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Materials and process issues in the fabrication of high performance VLWIR HgCdTe infrared detectors

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Issues relating to fabrication of VLWIR HgCdTe detectors with high performance and uniformity are reviewed. The primary mechanism operative in the activation of p type dopants in HgCdTe is discussed along with implications for fabrication of high performance detectors. Origin of native defect related deep centres in limiting the minority carrier lifetime is explored.

Keywords: HgCdTe photodiodes, R_0A product, heterostructures, lateral collection.

1. Introduction

The last two decades have witnessed a significant progress in HgCdTe material and detector technology [1]. Greatly improved control of unwanted impurities, point defects, and extended defects, and improvement in device architecture, device growth and junction formation techniques, and surface passivation have all contributed to this progress.

In order to make further improvements in the device technology, understanding of the underlying issues is essential. In this paper, we briefly review issues relating to maximisation of zero bias resistance-area product (R_0A), quantum efficiency, and uniformity. We discuss the primary mechanism operative in the activation of p type dopants in HgCdTe. We explore the origin of native defect related Shockley-Read centres which could degrade the minority carrier life time, and possible ways to minimise their concentration via post growth anneal.

2. Issues relating to HgCdTe VLWIR detectors for low background applications

VLWIR HgCdTe detectors need to be operated at low temperatures of 40 K to 50 K because as the cut off wavelength increases (or band gap decreases), thermally generated carriers (noise) increase at a given temperature of operation. Lower temperatures of operation will minimise this noise. Also as the background photon flux decreases, thermal noise reduction below that of the background demands lower operating temperatures.

The next generation of focal planes requires detector arrays exhibiting background limited detectivity at extremely low backgrounds and with extremely high uniformity and operability – much in excess of 90%. The two essential detector performance parameters on which detectivity depends are: R_0A, the zero bias impedance-area product, and quantum efficiency (QE), which is a measure of the number of electron-hole pairs that the detector can collect per photon of energy incident on it. R_0A is a measure of the thermal noise of the detector, with higher R_0A indicating lower thermal noise. Higher QE represents higher signal.

In order to maximise the thermal noise limited detectivity, so that the detectors can become background noise limited, we need to maximise the R_0A value and thereby minimise detector thermal noise, while at the same time maximising the quantum efficiency.

2.1. R_0A maximisation

At operating temperatures for low backgrounds, the detectors are generation–recombination (g-r) or tunnelling current limited. Key factors contributing to these currents include:

- substrate defects, especially Te precipitates and dislocations that propagate into base layer and eventually into the heterostructure/space charge region,
- dislocations generated in the base layer during growth, that propagate into the heterostructure/space charge region,
- heterostructure interface dislocations created during cap layer growth due to lattice mismatch of the layers,
- deep level impurity centres in the space region due to group V dopants such as phosphorus, arsenic, etc.,
- native defect related deep level centres in the space charge region which aid trap assisted tunnelling.

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To maximise detector R_oA, one should use substrates with low etch pit density and free of Te precipitates. Te precipitates less than 1 m in size, which can remain undetected by IR microscopy, in CdZnTe substrates can be minimised with anneals in CdZn atmospheres [2,3].

The minimisation of heterostructure interface dislocations is even more critical than the minimisation of the base layer/substrate interface dislocations for VLWIR detector operation at temperatures around 40 K, since the devices are g-r and tunnelling dark current limited. Growth of a precisely lattice matched quaternary HgCdZnTe wider band gap cap layer may ultimately result in minimisation of these base layer/cap layer heterostructure interface dislocations.

R_oA maximisation also needs minimisation of any deep level centres associated with group V dopants and native defects. Experiments where site transfer anneals are carried out at various Hg pressures will lead to an optimum pressure of Hg that gives high site transfer efficiency, and yet low concentration of deep level centres.

For very long cut-off wavelengths, interband tunnelling currents could pose a serious problem necessitating development of doping techniques to reproducibly grow/anneal base layers to yield N_D–N_A less than 10^{15} cm^{-3}.

2.2. Quantum efficiency maximisation

When the p-n junction is located in the compositionally graded region of the heterostructures, and its width is larger than the depetion width (Fig. 1, case 1), it can give rise to barriers in the valence band [4]. This can lead to inefficient collection of the minority holes, and thereby result in poor quantum efficiency. This can be alleviated with minimisation of heterostructure interdiffusion occurring at the growth temperature, and during the site transfer anneals required for group V acceptor activation.

