Liquid-phase-epitaxy-grown InAs$_x$Sb$_{1-x}$/GaAs for room-temperature 8–12 μm infrared detectors

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High-quality InAs$_x$Sb$_{1-x}$ (0 < x ≤ 0.3) films are grown on GaAs substrates by liquid phase epitaxy and electrical and optical properties of the films are investigated, revealing that the films exhibit Hall mobilities higher than 2 × 10$^4$ cm$^2$ V$^{-1}$ s$^{-1}$ and cutoff wavelengths longer than 10 μm at room temperature (RT). Photoconductors are fabricated from the films, and notable photoresponses beyond 8 μm are observed at RT. In particular, for an InAs$_{0.2}$Sb$_{0.7}$ film, a photoresponse of up to 13 μm with a maximum responsivity of 0.26 V/W is obtained at RT. Hence, the InAs$_x$Sb$_{1-x}$ films demonstrate attractive properties suitable for room-temperature, long-wavelength infrared detectors. © 2006 American Institute of Physics. [DOI: 10.1063/1.2209709]

Long-wavelength (8–12 μm) infrared detectors that can operate at room temperature (RT) have important infrared applications. At present, Hg$_1$−Cd$_x$Te is the most prevalent material in high-performance, long-wavelength infrared detectors. However, Hg$_1$−Cd$_x$Te lacks stability and uniformity over a large area, and only works under cryogenic conditions. InAs$_x$Sb$_{1-x}$ has recently attracted interest as a promising alternative to Hg$_1$−Cd$_x$Te. InAs$_x$Sb$_{1-x}$ is more stable and has higher electron and hole mobilities than Hg$_1$−Cd$_x$Te. When the value of x is in the midrange (0.2 < x < 0.6), InAs$_x$Sb$_{1-x}$ exhibits a large, positive "optical bowing" effect due to the ordering in the alloy, making possible its application in long-wavelength infrared detection. The main disadvantage of InAs$_x$Sb$_{1-x}$ is a lack of lattice-matched substrates. Despite its large lattice mismatch (7.2% < Δa/a < 14.6%) with InAs$_x$Sb$_{1-x}$, GaAs is an excellent substrate because of its infrared transparency and low noise. Moreover, it helps integrate detection and signal processing circuits and reduce parasitic capacitances. Therefore, the crucial point is the growth of high-quality InAs$_x$Sb$_{1-x}$ films on GaAs substrates (InAs$_x$Sb$_{1-x}$/GaAs). Epitaxial InAs$_x$Sb$_{1-x}$ films have been grown on GaAs substrates by molecular beam epitaxy (MBE) and metal organic chemical vapor deposition (MOCVD). Furthermore, demonstrations of RT long-wavelength infrared photodiodes and photoconductors using the MOCVD-grown InAs$_x$Sb$_{1-x}$/GaAs have been reported.

As is much more accessible and economical than MBE and MOCVD, liquid phase epitaxy (LPE) is often considered for the growth of InAs$_x$Sb$_{1-x}$ films. However, it is difficult to use LPE to carry out epitaxial growth with a large lattice mismatch (Δa/a > 1%). Popov et al. have grown InAs$_x$Sb$_{1-x}$ films on InAs substrates, indicating that when x < 0.7, the InAs$_x$Sb$_{1-x}$ films grown were polycrystalline (Δa/a > 2.1%). Recently, Dixit et al. have reported successful epitaxial growth of InAs$_x$Sb$_{1-x}$/GaAs by LPE with extremely low ramp-cooling rates (<0.8 °C/h), but they could incorporate only 6 at. % arsenic in InAs$_x$Sb$_{1-x}$ films due to the miscibility gap. This obstructs their InAs$_x$Sb$_{1-x}$/GaAs from being applied above the 9.5 μm wavelength range. In this letter, we report high-quality LPE-grown InAs$_x$Sb$_{1-x}$/GaAs suitable for RT long-wavelength infrared detectors.

The LPE growth was carried out in a conventional horizontal graphite sliding-boat system with an ambient of flowing Pd-membrane purified hydrogen at atmospheric pressure in a quartz reactor tube. The starting materials were 6N pure In and Sb, and undoped InAs (n < 10$^{12}$ cm$^{-3}$). The substrates used were well-polished, semi-insulating (100) GaAs wafers. In this work, samples were grown at constant growth temperatures rather than in ramp-cooling routines. Constant growth temperatures help the uniformity of the InAs$_x$Sb$_{1-x}$ films in the growth direction. The growth temperatures ($T_G$) for the samples are shown in Table I. The thicknesses of the InAs$_x$Sb$_{1-x}$ films range from 10 to 20 μm.

