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Low-resistance p-type polycrystalline GaSb grown by molecular beam epitaxy

Yingda Dong*, Dennis W. Scott, Yun Wei, Arthur C. Gossard, Mark J. Rodwell

Department of Electrical and Computer Engineering, University of California, Santa Barbara, CA 93117, USA

Abstract

We report the growth and electrical properties of low resistance carbon-doped polycrystalline GaSb (poly-GaSb) by molecular beam epitaxy using CBr_4. The resistivity of poly-GaSb is found to have strong dependence on film thickness and grain size of the polycrystalline film, particularly when the film thickness is comparable with the grain size. It is observed that grain size is mainly determined by growth temperature, while hole concentration is significantly affected by antimony to gallium beam flux ratio. With same doping level, grain size, and similar film thickness, the resistivity of carbon-doped poly-GaSb is more than one order of magnitude lower than that of carbon-doped poly-GaAs. This result is attributed to GaSb's favorable surface Fermi-level pinning in the valence band and higher hole mobility. It is proposed that carbon-doped poly-GaSb be used as extrinsic base material in InP heterojunction bipolar transistor to improve the device's high-frequency performance.
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1. Introduction

Polycrystalline silicon is an important material in the fabrication of advanced silicon bipolar transistors (BJTs) and metal-oxide-semiconductor field-effect transistors (MOSFETs). In SiGe BJTs, growth of a polysilicon extrinsic base over a dielectric spacer reduces the capacitance between the base and the collector (C_{BC}) [1,2]. In III–V heterojunction bipolar transistors (HBTs), the same technique can be applied, with low resistivity polycrystalline material employed as the extrinsic base layer of an HBT with buried SiO_2/SiN surrounding a collector pedestal, as shown in Fig. 1. With this structure, C_{BC} and R_B can be reduced simultaneously and both ECL logic speed and maximum oscillation frequency (f_{max}) will be significantly improved.

Applications of polycrystalline GaAs (poly-GaAs) in GaAs-based HBTs are however not as successful as those of polycrystalline silicon in the SiGe BJT technology. One reason could be

*Corresponding author. Tel.: +18058933812; fax: +18058938714.

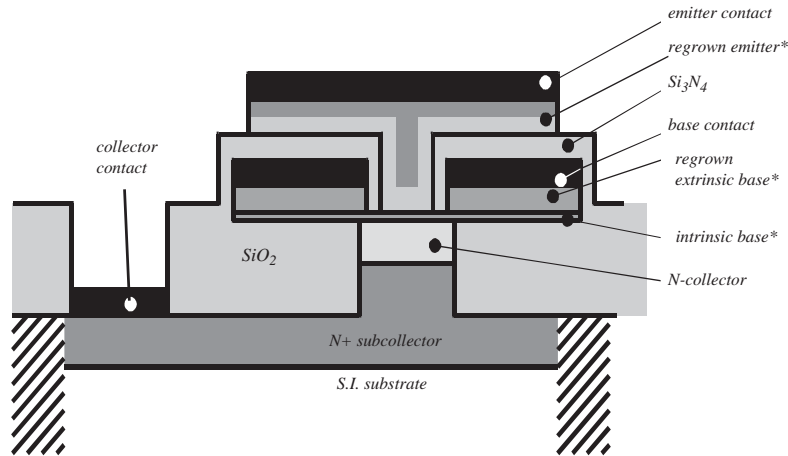*E-mail address:* yingda@ece.ucsb.edu (Y. Dong).

Fig. 1. Proposed InP HBT structure with regrown base and emitter.

ascribed to the difficulty of achieving low resistance poly-GaAs. Some efforts have been made to investigate the electric properties of poly-GaAs grown by different methods and a considerably wide range of resistivity values have been reported [3–5]. The lowest resistivity ever reported for poly-GaAs ($\sim 3 \times 10^{-3}\,\Omega\,cm$) [5], however, is still too high to achieve low sheet resistance for a relatively thin extrinsic base layer, which is necessary to the sub-micron double–poly HBT fabrication process.

