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Epitaxial growth of low threading dislocation density InSb on GaAs using self-assembled periodic interfacial misfit dislocations

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ABSTRACT

We report a fully relaxed low threading dislocation density InSb layer grown on a GaAs substrate using self-assembled periodic interfacial misfit dislocations. The InSb layer was grown at 310 ^{o}C by molecular beam epitaxy. The AFM measurement exhibited a root mean square (r.m.s.) roughness of 1.1 nm. ω-*2θ* scan results from x-ray diffraction measurement indicated that the InSb layer is 98.9% relaxed. Images from the transmission electron microscope measurement showed a threading dislocation density of 1.38×10^{8} cm^{-2}. The formation of interfacial misfit dislocations was also observed. The InSb layer exhibited a 33840 cm^{2}/V s room temperature electron mobility.

Keywords: Thin films; Epitaxial Growth; TEM; Structural; Semiconductors

1. Introduction

Indium antimonide (InSb) has attracted great interest because of its small bandgap energy of 0.18eV and its high electron mobility of up to 77,000 cm^{2}/V s [1].These properties make InSb an ideal candidate for thermal imaging applications and channel material in ultra-high speed transistor. Datta *et al.* [2] reported an InSb-based high electron mobility transistor (HEMT) with an operating speed up to 305 GHz. Furthermore, Sat *et al.* [3] suggested that the cutoff frequency for InSb-based HEMTs could likely exceed the 1 THz barrier. The lack of a semi-insulating (SI) substrate is an obvious disadvantage of InSb. A SI substrate minimizes the intrinsic parasitic capacitances in devices. The integration of an InSb layer on a SI GaAs substrate not only provides the benefit of minimal intrinsic capacitances, it also enables the co-existence of InSb-based devices and GaAs-based devices on a single wafer [4]. However, the direct growth of a InSb layer on a GaAs substrate was found to produce a high density of threading dislocations (TD) [5]. The use of an InP, InAlP and AlSb metamorphic buffer layer to overcome the constraint of a large lattice mismatch was reported [6,7]. However, a thick buffer layer (>1 μm) may not be favorable for integration with other GaAs-based devices as the resulting non-planar device arrangements require lengthy interconnect metal contacts. A two-step method, which requires a low temperature InSb growth followed by a higher temperature growth, was reported to promote the two dimensional (2D) growth mode [8,9]. However, the surface roughness remained high (>3nm r.m.s).

The formation of dislocations at the interface to relieve the strain energy is inevitable in a high lattice-mismatched heteroepitaxial system such as InSb on GaAs. Typically, two types of interfacial misfit (IMF) dislocations can be found at the heteroepitaxial interface: (1) a 60° dislocation and (2) a 90° dislocation [10]. In contrast to the 60° dislocation, a 90° dislocation rarely contributes to the formation of TDs [11]. Modelling results showed that a 90° dislocation relieves twice the amount of strain energy compared to a 60° dislocation [12]. A novel method to form a 90° IMF dislocation array at the GaSb/GaAs interface has been reported [11,13,14]. Density of TDs in the GaSb film grown on a GaAs substrate was reported to be significantly reduced to 5.4×10^{5} cm^{-2} [15]. In this letter, we report the heteroepitaxial growth of a low TDs density InSb layer on a (100) GaAs substrate using the IMF technique. The properties of the InSb layer were characterized by atomic force microscopy (AFM), high resolution x-ray diffraction (HR-XRD), and transmission electron microscopy (TEM).

2. Experimental details

The samples were grown using a solid-source molecular beam epitaxy reactor with valved cracker As and Sb sources. After oxide desorption of the GaAs substrate at 580 ℃, a 200 nm GaAs buffer was deposited to provide a smooth GaAs surface for the subsequent growth. Subsequently, a growth interruption was introduced with the closing of the As valve because the formation of the IMF layer can be impeded by the presence of an As-rich surface [16].

