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Microstructure Studies of GaInAsSb/GaSb Heterostructure

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Abstract

The preliminary results of scanning transmission electron microscope investigations on the interface of GaInAsSb/GaSb heterostructure were presented. STEM images show that the various dislocations and stacking faults were produced by the lattice mismatch between the quaternary GaInAsSb alloy and GaSb substrates. All these defects were the ways for relieving the misfit strain, including 60° dislocation, 90° dislocation and stacking faults. It is also found that only the 90° dislocation can form the ridges on the surface of the epilayer.

Key Words: Dislocation, Stacking fault, STEM, GaInAsSb

1. Introduction

In recent years, there has been considerable interest in the quaternary GaInAsSb alloy system for photodetectors, which have a variety of commercial applications in air pollution, industrial process control, automobile emission monitoring, and future lightwave communication system using novel fiber materials. Since the quaternary GaInAsSb alloy system has a direct bandgap adjustable in wavelength from 1.7 to 4.5 μm when grown lattice-matched on GaSb, it may provide the basis for emitters and detectors over this entire region.

For most applications, the operation at room temperature is very important for attaining the desired system performance at reasonable cost[1]. The quaternary GaInAsSb alloys are the most promising candidate materials alternative to the HgCdTe system for use in infrared detectors. Furthermore, the GaInAsSb quaternary alloys have more advantages than ternary alloys. It is due to the flexibility offered by quaternary alloys, which select both energy bandgap, E_g, and lattice constant, a_0, independently. Such flexibility is important since:

(1) The high quality material required for most devices can be produced when the epitaxial layer is lattice matched to the substrate, if the lattice parameter is mismatch, it will be easy to generate dislocations which can degrade the material qualities and device performance.

(2) The high quality interfaces required for heterostructure devices are obtainable only when E_g is changed without changing a_0.

In a ternary alloy, E_g and a_0 are generally both functions of a single composition parameter, so they can not be selected independently.

As mentioned above, the generation of threading dislocations as well as misfit dislocations is the fundamental concern for lattice mismatch heteroepitaxy. In order to reduce the densities of dislocations of the materials for device applications, it is essential to investigate both the formation and motion of these dislocations.

2. Experimental

The Quaternary GaInAsSb alloy was grown on GaSb substrates by MOCVD using a conventional atmospheric pressure horizontal reactor. The sources of Ga, In, Sb and As atoms were Trimethylogallium (TMGa), Trimethlogallium (TMIn), Trimethlogllium (TMSb) and arsine (AsH_3) Diluted to 10% in Hydrogen, respectively. TMGa, TMIn and TMSb were held in temperature baths at − 12, 17 and − 10℃, respectively, and carried by Pd-diffused hydrogen into the reactor. The substrates were n-type GaSb oriented 2°-off (001) towards [011]. GaSb substrates were chemically polished by a solution of $HNO_3 : HCl : CH_3COOH = 0.2 : 2 : 20$, before being put into reactor. The growth temperature was 610℃. The Ⅲ/Ⅴ ratio was 0.792, TMGa/(TMGa + TMIn) ratio was 0.466, TMSb/(TMSb + AsH_3) ratio was 0.869. The growth time was 50 min. The observations were made by using the vacuum generator HB501 scanning transmission electron microscopy and the AFM.

3. Results and Discussion

3.1. *STEM Image of Interface of GaInAsSb / GaSb Heterostructure*

Fig. 1 is the interface STEM image with bright field mode of the $Ga_{0.88}In_{0.12}As_{0.18}Sb_{0.82}$ epilayer. The threading dislocation, 60° dislocation, 90° dislocation, and stacking fault were observed. By these produced defects, the mismatch relaxation in a single layer has been widely studied in the literature [2], and three stages can be distinguished[3].

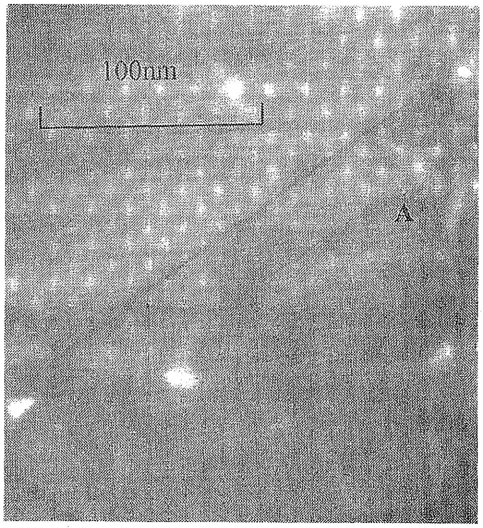

Fig. 1. STEM image close to substrate.

