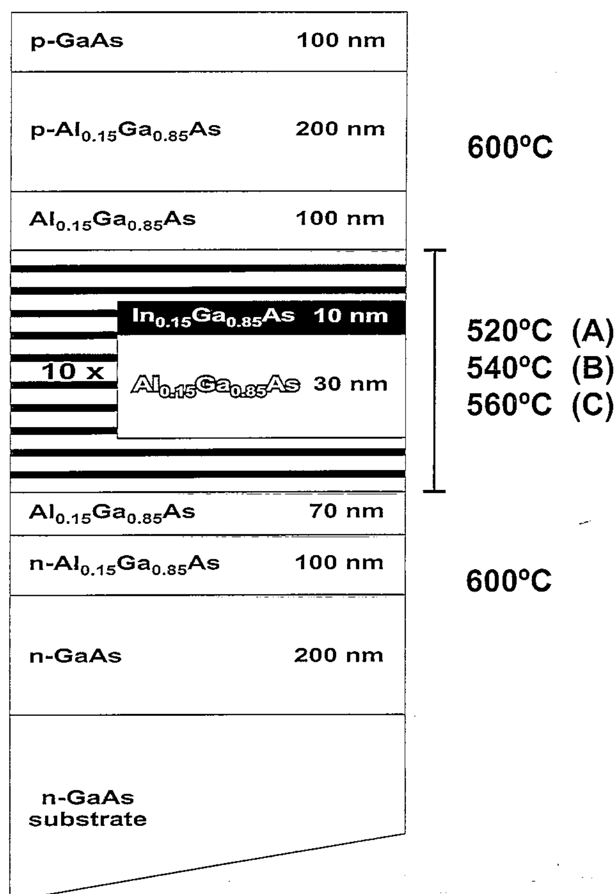


Fig. 1. Sample design scheme.

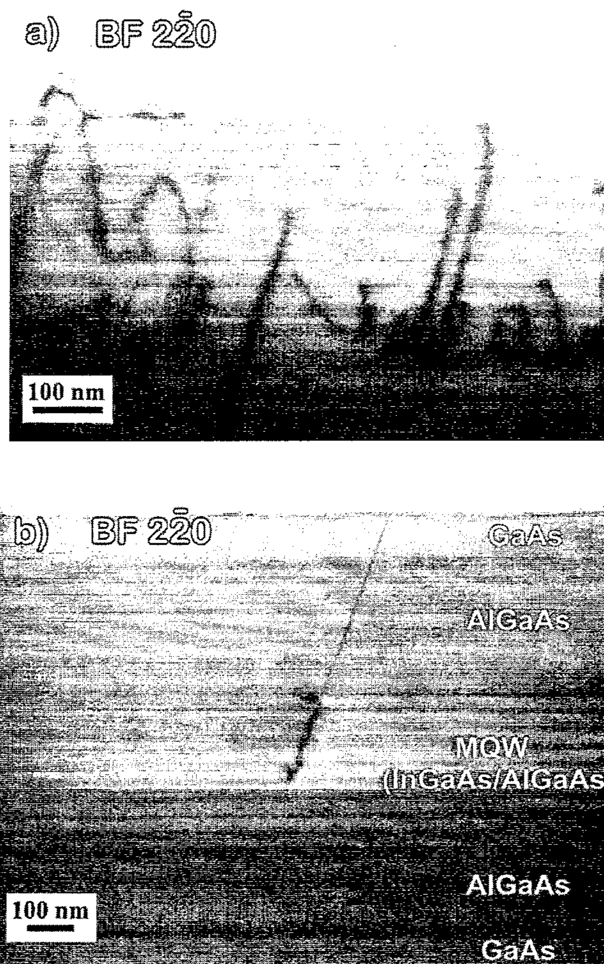
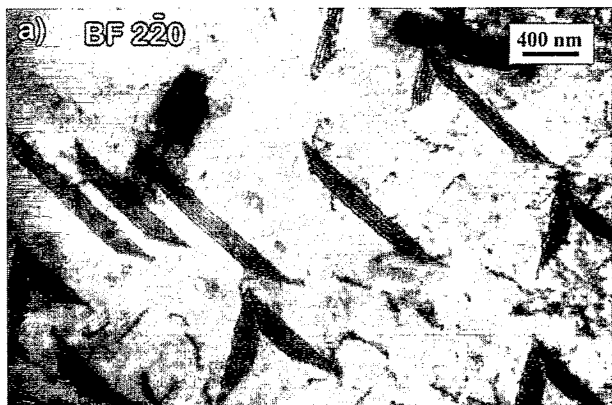


Fig. 2. (2-20) Bright-field XTEM images of (a) sample B and (b) sample C, grown at 540 and 560°C, respectively.