Also, barriers in the valence band to the collection of minority carrier holes can arise when the band gap of the cap layer of the heterostructure is too high, causing the Fermi level to be farther away from the valence band edge (Fig. 1, case 2). In this case, the ohmic contact to the capping layer will be poor, and will block the minority carrier holes from being collected efficiently. This will result in poor quantum efficiency. Growth of a cap layer of optimum composition such that its band gap is high enough to provide good passivation, while not so high that it becomes a blocking contact to the collection of holes, is essential.

Since most of the second and third generation focal plane arrays are back side illuminated, the base layer thickness also needs to be optimised so that quantum efficiency is not lost with thickness greater than that needed for optimum absorption of photons.

Wider band gap cap layer thickness also needs to be optimised so that it is thick enough to provide good passivation of the VLWIR base layer, but not so thick that the implanted ions cannot get past the cap layer thereby causing the junction to occur in the wider band gap.

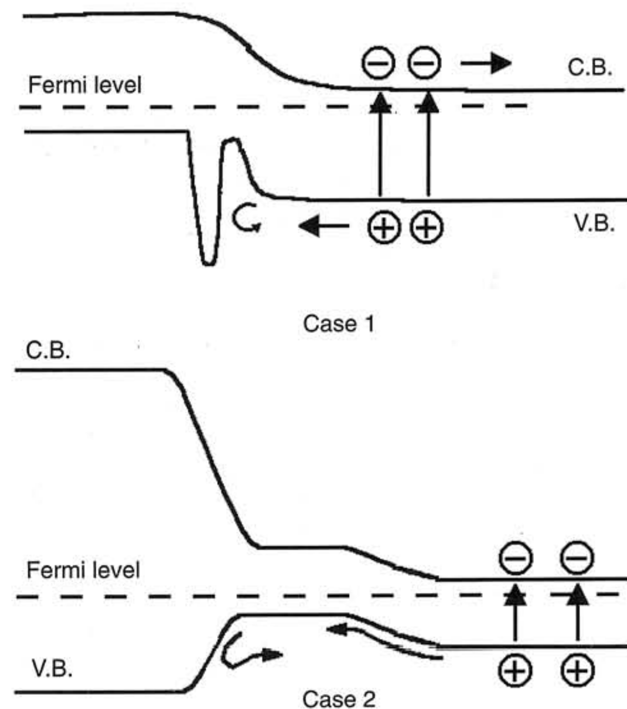

Fig. 1. Decrease of quantum efficiency because of creation of valence bandbarriers to the collection of holes as a result of excessive compositional grading (case 1) or as a result of the cap layer bandgap being too high causing Fermi level to lie farther away from the valence band edge (case 2).

2.3. Uniformity and operability maximisation

Molecular beam epitaxy (MBE) growth technology has advanced to a state where the absorber layer and the cap layer both are extremely uniform with regard to thickness, composition, carrier concentration, Te precipitates, dislocation density, native defect concentration etc. However, a non-uniformity in any of these, not detectable by any direct diagnostic means, could still exist, and affect the performance uniformity and operability.

Through the implementation of the buried planar heterostructure and lateral collection photodiode approaches described below, it is possible to improve the uniformity, and operability at all temperatures of operation while improving noise and detectivity in the g–r current limited temperature region.

2.3.1. Buried planar heterostructure photodiode architecture

Figure 2 shows a cross-section of the buried planar heterostructure detector. The optically active VLWIR HgCdTe base layer and p-n junction are both buried in a single crystalline wider band gap HgCdTe layer. In addition, the architecture has a planar geometry, and the device is configured so that the p layer is doped high while the n layer is doped low, so that most of the depletion region oc-

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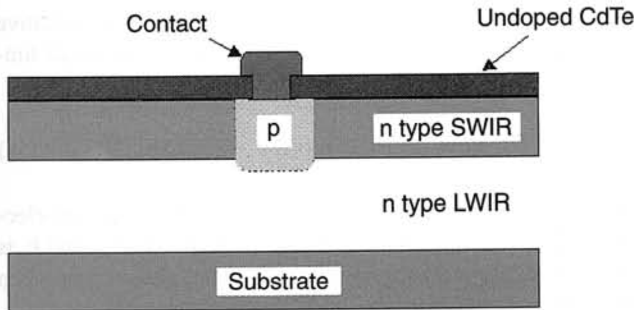


Fig. 2. p on n buried planar heterostructure device architecture.

curs in the n region. Because of the p on n device configuration, even with low n doping of the base layer, the extremely high mobility of electrons compared to that of holes in VLWIR HgCdTe does not result in inversion of the VLWIR layer. Hence this device structure is relatively free from the problems of current leakage and non uniformity of passivation associated with n on p devices, where such inversion occurs.