In the first one, the relaxation is slow since only the bending of existing dislocations coming from the substrate occurs as described by Matthews *et al*[4]. This stage, the epitaxial thickness to large lattice mismatch is so small that it is not observed in Fig. 1.

In the second one, when the layer is thick enough for the multiplication of dislocations[5], a fast relaxation occurs[6]. In a diamond or a zinc-blend structure, the Burgers vector for a complete misfit 60° or 90° dislocations usually lies on the {001} interface. The misfit strain relieved by a pure edge 90° dislocation is twice times more than that by a 60° dislocations; therefore, the formation of 90° dislocations would provide the most energetically favorable strain relaxation. While 90° dislocations can relieve strain most effectively at the interface without any threading component, 60° misfit dislocations are highly active sources for the generation of threading dislocations[7]. The significance of the sole existence of the 90° dislocations in our system is that they are more beneficial to mismatched growth than the 60° dislocations, which were seen in Fig. 1, and these dislocations propagate towards the epilayer surfaces.

In the third stage, an inhibition of the relaxation occurs for much thicker layer, due to work-hardening process in the material[8]. For the sake of thin epilayer, this stage is not observed in this STEM experiment.

Fig. 1 shows an overlapping stacking fault labeled A in the < 110 > and < $\bar{1}11$ > directions. Stacking faults are important defects in Ⅲ-Ⅴ epitaxial growth[9], but their nature can vary. They have been characterized as Frank stacking faults[10], Schockley stacking faults[11], or mixing of Frank and Schockley stacking faults[12]. The identification of the nature of the stacking faults is important because they do not result from the same nucleation mechanism. Frank stacking faults are formed by the collapse of vacancies while Schockley stacking fa[illegible]s are growing defects. We demonstrat[illegible] that strain relaxation of GaInAsSb/[illegible]Sb heterostructure occurs through the fo[illegible]ation of thin twin lamellae on {111} [illegible]anes provided that the biaxial stress fiel[illegible] is tensile. In this case the geometrical [illegible]rrangement of the atoms on the closed-p[illegible]ked {111} planes requires that a misfit [illegible]0° 1/6 < 112 > Schockley partial dislo[illegible]ion have to nucleate in order to resolve [illegible]rt of the mismatch induced strain. Altho[illegible] a 30° Schockley partial dislocation cou[illegible] in principle follow the 90° partial disloca[illegible]n forming a dissociated 60° 1/2 < 11[illegible] dislocation, it is energetically favorable [illegible] introduce further 90° partial dislocation [illegible]n adjacent {111} slip plane. A stackin[illegible] fault is characterized by its displacem[illegible]t vector R which can take 1/6 < 112 [illegible] or 1/3 < 111 > values and by the [illegible]artial dislocations bounding it which can [illegible]ve Burgers vectors b = 1/6 < 112 > o[illegible]/3 < 111 >. A fault bounded by a 1/[illegible] < 112 > partial dislocation is a Schock[illegible] fault and a fault bounded by a 1/3 < [illegible]1 > partial dislocation is a Frank fault[illegible]

3.2. *Dislocations in [illegible]ilayer*

In Fig. 2, the 6[illegible] dislocation is on {111} planes. The fo[illegible]ation of 60° dislocation displacement re[illegible]ts in the creation of a pair of planes (c[illegible]on and anion lattice planes). Since s[illegible] Ⅱ-Ⅵ and Ⅲ-Ⅴ compound semicon[illegible]tors have a zinc-blend structure consisti[illegible] of two interpenetrating face-centered [illegible]cubic sublattice, there are two types of [illegible]rtial edge dislocations depending on wh[illegible]er the extra-half planes are bound by [illegible]tial edge dislocations terminated on cat[illegible]s and anions, respectively. These disl[illegible]ations lie on the same {111} plane and have the same Burgers vector, and they constitute a "pile-up" from some multiplication source. Misfit dislocation multiplication produres closed-spaced bunches of dislocations which lie inside the epitaxial layer. The 90° dislocations are favorable on energy since they are mostly effective at accommodating the lattice mismatch. As a result, the GaInAsSb epilayer is almost fully relaxed with 60° and 90° dislocations, and very little threading dislocations propagating to the surface.

