High-Quality 100 nm Thick InSb Films Grown on GaAs(001) Substrates with an $\text{In}_x\text{Al}_{1-x}\text{Sb}$ Continuously Graded Buffer Layer

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ABSTRACT: In this paper, we report the growth of a high-quality 100 nm thick InSb layer on a (001) GaAs substrate for InSb-based high-speed electronic device applications. A continuously graded buffer (CGB) technique with $\text{In}_x\text{Al}_{1-x}\text{Sb}$ was used to grow high-quality InSb films on GaAs substrates. The CGB layer was grown by continuously changing the growth temperature and composition of the aluminum and indium during the growth of the buffer layer. Degradation of electrical properties, which normally accompany carrier-defect scattering in a heteroepitaxial layer, was minimized by using the CGB layer. The electrical properties of the InSb films were characterized by Hall measurements, and the electron mobility of the 100 nm-thick InSb film had the largest value, of 39 290 cm$^2$/V·s, among reports of similar thickness. To investigate the relationship between electrical and structural properties, the 100 nm thick InSb film was characterized by energy-dispersive spectroscopy and transmission electron microscopy.

1. INTRODUCTION

Indium antimonide (InSb), with its narrow band gap of 0.17 eV and ultrahigh electron mobility exceeding 80 000 cm$^2$/V·s, has attracted considerable attention for various applications. These exceptional properties allow InSb to be used in high-performance electronic and optoelectronic devices such as magnetic sensors,1,2 micro-Hall sensors,3 long-wavelength photodetectors,4−6 and electronic logic devices.7,8 Unfortunately, the homo-epitaxy of InSb on an InSb substrate is limited by the high cost of the substrate, compared with conventional semiconductor materials such as gallium arsenide (GaAs) or Si, and its high conductivity, which is undesirable for device fabrication which require an insulating substrate. For these economic and physical reasons, InSb films are generally grown on semi-insulating (SI) GaAs substrates. However, the physical properties of InSb (lattice-constant = 6.48 Å) are degraded by defects when grown on GaAs substrates because of a 14% lattice mismatch between the InSb and GaAs.

This lattice mismatch leads to serious defects, such as threading dislocations (TDs), micro-twins (MTs), and cracks located around the interface. However, the TDs, which cause electron-defect scattering at the surface, are reduced as the thickness of the InSb film increases.9 As a result, improved electron mobility is exhibited in thicker InSb films. Unfortunately, a thin layer is preferable for high-speed electronic devices because switching between on and off states is difficult with a thick layer. Thus, reducing defect density in thin-layer InSb on GaAs is crucial for real applications.

Defects due to lattice mismatch can be minimized by using certain growth techniques. For example, a metamorphic-buffer technique has been developed on commercially available substrates such as GaAs, and Si.10,11 According to several reported studies on the growth of high-quality InSb on GaAs substrates, the lattice mismatch problem can be partially solved by using a step-graded buffer (SGB) of AlSb and InAlSb.12−20 It is also important to provide optimized growth conditions, such as growth temperature and growth rate, for the SGB of the AlSb and InAlSb.21−20 It is also important to provide optimized growth conditions, such as growth temperature and growth rate, for the SGB of the AlSb and InAlSb.21−20 For instance, Sato et al. grew a 2 μm thick InSb active structure on GaAs substrates using an AlSb and In$_{1-x}$Al$_x$Sb SGB technique, while Biefeld and Phillips used an InAlSb buffer layer.21,22 In addition, Mishima et al. used an...

Received: August 29, 2018
Accepted: October 22, 2018
Published: November 1, 2018
In$_{1-x}$Al$_x$Sb/In$_{1-x}$Al$_x$Sb short-period super-lattice layer to reduce dislocations by changing the composition of In and Al. As a result, they reported a decrease in defects between the InSb/In$_{1-x}$Al$_x$Sb and In$_{1-x}$Al$_x$Sb/In$_{1-x}$Al$_x$Sb interfaces and have proposed analytical modeling for In$_{1-x}$Al$_x$Sb buffer layers based on structural analysis.

In this work, we apply an In$_{1-x}$Al$_x$Sb continuously GB (CGB) technique to grow a 100 nm thick InSb film. The technique effectively minimized the defects associated with lattice mismatch between the GaAs and InSb. The composition of In and Al in the In$_{1-x}$Al$_x$Sb buffer layers was manipulated continuously by controlling the substrate temperature and the In and Al cell temperatures. The InSb thin film grown with the In$_{1-x}$Al$_x$Sb CGB technique exhibited the highest-electron mobility among reported InSb films of equivalent thickness.

