Recent advances in Sb-based materials for uncooled infrared photodetectors

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In this paper, we review recent work on Sb-based materials for IR detector applications. The materials investigated in this work are InSb, AlInSb, and InAsSb grown by solid source molecular beam epitaxy (MBE). The photodiodes investigated are InSb p-i-n structures and InAs1-xSbx heterojunction and homojunction device structures grown on (100) and (111)B semi-insulating GaAs and Si substrates. The InAsSb was grown to result in a cutoff wavelength of \( \sim 8 \) \( \mu \)m which is useful for proximity fuse applications. Device performance is comparable to industry standard technologies such as HgCdTe and thermal detectors. The material parameters for device structures were investigated through theoretical calculations based on fundamental mechanisms.

1. Introduction

Infrared (IR) photodetectors are of great importance for military and civilian applications such as proximity detection, thermal imaging, and heat seekers. IR detectors are divided into two types: thermal and photon detectors. Thermal detectors absorb incident radiation which increases the temperature of the device. This change in temperature changes the resistance (bolometric detector) or the internal electric polarization (pyroelectric detector) which is measured by an external circuit. Thermal detectors exhibit a flat response that is high enough for IR thermal imaging (D^* \sim 10^8 \, \text{cmHz}^{1/2}/W), but since they are based on thermal effects, they are inherently slow. And although they can operate at room temperature, arrays based on thermal detectors must be thermoelectrically (TE) cooled for thermal stability.

For applications which require higher speeds, photon detectors are required. These detectors respond to incident light on an electronic level, thus they are characterized by high response speeds compared to thermal detectors and are generally cooled for \( \lambda > 3 \) \( \mu \)m to reduce competition from thermally induced transitions.

HgCdTe (MCT) is a well established material system which has been the dominant system for mid and long-wavelength IR photon detectors. However, MCT suffers from instability and non-uniformity problems over large areas due to the high Hg vapor pressure. Due to these problems, alternate material systems have been investigated. There has been interest in the use of heteroepitaxially grown InSb, InAsSb alloys, and strained layer superlattices as an alternative to MCT driven by the advanced material growth and processing technology available for Sb-based materials [1–5]. The material parameters of InSb, InAsSb, and HgCdTe are shown in Table 1.

For the mid-wavelength infrared (MWIR) atmospheric window (3–5 \( \mu \)m), bulk InSb is an established material system. State-of-the-art InSb arrays are fabricated by implanting a p-n junction on one side of a bulk InSb substrate, and thinning the other side using mechanical polishing. The chip must be thinned to \( \sim 10 \) \( \mu \)m due to absorption in the substrate. The chip is then In bump-bonded to a Si readout circuit. Thus, the radiation is incident on the back surface of the substrate. This thinning is a large yield limiter in the InSb array fabrication process. Using GaAs or Si substrates eliminates the need for substrate thinning since they are transparent over the useful wavelength range. In addition, these substrates can be easily etched away using wet chemical etching if desired. The work described in this paper resulted in high quality InSb on (100) and (111)B GaAs, and Si substrates.
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Table 1. Physical properties of InSb, InAsSb, and HgCdTe.