In our study, p-type GaSb has been chosen as a potential candidate for the polycrystalline extrinsic base material for InP-based HBT. Besides GaSb's low-energy bandgap and therefore high carrier mobility, another advantage is attributed to p-type GaSb's favorable surface Fermi-level pinning position. It has been widely known that polycrystalline materials are composed of small crystallites joined together at grain boundaries. At the grain boundaries between the crystallites, there exist a substantial number of traps that pin the surface Fermi level and greatly influence the bulk conductivity [3,6]. It has been found that p-type GaSb's surface Fermi-level is pinned in the valence band [7]. As Fig. 2 shows, this Fermi level pinning causes a band-bending barrier on the grain boundaries of p-type poly-GaSb much lower than that on the grain boundaries of p-type poly-GaAs, which has the surface Fermi-level pinned in the mid-bandgap [8]. Similarly, the grain boundary barrier of p-type poly-GaSb will also be lower than
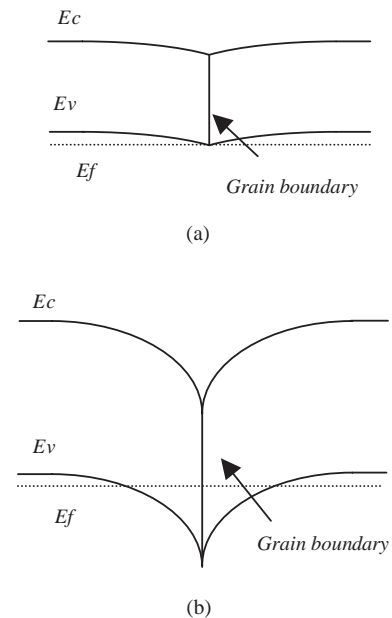

Fig. 2. Schematic diagrams of suggested energy band structure near grain boundaries in p-type of (a) GaSb and (b) GaAs.

that of p-type poly-InAs, which has the surface Fermi level pinned above the conduction band [9]. For these reasons, the conduction between grains in poly-GaSb is much less limited and lower bulk resistivity is more likely to be achieved.

In this work, we study the growth and electrical properties of low resistance carbon-doped polycrystalline GaSb (poly-GaSb) by molecular beam

epitaxy using CBr_4. We examine several aspects of growth conditions and film properties that significantly affect poly-GaSb film's resistivity. It is found that the resistivity of poly-GaSb film depends strongly on film's thickness and grain size, particularly when the film thickness is comparable with the grain size. We demonstrate that with similar doping level, grain size, and film thickness, the resistivity of poly-GaSb is more than one order of magnitude lower than that of poly-GaAs.

2. Experimental procedures

Poly-GaSb samples were grown in a Varian Gen II system equipped with a valved and cracked Sb source. High purity carbon tetrabromide (CBr_4) was delivered into the growth chamber through an ultra-high vacuum leak valve. A $0.3 \, \mu m$ thick SiO_2 film was deposited on a semi-insulating (1 0 0) GaAs substrate by plasma-enhanced chemical vapor deposition (PECVD) and the sample was then cleaned prior to being loaded into the MBE chamber. The growth rate was $0.2 \, \mu m/h$ and the CBr_4 flux was fixed with the same leak valve setting.

Hall measurements were performed at 295 K using the van der Pauw geometry at a magnetic field of 4000 G. Indium metal contacts were formed at 200–250°C. Hall measurements were performed on as-grown samples without any additional annealing.

3. Results and discussions

Three 1000 Å poly-GaSb samples were grown at different substrate temperatures with fixed antimony to gallium beam flux ratio of 5. Fig. 3 shows the dependence of poly-GaSb samples' hole concentration and mobility on the growth temperature. It is clear that the growth temperature strongly influenced the hole mobility, but only weakly affected the hole concentration. This behavior could be related to the grain size's growth temperature dependence, which is shown in Fig. 4 together with the resistivity of the same poly-GaSb films. Fig. 5 displays the SEM images of poly-GaSb grown at (a) 520°C and (b) 475°C at a magnification of 35,000. Clearly the grain size becomes much smaller at a lower growth temperature, decreasing from 250 to 80 nm. If the potential
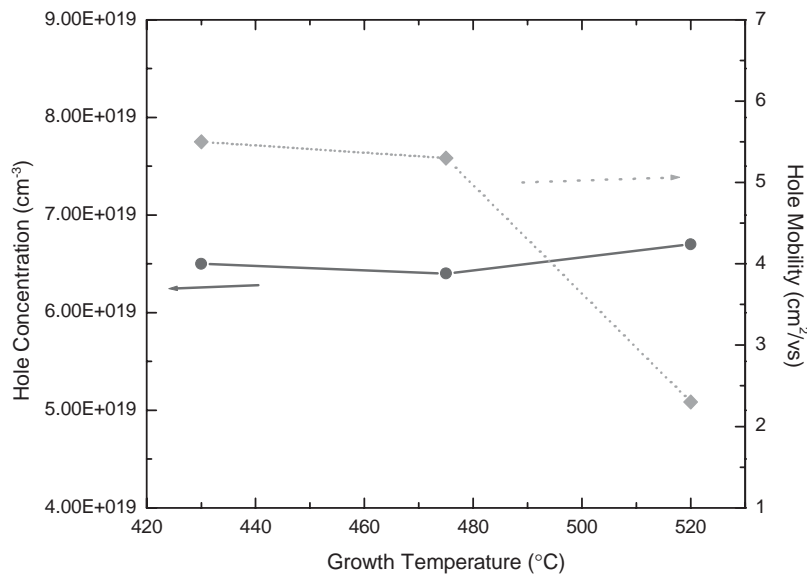