2. EXPERIMENTAL PROCEDURE

Four InSb thin-film samples and an InAlSb terminated sample were grown using the In$_{1-x}$Al$_x$Sb CGB technique with varying Al compositions. The samples were grown on SI GaAs(001) substrates in a Riber compact 21E solid source molecular beam epitaxy system.

The native surface oxide of the SI-GaAs(001) substrates was removed by heating the sample to 620 °C under an As$_2$ atmosphere using a valved cracker cell. Subsequently, a 0.1 μm thick GaAs buffer layer was grown at 580 °C to form a smooth surface, while a (2 × 4) reflection high-energy electron diffraction (RHEED) pattern was maintained. The growth atmosphere was changed from As$_2$ to Sb$_2$, while decreasing the temperature of the substrate ($T_{\text{sub}}$) to 450 °C. The RHEED pattern of the substrate changed from a (2 × 4) pattern to a (1 × 3) pattern under Sb$_2$ atmosphere, which was maintained until the opening of the Al shutter.

Sb was supplied by a Veeco Sb-valved-cracker cell. The beam equivalent pressure of Sb$_2$ was 1.7 × 10$^{-7}$ Torr. An In$_{1-x}$Al$_x$Sb CGB buffer layer was grown on GaAs substrates in a two-phase growth procedure. Figure 1a shows the scheme of the entire growth process in this work, including the continuous changes in the temperature of the cells, growth time, expected lattice constants, and the composition of In and Al.

In the first phase, the In composition was continuously increased in the CGB layer for an hour. The temperature of the Al cell was kept constant at 1130 °C, equivalent to a 1.45 Å/s AlSb growth rate, while $T_{\text{sub}}$ was maintained at 450 °C. Meanwhile, the temperature of the In cell was increased from 600 to 755 °C, which is the In growth rate equivalent to 0 and 3.80 Å/s of InSb, respectively. The In composition in the In$_{1-x}$Al$_x$Sb was gradually raised from 0 to 70% as the In cell temperature increased, and a gradual change in the lattice constant of the InAlSb CGB layer was expected. All samples were prepared using the same growth process in the first phase.

In the second phase, the Al composition of the In$_{1-x}$Al$_x$Sb CGB was decreased with a reduction in growth temperature. While the In cell was maintained at a constant temperature of 755 °C, which corresponds to a growth rate of 3.80 Å/s, the temperature of the Al cell was linearly decreased from 1130 °C to the targeted temperatures ($T_{\text{Al}}$), as $T_{\text{sub}}$ decreased from 450 to 400 °C. The total time for CGB and InSb growth was kept constant at 1 h. Various $T_{\text{Al}}$ of 850, 875, 900, 925, and 950 °C were used for five different samples.

Figure 1b shows the Al flux variation for each of the $T_{\text{Al}}$ (850, 875, 900, and 925 °C) as a function of time. The horizontal dash-dot line indicates zero flux of Al corresponding to an Al cell temperature of 940 °C. Al flux is nearly ignorable below the cell temperature of 940 °C. This was confirmed by energy-dispersive spectroscopy (EDS) measurements, which indicated the growth of pure InSb below the cell temperature of 940 °C. On the basis of the results in Figure 1b, the expected growth time of the InSb film for each sample was 1,080, 852, 544, and 264 s. The growth of the fifth sample with the $T_{\text{Al}}$ of 950 °C was terminated with an InAlSb alloy.

Figure 1c shows the calculated thicknesses of the InSb film grown with each sample. For $T_{\text{Al}}$ of 850, 875, 900, and 925 °C, thickness was 410, 324, 207, and 100 nm, respectively. This thickness calculation is in agreement with our previous study. Using transmission electron microscopy (TEM) and secondary ion mass spectroscopy, which has a depth resolution of about 1 nm, the thickness of an InSb layer grown for 1080 s was confirmed to be 0.41 μm. Although the Al cell temperature linearly decreased, the Al flux variation was inversely exponential to a function of time. This implies that the Al growth rate decreased exponentially as a function of time.

Electron mobility and doping concentration were characterized using an Ecopia HMS-3000 Hall measurement system. It was implemented with a standard van der Pauw 4-probe

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InSb, 6.48 Å
In$_{0.7}$Al$_{0.3}$Sb
In$_{1-x}$Al$_x$Sb (x:0.7 → 1)
In$_x$Al$_{1-x}$Sb (x:0 → 0.7)
AlSb, 6.14 Å

Figure 1. (a) Schematics of the growth method using the In$_{1-x}$Al$_x$Sb metamorphic continuous grading buffer (CGB) technique, with changing In and Al flux. The scale is not real. (b) Time dependence of Al fluxes measured by simulating the 2nd phase operation. Below a $T_{\text{Al}}$ of 940 °C, no flux was measured from the Al cell. (c) Calculated InSb thickness based on the results of (b).
method on a sample area of (5 mm × 5 mm) at room temperature.