<table>
<thead>
<tr>
<th>Lattice structure</th>
<th>T (K)</th>
<th>InSb</th>
<th>InAsSb&lt;sub&gt;0.33&lt;/sub&gt;Sb&lt;sub&gt;0.67&lt;/sub&gt;</th>
<th>HgCdTe&lt;sub&gt;0.8&lt;/sub&gt;Cd&lt;sub&gt;0.2&lt;/sub&gt;Te</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>zinc blende</td>
<td>zinc blende</td>
</tr>
<tr>
<td>Lattice constant (Å)</td>
<td>300</td>
<td>6.4794</td>
<td>6.36</td>
<td>6.4645</td>
</tr>
<tr>
<td>Thermal Expansion Coefficient α (10⁻⁶K⁻¹)</td>
<td>300</td>
<td>5.04</td>
<td>4.3</td>
<td></td>
</tr>
<tr>
<td></td>
<td>80</td>
<td>6.50</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Density ρ (g/cm³)</td>
<td>300</td>
<td>5.7751</td>
<td>7.63</td>
<td></td>
</tr>
<tr>
<td>Melting point</td>
<td></td>
<td>803</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Energy gap E₉ (eV)</td>
<td>4.2</td>
<td>0.2357</td>
<td>0.138</td>
<td>0.064</td>
</tr>
<tr>
<td></td>
<td>80</td>
<td>0.228</td>
<td>0.136</td>
<td>0.09</td>
</tr>
<tr>
<td></td>
<td>300</td>
<td>0.180</td>
<td>0.100</td>
<td>0.165</td>
</tr>
<tr>
<td>Thermal coefficient of E₉</td>
<td>100-300</td>
<td>-2.8 × 10⁻⁴</td>
<td>+3 × 10⁻⁴</td>
<td></td>
</tr>
<tr>
<td>Effective masses: m&lt;sub&gt;e&lt;/sub&gt;/m&lt;sub&gt;i&lt;/sub&gt;</td>
<td>77</td>
<td>0.0145</td>
<td>0.0101</td>
<td>0.005</td>
</tr>
<tr>
<td></td>
<td>300</td>
<td>0.0116</td>
<td></td>
<td></td>
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<tr>
<td></td>
<td>4.2</td>
<td>0.0149</td>
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<td></td>
</tr>
<tr>
<td></td>
<td>4.2</td>
<td>0.41</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Mobilities: μ&lt;sub&gt;e&lt;/sub&gt; (cm²/V·sec)</td>
<td>77</td>
<td>10⁶</td>
<td>5 × 10⁵</td>
<td>2.5 × 10⁵</td>
</tr>
<tr>
<td></td>
<td>300</td>
<td>8 × 10⁴</td>
<td>5 × 10⁴</td>
<td></td>
</tr>
<tr>
<td>μ&lt;sub&gt;h&lt;/sub&gt; (cm²/V·sec)</td>
<td>77</td>
<td>10⁴</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>300</td>
<td>800</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Intrinsic carrier concentration (cm⁻³)</td>
<td>77</td>
<td>2.6 × 10⁹</td>
<td>2 × 10¹²</td>
<td>8 × 10¹³</td>
</tr>
<tr>
<td></td>
<td>200</td>
<td>9.1 × 10¹⁴</td>
<td>8.6 × 10¹⁵</td>
<td></td>
</tr>
<tr>
<td></td>
<td>300</td>
<td>1.9 × 10¹⁶</td>
<td>4.1 × 10¹⁶</td>
<td>3 × 10¹⁶</td>
</tr>
<tr>
<td>Static dielectric constant ε&lt;sub&gt;s&lt;/sub&gt;</td>
<td>17.9</td>
<td></td>
<td></td>
<td>18.5</td>
</tr>
<tr>
<td>High freq. dielectric constant ε&lt;sub&gt;∞&lt;/sub&gt;</td>
<td>16.8</td>
<td></td>
<td></td>
<td>13.0</td>
</tr>
</tbody>
</table>

Following the optimization of heteroepitaxial InSb for the MWIR, InAsSb was investigated for uncooled λ < 8 μm applications. The InSb MWIR imaging arrays were cooled to 80K. Many applications which require portability, lower cost, higher reliability, and ease of use make cooled arrays less desirable. Thus, there is great interest in detectors which can operate at room temperature without the need for cryogenic or thermoelectric coolers. As mentioned previously, thermal detectors are capable of room temperature IR detection, but require TE cooling. In addition, applications such as projectile fuzes and situational awareness require high speed detection.

For this reason, InAsSb was investigated for uncooled IR detection. Earlier data suggested that InAsSb can exhibit a cut-off wavelength up to 10 μm at 77 K and 12.5 μm at 300 K. Some recent experimental results demonstrated that the cutoff wavelength of InAsSb epitaxial materials can be longer than 12.5 μm at near room temperature which may be due to structural ordering [6, 7]. However, the exact mechanism has not been determined yet. The bandgap of InAsSb versus composition is shown in Fig. 1. In comparison to HgCdTe, InAs₁₋ₓSbₓ exhibits the inherent advantages of high stability, well behaved donor and acceptor impurities, high mobility, and availability of low cost and high quality substrates such as GaAs. These results show promise for LWIR InAsSb photodetectors operating at room temperature. However, InAsSb photodetectors are limited at high temperatures due to strong thermal generation and recombination of charge carriers. Several solutions have been proposed to suppress the noise due to Auger recombination [8, 9]. These include the optimization of the detector structure by controlling the composition, doping level, and thickness. A comparison of theoretical performance of uncooled sensors operating at wavelengths of λ < 8 μm is shown in Table 2.