The substrate temperature was subsequently decreased to 310 ^{o}C and the sample was exposed to a Sb_2 species with a beam equivalent flux of ~1.2×10^{-6} Torr for 4 minutes. After exposure to the Sb flux, the RHEED pattern changed from a (2×4) As-stabilized surface reconstruction to a (1×3) Sb-stabilized surface reconstruction, which indicated the absorption of Sb atoms on the GaAs surface. This observation implied that an anions exchange between As atoms and Sb atoms had taken place. After the Sb exposure step, the growth of a 730 nm-thick InSb layer was started by opening the indium shutter. A Sb/In beam equivalent pressure (BEP) ratio of 4 and a growth rate of 0.9 μm/h were used. A spotty RHEED pattern was observed at the beginning of the growth and gradually evolved to streaky (1×3) lines after ~20 seconds of growth, which suggested a transition from the three dimensional (3D) growth mode to the two dimensional (2D) growth mode.

3. Results and discussion

Bright field TEM images of the InSb layer on GaAs are shown in Fig. 1(a)-(c). Fig. 1(a) shows a clean InSb/GaAs interface with only a few dislocations rising from the interface. We observed only 2 TDs (marked by white arrows) that reached the surface of the 720 nm-thick InSb layer over a length of 1.7 μm across the surface. Hence, the TD density of the InSb layer is estimated to be 1.38×10^{8} cm^{-2}.

The existence of threading dislocations could likely be due to the coalescence of islands during the initial phase of InSb growth, when the spotty RHEED pattern was observed.

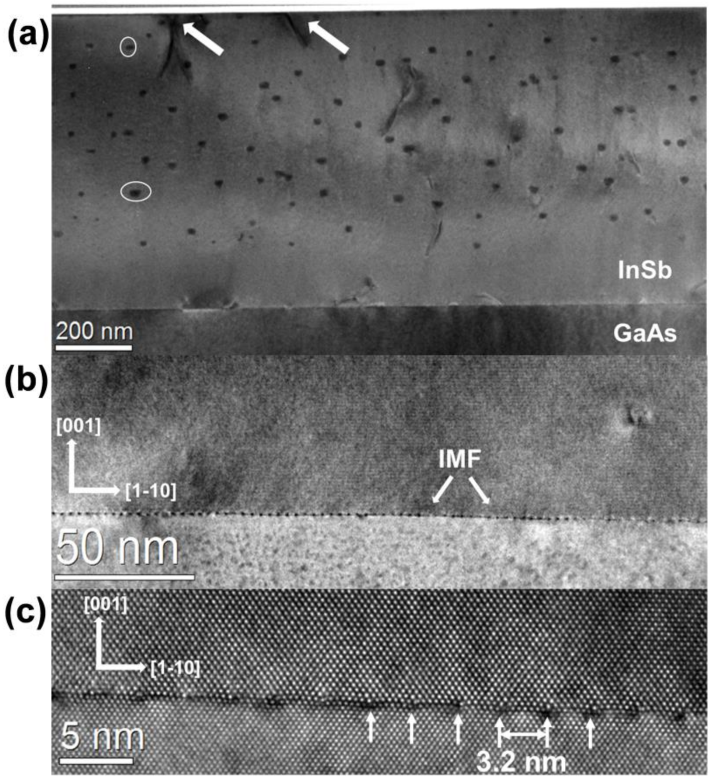

Fig. 1. Cross-sectional TEM images of a 730 nm InSb layer on a GaAs (100) substrate. (a) A low magnification (12 k) TEM image. The In rich clusters are marked by a white circle. The locations of two threading dislocations are indicated by the white arrows. (b) A 100K magnification TEM image illustrating highly periodic misfit dislocations at the GaAs/InSb interface. (c) A 400K magnification TEM image showing the separation between two adjacent periodic interfacial misfit dislocations.