Fig. 3. The dependence of hole concentration and mobility of poly-GaSb samples on growth temperature.
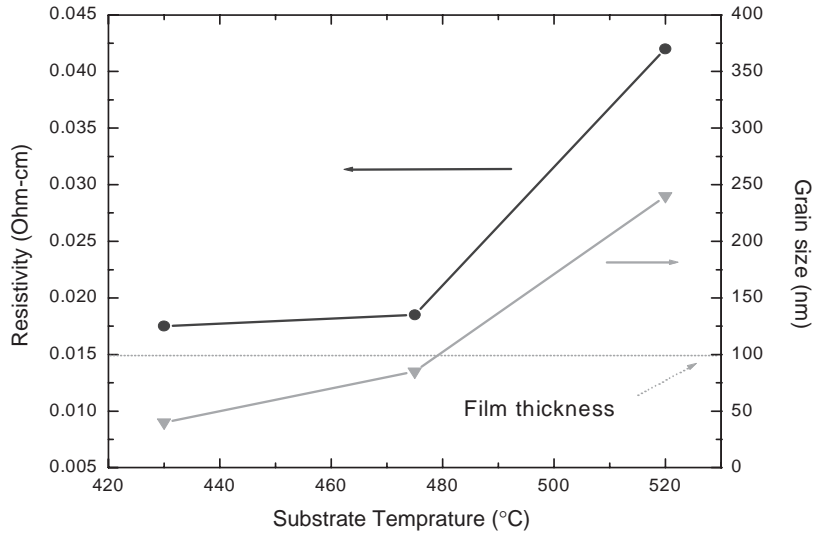
Fig. 4. The dependence of resistivity and grain size of poly-GaSb samples on substrate temperature.
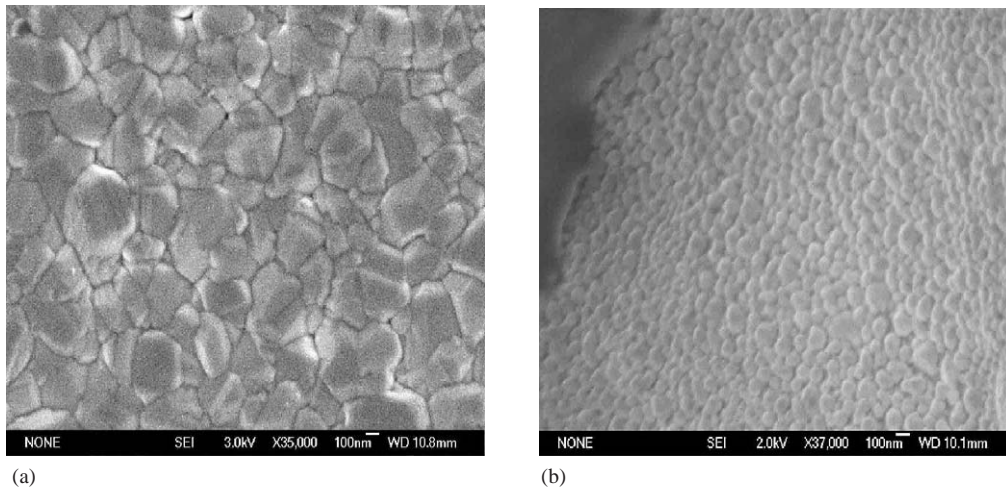

(a)

(b)

Fig. 5. SEM images of poly-GaSb samples grown at (a) 520°C and (b) 475°C.

energy barrier on the grain boundaries impedes the motion of carriers from one crystallite to another and thereby decreases the bulk conductivity hence the apparent bulk hole mobility, it is surprising to find that the hole mobility increases as the grain size decreases, and the number of associated grain boundaries increases. Note, however, that with smaller grain sizes, the total boundary area connecting the crystallites may become larger, which can significantly improve the carrier con-

duction. This may explain the increasing trend of hole mobility with the decrease of grain size.