low and high defect density, whereas sample A presents a homogeneous higher defect density. Both have some dislocations originated at the AlGaAs/GaAs and InGaAs/AlGaAs interfaces that thread through the whole structure. The observed planar defects are mainly stacking fault tetrahedra. A wavy appearance is associated with these defects, particularly for sample B, as observed in Fig. 3(a). This fact could be due to interface roughness. Likewise, the surface micro-morphology observed by SEM is in agreement with the defect distribution detected by TEM. A faceted morphology and a high density of deep triangular pyramids are observed in both samples. The latter suggests a low mobility of Al atoms that increases direct nucleation on the terraces with respect to the steps, hence producing facets and stacking faults in samples grown at low temperatures. The distribution of these defects is quite inhomogeneous, showing regions with low and high densities. Fig. 4(a) shows a SEM micrograph of sample B that corresponds to an area with a high density of pyramids.

On the other hand, the behaviour of sample C is different compared with samples A and B. XTEM and PVTEM micrographs of sample C are shown in Fig. 2(b) and Fig. 3(b), respectively. A low threading dislocation density is observed in the AlGaAs layers below the superlattice, without penetrating into it. A very low concentration of planar defects, originating in InGaAs/AlGaAs layers (see Fig. 2(b)), and some isolated intersections between them, which are located in different (111) planes, are also evident. The surface morphology is almost speculate and smooth without any facets, but regions with some triangular pyramids are observed as shown in Fig. 4(b). On the contrary, sample C* shows a defect-free epilayer, and only some misfit dislocations located at the first InGaAs/AlGaAs heterointerface are to be seen.

Low-temperature photoluminescence (PL) measurements are used as a way to assess the SLS quality. From samples grown on (111)B substrates, only sample C luminescence is enough for PL detection. Fig. 5 shows the PL spectra of samples C and C*. The PL



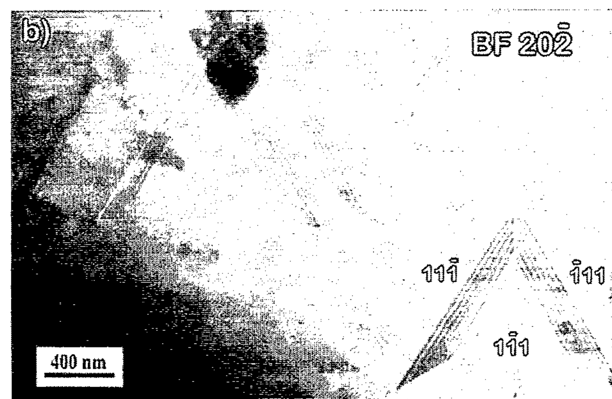


Fig. 3. Bright-field (2-20) and (20-2) PVTEM images of (a) sample B and (b) sample C, respectively. The *R* vectors of the stacking faults are indicated in Fig. 3(b). Note that one of the stacking faults is not visible as expected from the invisibility criteria.

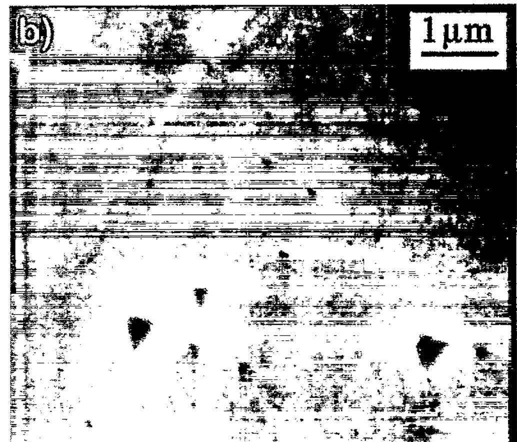


Fig. 4. SEM images of (a) sample B and (b) sample C, grown at 540 and 560°C, respectively. Fig. 4(a) shows an area with a faceted surface morphology and a high density of triangular pyramids, while Fig. 4(b) shows a smoother surface, although some not very deep triangular defects are still present.

peak intensity of sample C is clearly smaller than that recorded for C*. The peak full width at half maximum (FWHM) is broad (17 meV), compared with that obtained for its reference (6 meV), and a slight shift towards high energy occurs. To explain the PL behaviour, we propose the following mechanism.

The unique structural difference between samples C and C* is their defect density. Therefore, knowing that their growth temperature (560°C) is higher than the ideal one for InGaAs growth (540°C) [11], In species will tend to diffuse across the existing defects. This In diffusion will change the barrier height of the AlGaAs layers, becoming an AlInGaAs alloy and will increase the well width. Therefore, the In content of the wells will be lower for the deeper ones.