The microstructure analysis was carried out using a 200 kV TECNAI F20 G2 transmission electron microscope (FEI, Hillsboro, OR) with the sample mounted in a double-tilt holder (Gatan, Pleasanton, CA). TEM images were obtained with an UltraScan 1000 (2k × 2k) CCD camera (Gatan Inc.) and a Fischione model 3000 ADF detector. EDS mapping and line profile were obtained with a Talos F200X transmission electron microscope with a Bruker Super EDS System.

3. RESULTS AND DISCUSSION

Figure 2a shows the electrical properties of the InSb film samples for various film thicknesses, characterized by a Hall measurement system at room temperature. For 100, 207, 324, and 410 nm thick InSb films, electron mobility was 39 290, 38 920, 41 740, and 46 300 (cm²/V·s) and electron carrier density was 1.7, 1.9, 2.8, and 2.5 (×10¹⁵/cm³), respectively. Meanwhile, the electron mobility of the InAlSb terminated sample grown with a T substrates of 39 290 (cm²/V·s) at room temperature in the first phase. Previous reports of mobility of InSb on GaAs with a thickness of 100 nm were limited to 1 400 (cm²/V·s). Comparisons with other thicknesses show that there was also a considerable amount of improvement in electron mobility over other reported values. These enhancements are attributed to the reduction in defects in the InSb film by separating the InSb film from defects near the GaAs interface, using the In_xAl_{1−x}Sb CGB technique.

Figure 2b compares the electron mobility of InSb films in this work to other reported results. The electron mobility of the 100 nm thick InSb film in this work was 39 290 (cm²/V·s), while previous reports of mobility of InSb on GaAs with a thickness of 100 nm were limited to 1 400 (cm²/V·s). Comparisons with other thicknesses show that there was also a considerable amount of improvement in electron mobility over other reported values. These enhancements are attributed to the reduction in defects in the InSb film by separating the InSb film from defects near the GaAs interface, using the In_xAl_{1−x}Sb CGB technique.

Figure 3a shows a dark field TEM image of the 100 nm thick InSb film with horizontal yellow-dashed dot lines, which are designated as points A and B. They will be explained in Figure 3b. The distance between the GaAs substrate and points A and B is 1.22 and 0.3 μm, respectively. The thickness of the In_{x}Al_{1−x}Sb CGB layer including the 100 nm-thick InSb film is 2.33 μm. Fringes appear because of inevitable distortion produced by the focused ion beam (FIB) milling process. Defects such as TDs and MTs are distinctly shown near the interface between the GaAs and InAlSb CGB, resulting from stress caused by lattice mismatch between the GaAs and (In_x)Al_{1−x}Sb. The defects are mostly located in an area between the interface and point A. However, the nucleation and glide of dislocation is reduced from point A to the InSb.

Figure 3b shows the distribution of elements, including Ga, In, Al, As, and Sb, from the GaAs substrate to the InSb film surface. The growth rate was assumed to be 0 Å/s at the start of first phase because the measured In flux was almost zero at the cell temperature of 600 °C. In elements, however, were detected at the interface of the GaAs/In_{x}Al_{1−x}Sb. This is mainly attributed to overshoot of the In flux from the In source during the sudden opening of the shutter, and partially to the intermixing of In during growth.

The In composition gradually increased from the substrate interface to the InSb film surface. Interestingly, the gradual trend in composition changes at point A. The slopes of In intensity and Al intensity as a function of length become smaller across point A. Here, we can conclude that point A is related to a change in the first and second phases during growth (see Figure 1a). The trend is attributed to the change in slope of the composition, resulting from the different slopes in the change of In and Al fluxes in the first and second phases. That is, in the first phase, only the In cell was gradually heated up. Meanwhile, in the second phase, only the Al cell was gradually cooled down. The slope of each cell temperature was different.

The EDS results confirm that Al atoms do not exist near the surface. The thickness of the InSb as measured from the EDS is approximately 104 nm. This is in quite good agreement with the thickness of the InSb calculated from the Al flux (see Figures 1c and 2b). The EDS data once again confirms that the growth of the InSb film was pure without any Al near the surface.