Table 2. Comparison of the detectivities (cmHz¹/²/W) of HgCdTe, InSb, and InAsSb in the λ < 8 μm range at 300 K

<table>
<thead>
<tr>
<th>λ(μm)</th>
<th>MCT</th>
<th>InSb</th>
<th>InAsSb</th>
</tr>
</thead>
<tbody>
<tr>
<td>5</td>
<td>2 × 10⁹</td>
<td>7 × 10⁸</td>
<td>2 × 10⁹</td>
</tr>
<tr>
<td>6</td>
<td>7 × 10⁸</td>
<td>7 × 10⁸</td>
<td>7 × 10⁸</td>
</tr>
<tr>
<td>7</td>
<td>4 × 10⁸</td>
<td>4 × 10⁸</td>
<td>4 × 10⁸</td>
</tr>
<tr>
<td>8</td>
<td>2 × 10⁸</td>
<td></td>
<td>2 × 10⁸</td>
</tr>
</tbody>
</table>
Previous work on heteroepitaxial InSb and InAsSb has been primarily on studying the optimum growth conditions and material characterization. The growth of InSb by MBE [10–28] and MOCVD [2,29–33] on various substrates such as GaAs, Si, and Al2O3 has been studied extensively. Much work has also been done on the growth and characterization of InAsSb, also by MBE [3–5,34–45] and MOCVD [7,46–51]. Although work at Bell Labs covered the entire InAsSb composition range, most work in detectors from this material system concentrated on As-rich materials for MWIR applications. The work described here is on Sb-rich material on GaAs substrates.

2. Infrared photodetector design and operation

Consider a simple p–n homojunction photovoltaic structure. Electron-hole pairs generated within a diffusion length of the junction diffuse to the depletion region and are accelerated in opposite directions. In this way, minority carriers become majority carriers on the other side of the junction which shifts the current-voltage curve down as shown in Fig. 2. The magnitude of the photogenerated current is given by:

\[ I_{ph} = \eta q A \Phi \]  

where \( \eta \) is the quantum efficiency, \( q \) is the electronic charge, \( A \) is the junction area, and \( \Phi \) is the incident flux. Generally, the gain in photodiodes is \( \approx 1 \) when not operated in avalanche mode. Photodiodes can operate at any point on the current-voltage characteristic, but reverse bias is generally used for high frequency applications to reduce the RC time constant of the device. Many applications utilize photodiodes under zero bias. A figure of merit for diodes in this condition is the differential resistance-area product given by:

\[ R_o A = \left( \frac{\delta J}{\delta V} \right)_{V=0}^{-1} \]  

where \( J = I/A \) is the current density. The resistance-area product is a measure of the leakage current of the device: a high \( R_o A \) indicates a high quality diode. Many mechanisms contribute to the reduction of the \( R_o A \) [52]: diffusion current in the bulk p and n regions, generation-recombination in the depletion region, band-to-band tunneling, intertrap-to-band tunneling, ohmic leakage across the depletion region, and surface leakage. High quality bulk IR photodiodes are generally limited by diffusion outside the depletion region, generation-recombination within the depletion region, and tunneling through the depletion region.

To achieve the highest possible detectivity, the ratio of the optical carrier generation to the thermal carrier generation should be maximized. Also, the volume of the region of non-zero photoelectric gain where thermal generation and recombination occurs should be minimized. The optimization of the photovoltaic detector structure for a given material can be made by a proper selection of doping levels and layer thicknesses.

The detectivity of a thermal noise limited photodetector is given by:

\[ D^* = \frac{\lambda \eta q}{hc} \times \frac{(R_o A)^{1/2}}{(4kT)^{1/2}} \]  

where \( \lambda \) is the wavelength, \( h \) is Planck’s constant, \( c \) is the velocity of light, \( R_o A \) is the zero-bias resistance-area product given previously, \( k \) is Boltzmann’s constant, and \( T \) is the temperature. For the best performance under given operating conditions (wavelength
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and temperature), the value of $\eta(R_0 A)^{1/2}$ should be maximized. Calculations for various dark current mechanisms such as diffusion, generation-recombination, Auger, and tunneling, have shown that lightly doping the active region of the detector improves room temperature performance by minimizing Auger recombination which is the limiting mechanism in this temperature region [53].

The photoelectric gain is close to unity at distances within the corresponding diffusion length from the p–n junction. To ensure good photosresponse, the optical generation should occur in areas of near unity photoelectric gain. This requires that the illuminated side of the p–n junction be thinner than the diffusion length. But if the device is too thin, the absorption decreases which causes a decrease in the total quantum efficiency. The compromise between absorption and photoelectric gain to maximize the quantum efficiency gives the optimum thickness of the absorber region which is determined to be around 5 μm for an InAs$_{0.15}$Sb$_{0.85}$ photodetector. Calculation of the quantum efficiency is described in Section V.