The inset of Fig. 1 shows a cross section image of a typical sample (C) observed under a scanning electron microscope (SEM), indicating a sharp interface between the InAs$_x$Sb$_{1-x}$ film and GaAs substrate. The energy dispersive x-ray analysis (EDAX) system of the SEM was used for compositional analysis of the InAs$_x$Sb$_{1-x}$ films. By linescan of EDAX perpendicular to the surface in the cross section, the distribution of the constituent elements along the depth direction can be obtained. The EDAX linescan result of sample C is shown in Fig. 1, indicating the composition uni-

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formity of the InAs$_{x}$Sb$_{1-x}$ film and confirming the sharpness of the interface. Hence, in our LPE growth, severe diffusion at the interface did not occur even for sample C with a high growth temperature of up to 549 °C, and InAs$_{x}$Sb$_{1-x}$ films with uniform composition can be grown by applying constant growth temperatures. The compositions ($x$) of the samples obtained from EDAX are shown in Table I.

Figure 2 shows a typical x-ray diffraction (XRD) pattern of the samples, indicating that the InAs$_{x}$Sb$_{1-x}$ films are polycrystalline with (100)-preferred orientation. For comparison, the inset shows the predominant (400) reflection peaks from the three samples and an InSb film. The (400) peaks obviously shift towards higher angle with the increase of arsenic mole fraction ($x$). According to Vegard’s law, the compositions of the InAs$_{x}$Sb$_{1-x}$ films can also be calculated from those XRD peaks and the results, shown in Table I, are consistent with those obtained from EDAX, considering that the accuracy of EDAX, is within ±1 at.

The polycrystalline structure of the (InAs)$_{x}$(Sb)$_{1-x}$ films confirms that, as mentioned above, it is quite difficult to grow epitaxial InAs$_{x}$Sb$_{1-x}$/GaAs by LPE. In general, epitaxial structures are preferred over polycrystalline structures because grain boundaries that exist in polycrystalline structures degrade the material quality. For instance, scattering from grain boundaries greatly lowers mobility. However, the results of Hall measurements at RT, shown in Table I, reveal that the InAs$_{x}$Sb$_{1-x}$ films have high Hall mobilities greater than $2 \times 10^4$ cm$^2$ V$^{-1}$ s$^{-1}$ at RT, which are at the same level as those reported for epitaxial InAs$_{x}$Sb$_{1-x}$/GaAs with similar composition grown by MOCVD (Ref. 12) and MBE. This may be explained as follows. A lot of misfit dislocations exist in the epitaxial InAs$_{x}$Sb$_{1-x}$/GaAs grown by MOCVD and MBE due to the large lattice mismatch, and dislocation scattering has been regarded as one of the dominant scattering mechanisms for InAs$_{x}$Sb$_{1-x}$/GaAs. However, for our InAs$_{x}$Sb$_{1-x}$ films, the inside-grain dislocation density could be greatly reduced due to the nearly equilibrium growth conditions in LPE. Moreover, the highly (100)-preferred orientation would diminish the effect of grain boundaries. Hence, it may be counteracted by the reduction of the inside-grain dislocation density.

The Fourier transform infrared (FTIR) transmission spectra of the samples were measured at RT, and using that of a GaAs substrate as reference, the FTIR transmission spectra of the corresponding InAs$_{x}$Sb$_{1-x}$ films at RT were obtained and are shown in Fig. 3. The absorption edges of the InAs$_{x}$Sb$_{1-x}$ films shift toward the long-wavelength side with the increase of $x$, demonstrating a strong positive optical bowing effect. For a direct energy gap semiconductor, the absorption coefficient ($\alpha$) of InAs$_{x}$Sb$_{1-x}$ can be described as $\alpha=A(h\nu-E_g)^{1/2}$. Here $A$ is a constant, $\nu$ is the incident photon frequency, and $E_g$ is the band gap. Therefore, based on the data of $\alpha d$ ($d$ is the film thickness) calculated from the FTIR transmission results, $E_g$ at RT can be obtained from the linear fitting of $(\alpha d)^2$ (as shown in the inset of Fig. 3) and is shown in Table I. The dependence of $E_g$ on the composition of InAs$_{x}$Sb$_{1-x}$ at RT has been given by Woolley and Warner as $E_g(x)=0.35-0.771(1-x)+0.596(1-x)^2$ eV. The $E_g$ calculated from this equation ($E_g(x)$), also shown in Table I, does not agree with our results. The $E_g$ of InAs$_{x}$Sb$_{1-x}$ may be influenced by the degree of ordering and residual strain in the alloy, so the discrepancy may be attributed to the difference between growth processes in producing ordering and residual strain.