From Fig. 4, it can be found that the resistivity decreased rapidly with the substrate temperature dropping from 520°C to 475°C but remained relatively constant at lower temperature even though the grain size dropped further. Also from Fig. 4, it can be seen that for the poly-GaSb film grown at 520°C, the film thickness is smaller than the average grain size, while for the films grown at

475°C and 430°C, the film thicknesses are larger than the average grain sizes. It is reasonable to assume that when the film thickness is smaller than the grain size, the grain boundary areas are substantially reduced, severely limiting the conduction between polycrystalline crystallites. Under this circumstance, the resistivity of the polycrystalline film could become very sensitive to the grain size. This may explain why the resistivity decreased rapidly when the growth temperature was lowered from 520°C to 475°C but remained little changed at lower temperature.

The second set of poly-GaSb samples were grown with varied antimony to gallium beam flux ratios at the same growth temperature of 440°C. Fig. 6 shows the hole concentration and mobility of these samples as a function of the V/III ratio. The hole concentration shows a clear increasing trend for decreasing V/III ratio. This behavior could be explained by considering that carbon must displace Sb from its lattice site in order to behave as an effective p-type dopant. With lower Sb beam pressure, the chance of carbon being incorporated into the Sb sublattice will be larger and the hole concentration will therefore be increased. In this set of samples it was found that

the change of V/III flux ratio did not significantly affect the hole mobility. The dependence of the grain size on the V/III ratio was also weak.

The third set of poly-GaSb samples were grown with 1000, 1500, 2000 and 3000 Å film thicknesses, all at 440°C. Fig. 7 shows the resistivity of these four samples as a function of film thickness. The hole concentration, mobility, and sheet resistivity are listed in Table 1. It is clear that unlike monocrystalline material, the resistivity of poly-GaSb has a strong dependence on film thickness, with thicker films having lower resistivity. Consequently, the sheet resistivity is no longer proportional to the reciprocal of the film thickness; instead it increases very rapidly with decreasing film thickness. These results suggest that resistivity should be compared only between polycrystalline films similar thickness.

Since these samples were all grown at 440°C and had very similar average grain sizes, it is clear from the data of Table 1 that the hole mobility decreases as the ratio of film thickness to grain size is decreased. This is obviously consistent with the results obtained from Fig. 3.

The resistivity of the 3000 Å poly-GaSb film is $7.5 \times 10^{-3} \, \Omega \, \text{cm}$, over one order of magnitude
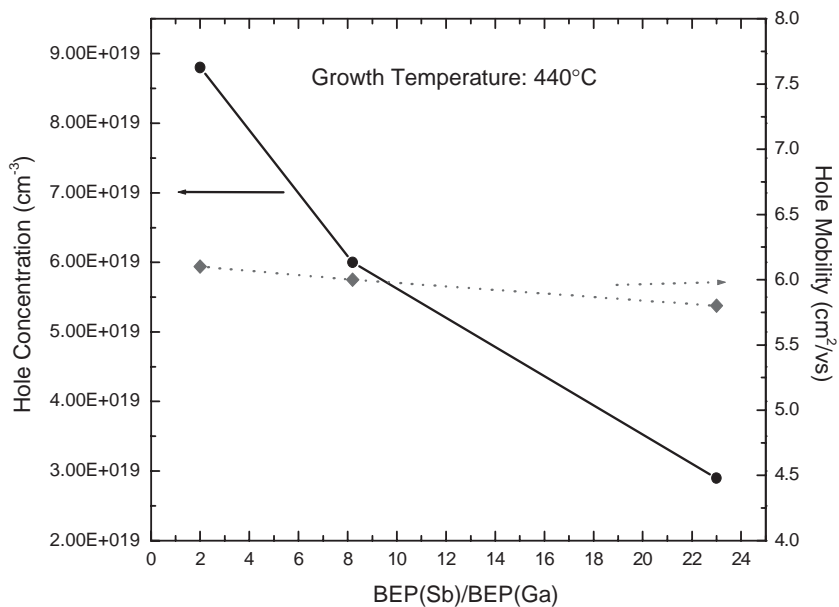