The highest performance can be obtained using n$^+$–p photodiodes illuminated through the thin n$^+$ region. Due to the Moss-Burstein effect, the cut-off wavelength of the n$^+$ region is significantly shifted toward shorter wavelength. Therefore, the n$^+$ region is optically transparent and acts as an optical window for IR radiation. At the same time, the Auger processes are largely suppressed, and the main contribution to the total dark current and noise comes from the limited volume of the p-type layer. The radiation is absorbed in the lightly doped p-type region of the structure. The absorption occurs primarily in the space charge region and within a distance of about 1/α where the photoelectric gain is close to unity. So in principle, the device can obtain the highest performance for a given wavelength and operating temperature.

However, there are a lot of thermal generation and recombination processes at the back surface for an n$^+$–p device with a thin p-region. The device also suffers from high series resistance (as compared to the junction resistance) originating from the high resistance of the p-region due to both low hole mobility and low thickness. Another problem arises from the thermal generation and recombination at the contact to the p-region. These problems can be alleviated by the use of a heterojunction contact layer.

The first detector structures described here were simple p-i-n structures. The p and n regions were doped in-situ during growth. Subsequent devices were grown on wider gap layers, and finally double-heterostructure devices were grown on wider gap barrier layers nucleated on GaAs substrates. These devices make it possible to optimize the parameters of the narrow gap absorber region and the heavily doped contact regions independently, thus offering important advantages. Consider illumination from the n$^+$ side. Since the n$^+$ region is of wider gap, radiation with energy $E_{gr} < h\nu < E_{gr}$ is absorbed only in the narrow gap semiconductor, predominantly in the vicinity of the n$^+$–p junction (or p$^+$–n junction with an undoped active region), which results in high quantum efficiency. Thermal generation can be limited to the absorber region only. Contacts to heavily doped regions do not contribute to the total thermal generation, since photoelectric gain is very low there due to the short diffusion length in heavily doped regions.

The wider gap layers can improve performance due to better carrier confinement and lower thermal generation as mentioned above. The use of these wider gap barriers also led to a two-color detector, which showed different response versus wavelength when a bias was applied. This is described in the device measurement section.

3. Growth and characterization of Sb-based materials

The InSb and InAsSb layers were grown by molecular beam epitaxy (MBE) on (100) and (111) semi-insulating GaAs and Si substrates. The layers were characterized by various characterization techniques. Structural characterization of the epitaxial layers was performed using high-resolution x-ray diffraction spectra and the composition was determined using Vegard's law. Hall measurements were performed using a Bio-Rad HL 5560 system and Van der Pauw method at both 300 K and 77 K. Field-dependent Hall measurements were also performed to study the effect of the InSb/GaAs interface and the parallel conduction channels which result from highly mismatched epilayers. Optical quality was assessed using photoluminescence (PL) and photoresponse measurements. Transmission and photoresponse measurements were carried out using a Fourier Transform Infrared (FTIR) spectrometer. Photoreponse measurements will be described in the device measurement section.

3.1. MBE growth and characterization

MBE is quite suitable for the growth of hetero-epitaxial Sb-based materials due to the excellent flux control as a result of in-situ monitoring techniques.
such as reflection high energy electron diffraction (RHEED) and precise thickness control using shuttering mechanisms. The samples were grown on (100) and (111)B oriented semi-insulating GaAs and (100) Si substrates in an EPI Modular Gen II solid source reactor with uncracked elemental sources. The substrates were epi-ready, and were In-mounted to molybdenum blocks before being loaded into the system. The growth of InSb on (100) GaAs and Si has been reported elsewhere [54–57]. X-ray diffraction FWHM’s of ~ 50 arcsec were obtained on a 6 μm thick epilayer on a 3” GaAs substrate. The variation of the FWHM across the substrate was typically +-3 arcsec.

Growth on GaAs coated Si typically resulted in x-ray FWHM’s of ~ 100 arcsec. The optimum growth conditions for (111) InSb were found to be quite similar to (100) growth but a slightly higher substrate temperature was required to obtain mirror-like surfaces. This difference has been attributed to the different bonding characteristics of the (111) surface. On the (111)B surface, there are three dangling bonds for the group V atom, and one dangling bond for the group III atom as compared to two for each atom on (100) material. Thus, at a given V/III ratio, a higher temperature is required for stoichiometric growth. Fig. 3 shows typical X-ray FWHM for (100) and (111)B InSb grown on GaAs substrates versus epilayer thickness. It is evident that there is negligible difference between the crystallinity of the two types of material. Hall measurements also exhibited similar characteristics as shown in Fig. 4, which shows 300 K mobility versus epilayer thickness. Note that the mobility approaches that of bulk InSb at thicknesses greater than ~2 μm.