To improve the flatness of the AlGaAs growth front, a higher growth temperature is required. However, a lower growth temperature is necessary to achieve the

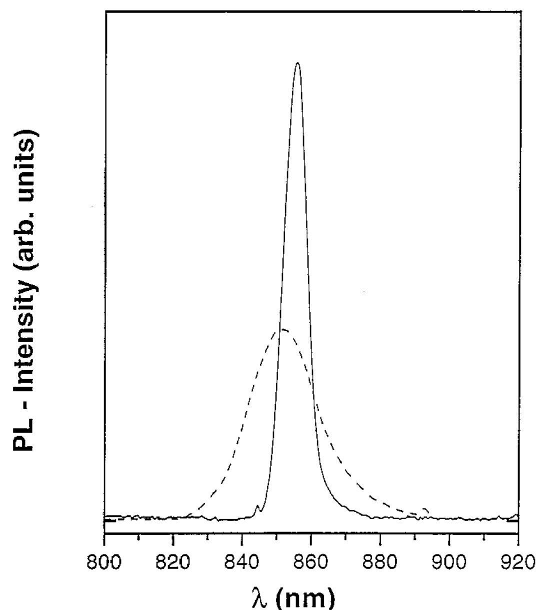


Fig. 5. Low-temperature (4 K) PL spectra of samples C (– – –) and C* (——), grown on (111)B and (100) substrates, respectively, at 560°C.

required InGaAs alloy quality. Since this trade-off is not suitable, another approach has to be considered to improve the structural quality on these SLSs. We propose that they should be grown on substrates where the distance between steps becomes smaller, i.e., where the substrate surface misorientation angle was larger. This would reduce the problems produced by the low mobility of Al adatoms, i.e., the defect density would be improved and would allow growth at lower temperatures compatible with the growth of InGaAs layers.

4. Conclusions

InGaAs/AlGaAs SLS grown by MBE at three different temperatures on (100) and (111)B GaAs substrates are studied by SEM, TEM and PL. The set of results obtained by TEM and SEM lead to the defect density in InGaAs/AlGaAs SLS heterostructures grown on (111)B GaAs substrates decreasing as the growth temperature rises, i.e., the microstructural quality is better at the higher proposed growth temperature of 560°C. On the other hand, the InGaAs layers structural quality and compositional homogeneity are better for (100) structures than for (111)B ones. We propose the use of larger substrate surface misorientation angles to improve the structural quality of the studied (111)B InGaAs/AlGaAs.

Acknowledgements

This work was made possible through grants from the Comisión Interministerial de Ciencia y Tecnología (CICYT) (MAT94-0538) and the Junta de Andalucía (Group 6020). The work was carried out at the Electron Microscopy Division of the University of Cádiz. The authors would like to thank J.M. Geraldía for the technical support during this study and Dr D. Araújo for his fruitful comments.

References

[1] D.L. Smith, *Solid State Commun.*, *57* (1986) 919–921.
[2] D.L. Smith and C. Mailhiot, *Rev. Mod. Phys.*, *62* (1990) 173–234.
[3] E.A. Caridi, T.Y. Chang, K.W. Goossen and L.F. Eastman, *Appl. Phys.*, *56* (1990) 659.
[4] T. Hayakawa, M. Kondo, T. Suyama, K. Tatakahashi, S. Yamamoto and T. Hijikata, *Jpn. Appl. Phys.*, *26* (1987) L302.
[5] T.S. Moise L.J. Guido, J.C. Beggy, T.J. Cunningham, S. Seshadti and R.C. Barker, *J. Electron. Mater.*, *21* (1) (1992).
[6] J.L. Sanchez Rojas, E. Muñoz, A.S. Pabla, J.P.R. David, G.J. Rees, J. Woodhead and P.N. Robson, *GaAs and Related Compounds*, Freiburg, 1993.
[7] M. Livingstone, I. Galbraith and B.S. Wherrett, *Appl. Phys. Lett.*, *65* (2) (1994) 28.
[8] X. Li, K.F. Longenbach and W.I. Wang, *Appl. Phys.Lett.*, *60* (12) (1992) 23.
[9] A. Chin, P. Martin, P. Ho, J. Ballingall, T.-H. Yu and J. Mazurowski, *Appl. Phys. Lett.*, *59* (15) (1991).
[10] A. Ching, P. Martin, U. Das, J. Ballingall and T.-H. Yu, *J. Vac. Technol. B*, *10* (2) (1992).
[11] Ana Sacedón, *Ph.D. Thesis*, Polytechnical University of Madrid, 1995.
[12] K. Yang, L.J. Schowalter, B.K. Laurich, I.H. Campell and D.L. Smith, *J. Vac. Sci. Technol. B*, *11* (3) (1993).