In addition to the TDs, black spots were observed in the InSb layer as shown in Fig. 1(a). Energy dispersive x-ray (EDX) analysis of these spots revealed that they are In-rich clusters, where the In to Sb atomic ratio is 13% higher compared to the ratio at the area outside the cluster. These In-rich clusters could likely be due to excessive indium flux or an insufficient Sb/In ratio was used in the growth of the InSb layer. Fig. 1(b) shows a high resolution TEM image taken at the InSb/GaAs interface where periodic IMF dislocations can be observed. A higher magnification TEM image for these periodic dislocations is shown in Fig. 1(c). It can be seen that the dislocations were uniformly distributed across the InSb/GaAs interface. The distance between two adjacent dislocations is ~3.2 nm.

The lattice constant of InSb and GaAs are 6.479 Å and 5.653 Å, respectively. Thus, it is energetically favorable for every seventh Sb atom in the InSb layer to skip one Ga atom in the GaAs layer to decrease the lattice strain energy of the Ga-Sb bonds at the InSb/GaAs interface [17]. This phenomena was observed in the TEM image as shown in white boxes in Fig. 2 The separation between two adjacent IMF 90^{o} dislocations can be calculated using the following equation: [17, 18]

$$S = \frac{b}{f} \tag{1}$$

where S is the separation distance, b is the Burgers vector of misfit dislocation and f is the lattice mismatch. The details of the 90° misfit dislocation with the Burgers vector $\frac{a}{2}[1\bar{1}0]$ (marked by the Burgers circuit in black dashed arrows) is shown in Fig.2. Substituting a $|b|$ value of 4.58 Å and a lattice mismatch of 14.6% into Eq. (1) give a separation distance between two adjacent dislocations of 3.19 nm. This calculated value is consistent with the 3.2 nm distance of separation observed in Fig. 1(c).

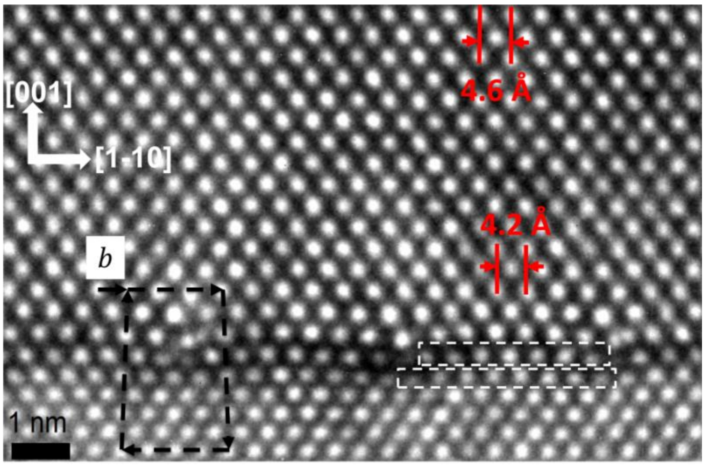

Fig. 2. A high resolution cross-sectional TEM image at the GaAs/InSb interface. Dislocations are indicated by black arrows. The black dashed arrows are a Burgers circuit around a 90° misfit dislocation with a Burgers vector b=a/2($1\bar{1}0$) at the interface. The box with a white line and the box with a dotted line show the atomic sites between adjacent dislocations in InSb and GaAs, respectively.

For bulk InSb, the distance between two adjacent atoms in the ($1\bar{1}0$) direction is expected to be 4.6 Å. However, it can be seen in Fig. 2 that this distance for atoms, which were located ~2 nm from the InSb/GaAs interface, was found to be ~4.2 Å. This indicated that the strain energy is not fully relaxed at the interface. At atomic sites that were ~5 nm from the InSb/GaAs interface, the distance between two adjacent atoms increased to ~4.6 Å, implying that the InSb layer was fully relaxed after 5nm of growth.