Initial InSb samples were grown with six nines (99.9999%) pure Sb source material resulting in a measured background carrier concentration of $n = 10^{16}$ cm$^{-3}$ at 80 K. The mobility versus temperature curve showed typical $n$-type behavior with peak mobilities of ~100000 cm$^2$/Vs at 80K for samples of 3–5 μm thickness. However, when higher purity (seven nines) Sb source material was used, the measured background concentration was ~10$^{15}$ cm$^{-3}$, and the peak electron mobility occurred at ~200 K. At lower temperatures, the measured electron mobility dropped drastically to ~3000 cm$^2$/Vs. The measured mobility versus temperature curve is shown in Fig. 5 for epilayers of 2.0, 5.0, and 9.2 μm thickness with $n \approx 10^{15}$ cm$^{-3}$ background concentration. Note that at temperatures below the peak in mobility, the mobility drops off sharply. This does not necessarily imply that the material is of poorer quality than the previously grown material, but that the mobility measurement is being dominated by a low mobility parallel conduction channel, most likely the highly dislocated region near the InSb/GaAs heterointerface. To study these results, field dependent Hall measurements were performed.

The InSb epilayer can be considered to be composed of three distinct layers: a highly dislocated interface layer, a bulk-like layer, and a surface accumulation (or depletion) layer. At low temperatures, the interface layer dominates the measurements because of the high number of carriers there. This was verified using several techniques:

First, InSb epilayers on GaAs were doped at levels near $10^{16}$ cm$^{-3}$. These layers showed 80 K mobilities back up to that obtained when six nines Sb source material was used. Previous work in optimizing InSb on GaAs mobility also required $n$-type doping to obtain high electron mobility at low temperatures [58].

Second, several samples were epoxied to glass
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Fig. 5. InSb on GaAs electron mobility versus temperature for three layer thicknesses.

slides epi-side down, mechanically polished to remove the indium, and chemically etched to remove the GaAs substrate. The resulting surface was mirror-like with no visible cracks or defects as a result of the substrate removal. This process was performed on the samples shown in Fig. 5. Much higher mobilities were obtained at 80 K than before the substrates were removed. The 5.0 \mu m sample in Fig. 5 exhibited an 80 K mobility of \sim 100000 \text{ cm}^2/\text{Vs} as indicated by the dot inserted at 80K in the figure. Before removal of the GaAs substrate, the 80K mobility of this same sample was measured at \sim 3000 \text{ cm}^2/\text{Vs}.

At temperatures below \sim 100 K for all three samples shown in Fig. 5, the measured carrier concentration remains constant and the sheet density is essentially equal for all three samples. This implies that a thickness independent layer (the interface) is the dominant conduction channel at low temperatures, because the bulk-like layer would show thickness dependence in the sheet density.

Third, calculated mobility spectra obtained from B-field dependent Hall measurements were used to investigate the number and type of conduction channels in the epilayer. The spectra are calculated from B-field dependent Hall coefficient and conductivity data at various temperatures as described elsewhere [59-61]. The calculated mobility spectra for the 9.2 \mu m sample at 100 K and 210K are shown in Fig. 6. Peaks in the spectra indicate conduction channels – negative (positive) mobilities indicate n-type (p-type). As shown in Fig. 6(a), there are several conduction channels at 210 K, the dominant one being the high mobility (\sim 120000 \text{ cm}^2/\text{Vs}) channel which is most likely the bulk-like layer. The low mobility peak is thus most likely a result of the highly dislocated region at the

InSb/GaAs interface. The middle n-type mobility (\sim 25000 \text{ cm}^2/\text{Vs}) may be due to a quasi-two dimensional surface layer. The positive (p-type) mobility peaks in the spectra are artifacts that come about as a result of experimental error or from the relaxed assumption that the carrier density and mobility are independent of magnetic field. As shown in Fig. 6(b), at 100 K, the low mobility peak increases in relation to the high mobility peak, with a conductivity on the same order of magnitude as the bulk-like layer. However, since the concentration (n \sim \sigma/\mu) is higher for the low mobility layer, it is expected to dominate the mobility measurement at this temperature. This is indeed the case, as the measured mobility at 100 K is on the order of a few thousand \text{ cm}^2/\text{Vs}. This drop in the conductivity of the bulk-like layer is expected as the intrinsic carrier concentration drops with decreasing temperature.