Here, we should point out that the T Al decreases beyond point A, which is the point where the second phase of the CGB layer start. Most of the defects are limited to the region between the interface and point A. This implies that the generation and glide of dislocations during the in-plane strain relaxation is suppressed. This suppression is attributed to an increase in the yield strength and limitation of the dislocation thread and glide, resulting from the decrease in growth temperature in the second phase.
slope of the composition may contribute to the reduction in defects above point A because of the formation of a virtual interface which reflects defects.27

To analyze the relationship between dislocations and strains of the InAlSb CGB layer, the lattice constants were compared along the growth direction by the constants from selected area electron diffraction (SAED) patterns. The lattice constant ratio along (110) ($a_{xy}$) over the one along (001) ($a_{zz}$) in the nondeformed zinc-blende crystal structure of III–V materials is $\sqrt{2}$. The lattice constant ratio ($R$) is calculated by dividing $a_{zz}$ with $a_{xy}$ from

$$R = \frac{a_{zz}}{a_{xy}}$$

If the $R$ is $\sqrt{2}$, the crystal structure at the point is cubic and if the $R$ is smaller or larger than $\sqrt{2}$, the cubic structures at the point are compressively and tensely deformed, respectively.

Figure 3c shows a calculated $a_{zz}$ lattice constant from the SAED image at the points, the SAED pattern along the growth direction of the sample and $R$. The dashed blue arrow indicates $\sqrt{2}$ which means that the crystal structure is cubic. The SAED image indicates that there is high-quality single crystallinity throughout the whole sample. The patterns mean a typical zinc blended structure. The calculated $a_{zz}$ lattice constants gradually increase from the GaAs surface to the InSb film surface.

$R$ is slightly increased to 1.41 at a point approximately 170 nm from the interface, while the $R$ of the AlSb layer at the interface with the GaAs substrate is about 1.403. The variation in the $R$ value means the crystal structure of the (In)AlSb GB layer at the surface of the GaAs substrate is deformed by the in-plane strain due to lattice mismatch.

Meanwhile the deformed structure changes to a cubic form because of relaxation of the in-plane strain following the generation and gliding of dislocations up to point B, which is about 170 nm from the GaAs substrate. The strain relaxation produced by dislocations is also shown in Figure 3a. The dislocations are reduced beyond point B. This is attributed to the low temperature growth of the (In)AlSb CGB buffer layer in the first phase with a $T_{sub}$ of 450 °C. Subsequently, the $R$ value gradually decreases as the GB layer is grown on the InSb surface. This means that deformation of the crystal structure is favorable, to relieve in-plane strain, rather than the generation and gliding of dislocations.

Figure 3d illustrates the lattice mismatch between the GaAs substrate and various points along the growth direction in the InAlSb CGB layer for $a_{xy}$ and $a_{zz}$. The lattice constants calculated from each of the SAED patterns. The difference in lattice mismatch $a_{xy}$ and $a_{zz}$ is minimized at a point ∼170 nm from the GaAs substrate because the in-plane strain is relieved by the generation and gliding of dislocations, as mentioned earlier. The difference increases beyond point B to near the InSb film surface.

Considering the dark field TEM image, $R$, the lattice mismatch, the in-plane strain is relieved by the generation and gliding of dislocations in the first phase of the growth below point A. This agrees with the dark field TEM image in Figure 3a where dislocations are mostly observed in the lower region of the CGB layer below point A. On the other hand, the in-plane strain is slightly increased with suppression of dislocations in the second phase of growth above point A, which is attributed to the decrease in $T_{sub}$.28

Figure 3. (a) TEM dark field image of the 100 nm thick InSb film on GaAs substrate with a 500 nm scale bar. The horizontal yellow-dashed-dot A and B lines indicate points 1.22 and 0.3 $\mu$m from the GaAs substrate. (b) EDS line profile of the sample along the white dashed-arrow in Figure 2a, and EDS 2D mapping image of In and Al with a 500 nm scale bar. (c) Calculated lattice constants and $R$ from SAED patterns at the points. (d) Lattice mismatch along $a_{xy}$ and $a_{zz}$ at a point near the GaAs lattice constant.
4. CONCLUSIONS

The In$_{Al_{1-x}}$Sb CGB technique, combined with control of the Al, In cell, and growth temperature, is an effective method for overcoming the problem of lattice mismatch between InSb and GaAs. Using the In$_{Al_{1-x}}$Sb CGB technique, an InSb film with a thickness of 100 nm was grown, which exhibits an electron mobility of 39.290 cm$^2$/Vs. This is a noticeable improvement over other reports with similar thickness. Reducing the thickness of InSb films while preserving their properties is crucial to applications in high-performance devices. The In$_{Al_{1-x}}$Sb CGB technique offers a potential method to obtain high-quality III–V films by heteroepitaxy for real applications.

ACKNOWLEDGMENTS

This work was mainly supported by KIST institutional program (2E26420). We express our thanks to Xuan, Advanced Analysis Center, KIST, for FIB.

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