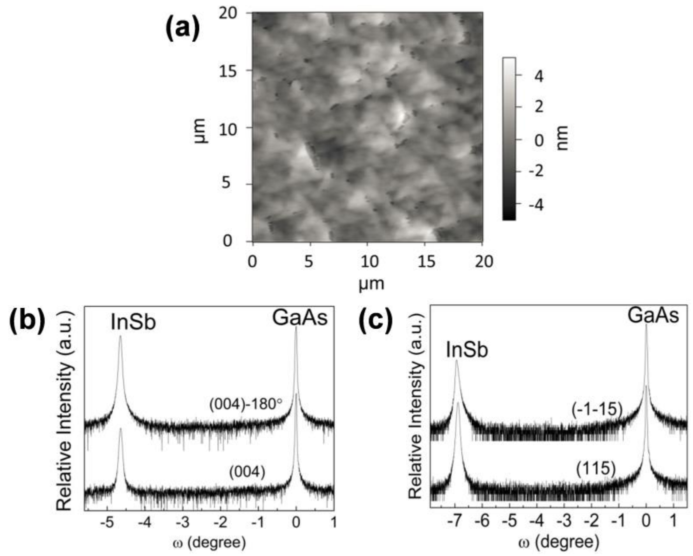

Fig. 3. (a) A 20 μm x 20 μm AFM micrograph of the InSb layer grown on a GaAs (100) substrate. (b) Two (004) ω-2θ scans with the opposite azimuth of 0^{o} and 180^{o}. The peak position of the GaAs peak is normalized to 0^{o}. (c) (115) and ($\bar{1}\bar{1}5$) ω-2θ scans of InSb grown on a GaAs substrate. The position of the GaAs peak is normalized to 0^{o}.

The surface topography of the InSb layer measured with AFM is shown in Fig. 3(a). The InSb layer exhibited a smooth surface with a r.m.s. roughness of 1.1 nm over a 20 μm×20 μm scan area. Some cavities were observed in the AFM micrograph. This could be due to the presence of TDs in the InSb layer. The value of surface roughness reported in this study is smaller compared to that of InSb grown using the two-step method [8], which is 3-7 nm, and that of direct growth InSb, which is ~20 nm [13]. To measure the strain relaxation of the InSb layer, the results of the ω-2θ scans at the (004), (115) and ($\bar{1}\bar{1}5$) planes are shown in Fig 3(b) and 3(c) with ω value of the GaAs peaks were normalized to zero. High interfacial misfit dislocation density may lead to a crystal orientation tilt in the layer. (004) ω-2θ curves at opposing azimuth angles will exhibit a different peak separation between InSb and GaAs peaks if a tilt is presented in the layer. As shown in Fig. 3(b), the peak separations are identical in the (004) ω-*2θ* curves at the azimuth angles of 0^{o} and 180^{o}. This indicates the absence of a tilt in the InSb layer. Using the peak separation between the InSb peak and the GaAs peak in (004), (115) and ($\bar{1}\bar{1}5$) ω-*2θ* curves, the perpendicular and parallel lattice constant is estimated to be 6.4785 Å and 6.4699 Å, respectively. The degree of relaxation of the layer can be calculated

$$\mathrm{R} = \frac{a_{f/\!/} - a_0}{a_{f0} - a_0} \times 100\% \qquad (2)$$

where a_0 is the lattice constant of the substrate, a_{f0} is the lattice constant of the bulk layer's material and a_f is the measured parallel lattice constant of the layer. Using this equation, the calculated relaxation of InSb is 98.9%. The InSb layer exhibited an electron mobility and n-type carrier concentration of 33840 cm^2/V s and 2.3×10^{16} cm^{-3}, respectively in a 300K Hall measurement. This value of electron mobility is comparable to that of an InSb layer grown using the direct growth method and the two-step methods [19, 20].

4. Conclusion

In summary, the growth of a 730 nm-thick InSb layer on a GaAs substrate using IMF dislocations has been first demonstrated. The InSb layer exhibited a smooth surface with a r.m.s. roughness of 1.1 nm, nearly 100% strain relaxation and a room temperature electron mobility of 33840 cm^2/V. Uniformly distributed self-assembly 90° misfit dislocations at the InSb/GaAs interface was observed in a high-resolution TEM measurement. The observed value of the Burgers vector and the separation distance between two adjacent dislocations agrees well with theoretical calculated values.

Acknowledgements

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