The optical quality of the heteroepitaxial InSb was assessed using time-dependent PL measurements to determine minority carrier lifetimes. Typical carrier lifetimes for early InSb material on both GaAs and Si substrates were \sim 30 nsec at 80 K. With improvement in the surface morphology on later growths through further optimization of growth conditions, the carrier lifetime was measured at \sim 240 nsec as shown in Fig. 7, which is comparable to that of bulk InSb at this con-
4. Detector fabrication

The fabrication techniques utilized in this work were all standard processing techniques such as photolithography, wet chemical etching, and e-beam metal vaporization. Because of the narrow bandgaps of these materials, ohmic contacts are quite simple to obtain. The InSb photodiodes typically consisted of a ~2 μm n⁺ region (= 10^{18} cm⁻³), a ~6 μm unintentionally doped region (n = 10^{15} cm⁻³ 77K), and a ~0.5 μm p⁺ (= 10^{18} cm⁻³) contact layer. Photodiodes were fabricated with simple 400 × 400 μm² mesa structures prepared by standard photolithography techniques and wet chemical etching (lactic:nitric acid). Ohmic contacts for both n- and p-type layers were made by depositing Au/Ti using an e-beam evaporator. The contact pattern was defined again by standard photolithography techniques and selective etching (KI:K₂H₂O for Au and HF for Ti etching) or using lift-off. Lift-off was used when AllnSb barrier layers were used since the gold etch would damage the surface. The chips were assembled using standard integrated circuit technologies including die separation, mounting, and Au-wire bonding.

For the realization of InAsSb photovoltaic detectors, several device structures were investigated. First, p⁺–InSb/p–InAs₁₋ₓSbₓ/n⁺–InSb double heterojunction structures were grown on semi-insulating GaAs substrates. The bottom n⁺–InSb layer with ~2 μm thickness was doped at 2 × 10^{18} cm⁻³ both for increased quantum efficiency due to the pronounced Moshkurstein effect and for a low series resistance due to high electron mobility. The thickness of the active layer was ~2–5 μm to ensure both high quantum efficiency and high optical gain. The top contact layer p⁺–InSb was 0.5 μm thick and doped at a level of 10^{18} cm⁻³.

Some device structures simply utilized the bottom InSb layer as a buffer layer with the bottom contact made directly to the InAsSb active region. Subsequent devices utilized a wider gap AllnSb buffer (or bottom contact) layer. This structure gave flexibility in the device in that the mesa could be etched down to the InAsSb active region or down to the AllnSb layer which resulted in a heterojunction. The only difficulties encountered in the fabrication of these devices was making contact to p-type AllnSb or GalnSb due to roughness from etching. A rough surface occasionally caused poor adhesion of the gold contact which could be pulled of the sample when making contact with a wire bonder. However, this only occurred on sub-optimized material.

5. Characterization and analysis of InSb and InAsSb detectors

The photodetectors were mounted in a liquid nitrogen cooled cryostat system and measurements were taken at temperatures between 77 K and 300 K. The relative photoresponse spectra were measured with an FTIR spectrometer. The absolute photoresponse was determined using a calibrated blackbody test set, which is composed of a blackbody source, preamplifier, lock-in amplifier, and chopper system. Responsivity measurements as a function of frequency showed that the thermal effect could be neglected at frequencies higher than 200 Hz. The measurements were carried out with a blackbody source at a temperature of 800°C and modulating frequency of ~ 400 Hz.

5.1. InSb photodiodes

InSb photodiodes were grown on 3” Si and (111) GaAs substrates. Fig. 8 shows the spectral response of an InSb photodiode grown on a 3” Si substrate by MBE. The detectivity measured for this device was ~ 3 × 10^{10} cmHz¹/₂/W. The morphology was mirror-like and the x-ray FWHM for this sample was 109 ± 3 arcsec across the entire wafer, indicating excellent...
crystallinity and uniformity. More recently, typical FWHM's have been consistently < 100 arcsec. The best x-ray FWHM obtained in this work for InSb heteroepitaxial growth is 56 arcsec for an undoped epilayer on (100) GaAs. InSb photodiodes grown on (111) GaAs have exhibited similar characteristics: typical x-ray FWHM's are 100 arcsec and the photoresponse surpasses that of the structures grown on Si as shown in Fig. 9.

The InSb diodes were modeled to determine optimum device structure and the measured dark currents were fitted to theoretical calculations to extract the dominant limiting mechanisms. The quantum efficiency (QE) was calculated for the device structure shown in Fig. 10. Through calculation of the photogenerated current density, the QE can be calculated using eq. (4).

\[
\eta = \frac{hc}{q\lambda} \times \frac{J_{ph}(V_a)}{I_o}
\]  

\[\eta = \frac{(1-r)\alpha L_n}{\alpha^2 L_d^2 - 1} \times e^{-\alpha(x_n+w)} \times \]  

Fig. 10. Diode structure used for the quantum efficiency calculation.