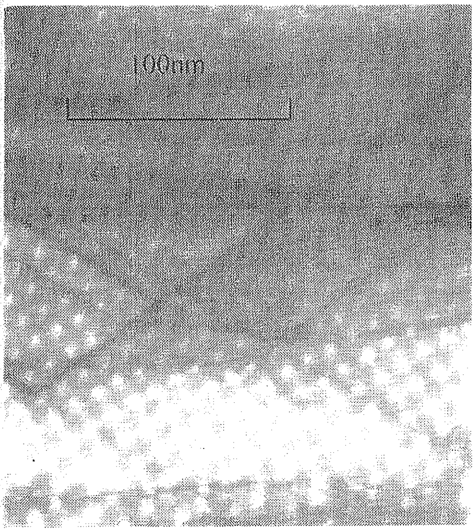

Fig. 2. STEM image between substrate and surface.

3.3. *90° Dislocation Induced Morphological Features*

Fig. 3 is an AFM surface image of the $Ga_{0.88}In_{0.12}As_{0.18}Sb_{0.82}$ epilayer, and many ridges may be observed. It seems likely that there is a finite residual strain in the deposit. This could arise because there is a significant driving force required for the introduction of misfit dislocations and thus any residual strain could be at the level which gives insufficient strain energy for dislocation introduction. Since {100} planes are not slip planes in the GaInAsSb epilayer, the 90° misfit dislocations whose Burgers vectors lie on the {001} plane are not expected to move by slipping under normal condition. The strain field of these bunches of dislocation is strong and they lie sufficiently close to the surface to give locally enhanced growth. Ridges are produced above these "pile-up" as the misfit strain is relieved. The surface ridges maintain their profile after the enhancement ceases.

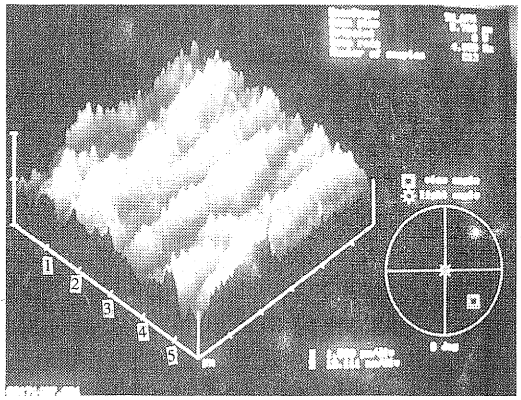

Fig. 3. AFM image of GaInAsSb epilayer.

Fig. 4 is the STEM image of the cross-section close to surface. It is observed that only 90° dislocations may form the ridges, while 60° dislocations may not. This is because 90° dislocations give a locally enhanced growth *in situ*., but 60° dislocations may slip all epilayer.

4. Conclusion

In summary, we found that the GaInAsSb epilayer is almost fully relaxed with 60° dislocation, 90° dislocation, very little threading dislocations propagating to the surface, and on the surface of the epilayer, only 90° dislocation can form the ridges.

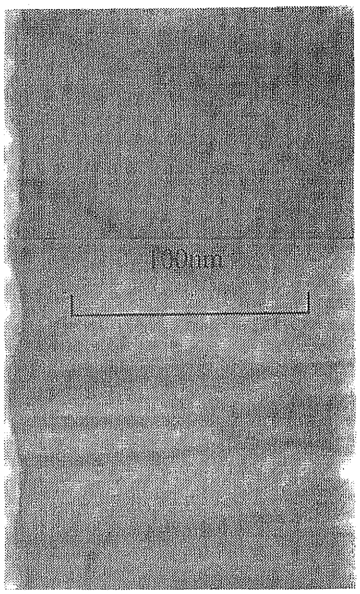

Fig.4. STEM image close to surface.

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