Photoconductors formed by 0.1 $\times$ 2 mm$^2$ bars of InAs$_{x}$Sb$_{1-x}$ films were fabricated on samples A and C. Au/Ti was used to form an Ohmic contact. The spectral photoresponse was measured at 77 K and RT using a FTIR spectrometer, and the absolute responsivity was calibrated based on the responsivity measurement at 77 K using a standard blackbody source with a temperature of 800 K. The bias voltage applied to the photoconductors was 1.5 V. Figures 4(a) and 4(b) show the spectral responsivities of the photoconductors.
conductors at 77 K and RT, respectively. A notable photocurrent above 8 μm was observed with both photoconductors at RT, and the cutoff wavelength was extended to about 13 μm. The maximum responsivities \( R_{\text{v, max}} \) and the cutoff photon energies \( E_{\text{c}} \) of the photoconductors are shown in Table II. At RT the responsivities of our photoconductors are higher than those reported for MOCVD-grown InAs\(_{x}\)Sb\(_{1-x}\)/GaAs.\(^\text{13,14}\) This confirms the high quality of our LPE-grown InAs\(_{x}\)Sb\(_{1-x}\)/GaAs. In particular, sample C with a midrange composition exhibited a discernible photoresponse beyond 8 μm at RT, showing great potential for RT long-wavelength infrared detection.

In comparison with the \( E_{\text{c}} \) obtained from the FTIR experiments, \( E_{\text{c}} \) is smaller by about 0.02 eV for both samples. The temperature dependence of \( E_{\text{c}} \) is described by Varshni's equation:\(^\text{20}\) \( E_{\text{c}}(T) = E_{\text{c}}(0) - \alpha T^2/\beta (\beta + T) \). Using the values of \( \alpha \) and \( \beta \) for InAs\(_{x}\)Sb\(_{1-x}\) fitted by Dixit et al.,\(^\text{16}\) \( \alpha = 3.01 \times 10^{-3} \text{ eV K}^{-1} \) and \( \beta = 341 \text{ K} \), the difference between the \( E_{\text{c}} \) at 77 K and RT (300 K) can be calculated to be 0.038 eV. This is in good agreement with the data of \( [E_{\text{c}}(77 \text{ K}) - E_{\text{c}}(300 \text{ K})] \) in Table II, suggesting a tight connection between the \( E_{\text{c}} \) and \( E_{\text{c}} \). Hence, the difference between \( E_{\text{c}} \) and \( E_{\text{c}} \) may be attributed to the existence of some band tail states, which are in a certain distance from the band edge.

In summary, we have grown high-quality InAs\(_{x}\)Sb\(_{1-x}\) films on GaAs substrates by LPE and fabricated photoconductors. The InAs\(_{x}\)Sb\(_{1-x}\) films exhibit Hall mobilities higher than \( 2 \times 10^{4} \text{ cm}^{2} \text{ V}^{-1} \text{ s}^{-1} \) and cutoff wavelengths longer than 10 μm at RT, implying that there are a low inside-grain disorder density and structural ordering in the InAs\(_{x}\)Sb\(_{1-x}\) films. A notable photoresponse beyond 8 μm was observed with the photoconductors at RT. In particular, for the InAs\(_{x}\)Sb\(_{1-x}\) film with a midrange composition of \( x=0.3 \), the \( E_{\text{c}} \) is as low as 0.098 eV and there is a photoresponse of up to 13 μm with a maximum responsivity of 0.26 V/W at RT. Hence, LPE-grown InAs\(_{x}\)Sb\(_{1-x}\)/GaAs has demonstrated attractive properties for use in RT long-wavelength infrared detectors.

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### Table II. Maximum responsivities \( R_{\text{v, max}} \) and cutoff photon energies \( E_{\text{c}} \) obtained from the responsivity spectra of the photoconductors.

<table>
<thead>
<tr>
<th>Sample</th>
<th>( R_{\text{v, max}} ) (V/W)</th>
<th>( E_{\text{c}} ) (eV)</th>
<th>( E_{\text{c}}(77 \text{ K}) - E_{\text{c}}(300 \text{ K}) ) (eV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>30.4</td>
<td>0.15</td>
<td>0.150</td>
</tr>
<tr>
<td></td>
<td>28.3</td>
<td>0.26</td>
<td>0.136</td>
</tr>
<tr>
<td>C</td>
<td>0.15</td>
<td>0.26</td>
<td>0.150</td>
</tr>
<tr>
<td></td>
<td>0.136</td>
<td>0.26</td>
<td>0.136</td>
</tr>
</tbody>
</table>

FIG. 4. Spectral responsivity of the photoconductors fabricated on samples A and C at (a) 77 K and (b) room temperature.