The QE of the device structure is the sum of the three regions: p- and n-regions, and the depletion region. These are given by Eqs. 5–7.

\[\eta = \frac{(1-r)\alpha L_n}{\alpha^2 L_d^2 - 1} \times \]  

(a) Quantum efficiency vs. Thickness of n-region (µm)

(b) Quantum efficiency vs. Thickness of n-region (µm)

Fig. 11. Calculated QE for a) InSb on GaAs and b) Bulk InSb Photodiodes.
\[
\eta_v = -\frac{J_{ph}h\alpha}{q\Lambda L_h} = \frac{(1-r)\alpha L_h}{\alpha^2 L_h - 1}\left[\alpha L_h + \gamma_1 e^{-\alpha x_n}\left(\frac{\gamma_1}{\gamma_1 - \gamma_2} - \gamma_2 \frac{x_n}{L_h}\right) + \gamma_2 \frac{x_n}{L_h}\right]
\]

\[
\eta_{DR} = (1-r)\left[e^{-\alpha x_n} - e^{-\alpha(x_n + w)}\right]
\]

The QE versus n-region layer thickness for a device using material parameters for heteroepitaxial InSb and bulk material is shown in Fig. 11. The parameters for heteroepitaxial InSb are:

- \(r\) = reflection coefficient with no anti-reflection coating = 0.35
- \(\alpha\) = absorption coefficient calculated from transmission measurements = 3400 cm\(^{-1}\)
- \(\gamma_1\) = surface recombination velocity coefficient at n-type surface = 0
- \(\gamma_2\) = surface recombination velocity coefficient at p-type surface = 0
- \(L_e = L_h\) - diffusion length of electrons and holes = 2 \(\mu\)m (= 10–20% of bulk material diffusion length determined using measured mobilities and carrier lifetimes)
- \(N_A\) = p-type doping level = \(1 \times 10^{18}\) cm\(^{-3}\)
- \(N_D\) = n-type doping level = \(1.5 \times 10^{15}\) cm\(^{-3}\).

Assuming a zero surface recombination velocity coefficient is a simplified case, but using \(\gamma_1 = \gamma_2 > 0\) does not change the shape of the curves much, just the magnitudes. This calculation cannot give exact values for QE, as there are many simplifications in the derivation. The main objective of the calculation is to find optimum structures for the best device performance, not to predict the exact performance. The values for the bulk material calculation were essentially the same except \(L_e = L_h = 15\) \(\mu\)m and \(\alpha = 5000\) cm\(^{-1}\). Figure 11 shows that the QE of a heteroepitaxial InSb diode (with 30 ns minority carrier lifetime) has a maximum at a thickness of a few microns, whereas bulk detectors have a peak at ~ 9 \(\mu\)m, which is a common thickness for commercial InSb detectors. However, with the recent MBE grown material exhibiting ~ 240 ns carrier lifetime, thicker devices are optimum.

In addition to the device QE, the p–n junction dark currents were analyzed to determine which leakage mechanisms determine the device performance. The current mechanisms considered in this analysis were [65]: diffusion due to carrier generation outside the depletion region (I\(_{DS}\)), generation-recombination in the depletion region (I\(_{G}\)), tunneling (I\(_T\)), shunt current due to leakage along the edges of the device and through dislocations (I\(_S\)), and surface diffusion (I\(_{DS}\)). These current mechanisms are described by Eqs. 8–12:

\[
I_{diff} = AT^3\exp(-E_g/kT)\{\exp(qV/kT) - 1\}
\]

\[
I_{DS} = BT^{3/2}\exp(-E_g/2kT)\{\exp(qV/kT) - qV/kT - 1\}^{1/2}
\]

\[
I_G = CT^{3/2}\exp(-E_g/2kT)(V_b - V)^{1/2}\{\exp(qV/2kT) - 1\}
\]

\[
I_T = D(V_b - V)^{1/2}\exp(-E_g/kT)\{\exp(qV/2kT) - 1\}^{1/2}
\]

\[
I_S = FVT^{3/2}\exp(-E_g/2kT)
\]

where \(E_g\) is the bandgap, \(k\) is Boltzmann's constant, \(V\) is the applied bias, \(T\) is the temperature, \(V_b\) is the built-in potential, and \(A, B, C, D, E,\) and \(F\) are fitting parameters. The shunt current is linear with bias and shows a temperature dependence following that of the intrinsic carrier concentration. The measured and fitted currents for a heteroepitaxial InSb photodiode are shown in Fig. 12. As expected for material grown with such a large lattice mismatch, shunt current dominates at 77K at lower biases, until tunneling becomes dominant.