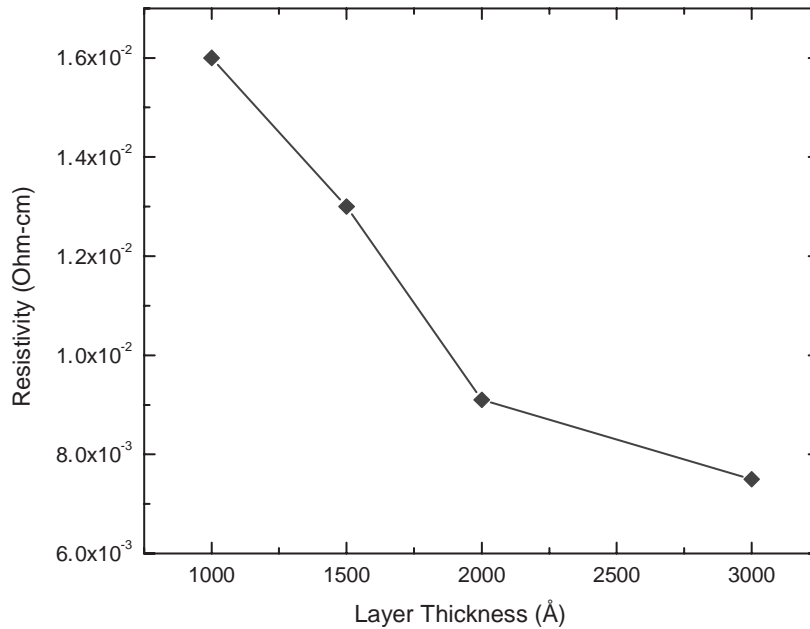Fig. 6. The hole concentration and mobility as a function of V/III beam flux ratio.

Fig. 7. The dependence of poly-GaSb resistivity on film thickness.

Table 1
Electric properties of poly-GaSb samples with different film thicknesses

Poly GaSb thickness (Å)	Hole density N_s (cm^{-3})	Mobility μ (cm^2/V s)	Bulk resistivity ρ (Ω cm)	Sheet resistivity ρ_s ($\Omega/\square$)
3000	8.2e19	10.2	7.5e-3	240
2000	8.0e19	8.6	9.1e-3	450
1500	8.1e19	5.8	1.3e-2	900
1000	7.8e19	5.1	1.6e-2	1550

lower than that of polycrystalline GaAs with the same carbon doping level in a thicker 4000 Å film [6]. The resistivity of poly-GaSb is expected to decrease further by using higher CBr_4 flux during growth to achieve higher carbon doping density, which is believed to have much room to increase beyond the current value of 8×10^{19} cm^{-3} used in our experiments.

4. Conclusions

Low resistance poly-GaSb films can be achieved by MBE growth using CBr_4 doping. The resistivity of poly-GaSb has strong dependence on film's thickness and grain size, particularly when the film thickness is comparable with the grain size. It has been found that growth temperature strongly influence the grain size, but has little effect on hole concentration. The hole concentration, on the other hand, is significantly affected by the antimony to gallium beam flux ratio, with low V/III ratios increasing the carbon incorporation. Therefore, low growth temperature and low V/III beam flux ratio are desirable to achieve low-resistivity poly-GaSb films. With the same doping level (8×10^{19} cm^{-3}), grain size (~ 100 nm) and similar film thickness, the resistivity of poly-GaSb is found to be more than one order of magnitude lower than that of poly-GaAs. This low resistivity is attributed to GaSb's favorable surface Fermi-level pinning position and higher hole mobility.

These results suggest that poly-GaSb is a good candidate for the use as an extrinsic base material in InP HBTs.

References

[1] T. Nakamura, T. Miyazki, S. Takahasji, T. Kure, T. Okabe, M. Nagata, IEEE Trans. Electron Devices ED-29 (1982) 596.

[2] T.H. Ning, R. Issac, P. Solomon, D.D. Tang, H. Yu, G.C. Feth, S. Wiedmann, IEEE Trans. Electron Devices ED-28 (1981) 1010.

[3] J.J.J. Yang, P.D. Dapkus, R.D. Dupuis, R.D. Yingling, J. Appl. Phys. 51 (1980) 3794.

[4] K. Mochizuki, T. Nakamura, T. Tanoue, H. Masuda, Solid-State Electron. 9 (1995) 1619.

[5] N.Y. Li, C.W. Tu, J. Crystal Growth 188 (1998) 45.

[6] J.YW. Seto, J. Appl. Phys. 12 (1975) 5247.

[7] W.E. Spicer, P.W. Chye, P.R. Skeath, C.Y. Su, K. Lindau, J. Vac. Sci. Technol. 16 (1979) 1422.

[8] X. Yin, H-M. Chen, F.H. Pollak, Y. Chan, P.A. Montano, P.D. Kirchner, G.D. Pettit, J.M. Woodall, J. Vac. Sci. Technol. A1 (1992) 131.

[9] C. Nguyen, B. Brar, H. Kroemer, J. Vac. Sci. Technol. B11 (1993) 1706.