This high quality InSb on GaAs was used for imaging. In collaboration with Lockheed Martin Fairchild Systems, IR imaging was obtained with heteroepitaxial InSb from a 256 \(\times\) 256 array which was fabricated from a 9 \(\mu\)m thick InSb layer on a 3" GaAs substrate. The p–n junctions were formed by ion implantation.
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Fig. 12. Measured and fitted dark currents for an InSb diode on GaAs.

The chip was In bump-bonded to a CMOS readout circuit and the GaAs substrate was chemically etched to remove any effect of thermal mismatch between the readout and the GaAs substrate. Good imaging was obtained as shown in Fig. 13. This is the first report of imaging from heteroepitaxial InSb.

Fig. 13. IR Images obtained from an InSb on GaAs FPA.

5.2. InAsSb photodiodes

Initial InAsSb photodiode structures were grown on InSb buffer layers nucleated directly on GaAs substrates. Subsequent structures were grown on an AlInSb buffer layer nucleated on GaAs. This wider gap buffer layer adds the advantages of a wider energy barrier to block carriers from the highly dislocated interface at the substrate which can improve the differential resistance-area product ($R_nA$), and also can be used as a contact layer in a single heterostructure diode. A typical absolute photoresponse curve of an InAsSb photodiode on an InSb buffer layer is shown in Fig. 14. The 300K Johnson-noise limited detectivity for this device was calculated at $\sim 1 \times 10^8$ cmHz$^{1/2}$/W. Because of the highly dislocated interface layer and the unpassivated surface, the $R_nA$ for this device was $\sim 10^{-3}$ Ω cm$^2$ at 300 K. The $R_nA$ was increased by an order of magnitude by incorporating an AlInSb buffer layer resulting in a peak detectivity of $\sim 3 \times 10^8$ cmHz$^{1/2}$/W. The absolute photoresponse for an InAsSb homojunction on an AlInSb buffer layer is shown in Figure 15. Typical detectivities of existing MCT material available for this wavelength region is in the mid-$10^7$ cmHz$^{1/2}$/W range without optical immersion, and in the low-$10^8$ cmHz$^{1/2}$/W range with optical immersion.

Fig. 14. Room temperature absolute photoresponse of an InAsSb photodiode on InSb/GaAs.

Fig. 15. Room temperature absolute photoresponse of an InAsSb photodiode on AlInSb/GaAs.
This shows the promise for this material to compete with MCT in high speed uncooled sensor applications.

The wider gap AllnSb layer can also be used as the bottom contact layer resulting in a single heterostructure photodiode. The heterojunction increases the $R_0$A of the device from $10^{-3}$ $\Omega$cm$^2$ to $10^{-1}$ $\Omega$cm$^2$. The current and differential resistance curves versus bias are shown for an InAsSb/AllnSb heterostructure device in Fig. 16. The peak differential resistance of 70 $\Omega$ at zero bias results in an $R_0$A of 0.11 $\Omega$cm$^2$ at 300K. The absolute photoresponse for this structure at 300 K is shown in Fig. 17. The Johnson-noise limited detectivity was estimated at $\sim 4 \times 10^8$ cmHz$^{1/2}$/W at 3 $\mu$m and $\sim 5 \times 10^7$ cmHz$^{1/2}$/W at 6 $\mu$m. The response curve shown was obtained under a small bias ($\sim 0.15$ V) since the device showed little or no response at zero bias. This may be due to the alignment of the band edges of InAsSb and AllnSb where electrons generated in the InAsSb region encounter a

6. Conclusions

InSb and InAsSb photodetector structures have been optimized through calculations based on fundamental processes. We have successfully grown InSb p-i-n, InAsSb homojunction, and InAsSb/AllnSb single heterojunction photodetector structures with determined doping level, thickness, and composition on GaAs substrates by solid source MBE. InSb photodiodes exhibited detectivities of $\sim 3 \times 10^{10}$ cmHz$^{1/2}$/W with photoresponse up to room temperature. The optimum device structure was investigated by modeling the quantum efficiency and device quality was assessed by modeling the dark currents. InAsSb photodetectors operating at room temperature using Sb-based materials with $\lambda < 8$ $\mu$m have been demonstrated. Photoresponse of InAsSb photovoltaic detectors with detectivities comparable to MCT and
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![Graph](image)

- Exp. MCT at 300K
- CQD InAsSb at 300 K
- --- Thermal detectors

Fig. 19. Experimental detectivities for MCT, InAsSb, and thermal detectors at 300K as well as the theoretical curves for MCT.

thermal detectors was obtained up to room temperature. Fig. 19 shows the comparison between the InAsSb described in this work and experimental values for MCT and thermal detectors as well as the theoretical values for MCT.

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