In contrast, in a conventional p on n mesa device architecture, the portion of the p-n junction comprising the VLWIR layer comes in contact with the surface, thus contributing significantly to leakage currents arising from surface and bulk generation-recombination and tunnelling mechanisms. This mesa architecture thereby necessitates an extremely exacting passivation of excellent quality for diode uniformity and reproducibility.

The passivation problem in the buried planar architecture is alleviated quite elegantly by the wider band gap single crystalline HgCdTe, with nearly the same lattice parameter as that of the base layer. The only region of the p-n junction which comes in contact with the surface in the buried planar heterostructure detector is the one comprising the wider band gap HgCdTe layer, which is much easier to passivate using the conventional ZnS dielectric or CdTe.

This device approach has been demonstrated to yield the highest performance devices in a spectral region ranging from 3 μm to over 20 μm [2,5,6].

2.3.2. Lateral collection photodiode configuration

By reducing the electrical junction area to only a fraction of the optical area, the thermally generated dark currents can be significantly suppressed, while maintaining a high collection efficiency, particularly if the entire region in the pixel optical area is within a diffusion length of the minority carrier (Fig. 3). Although this idea of utilising lateral collection to improve detectivity was first conceived by Holloway [7,8] for PbSnTe, the inability to reduce surface leakage related dark currents made it difficult to achieve the anticipated benefits in that material. In the buried planar heterostructure photodiodes, surface leakage related dark currents are minimised, since the active layer of the junction does not intersect the surface. Hence the lateral collection design is used in conjunction with buried planar heterostructure to obtain maximum detectivity VLWIR HgCdTe detectors.

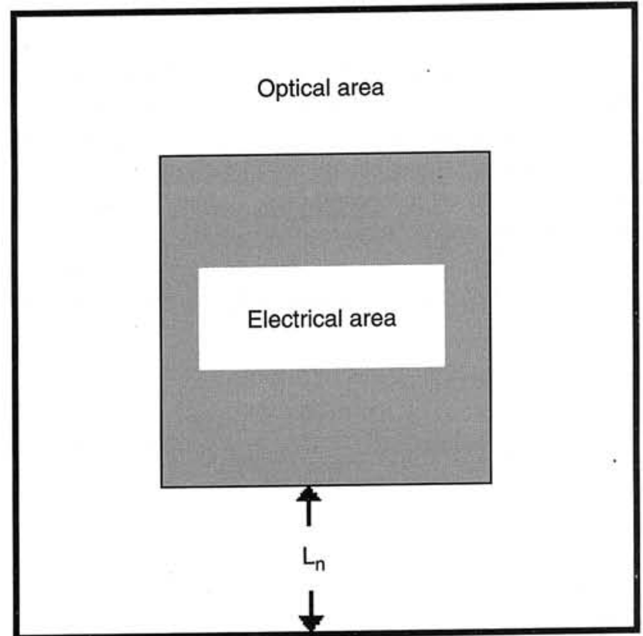


Fig. 3. Schematic of the lateral collection photodiode approach; electrical area is only a fraction of the optical area; diffusion length L_n of holes will allow for the collection of all the carriers photogenerated within the optical area shown.

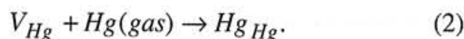
3. p-type doping issues

One of the important issues that plays a major role in the practical realisation of MBE grown VLWIR HgCdTe photodiode arrays exhibiting high performance, uniformity and operability is the formation of p-n junctions at low temperatures so, that the extremely low heterostructure compositional grading resulting from the MBE growth at T < 200°C is retained during the post-growth device fabrication anneals.

Amphoteric behaviour of group V elements in HgCdTe has been unambiguously established irrespective of the method of growth. In bulk and epitaxial HgCdTe grown by LPE, MOCVD and MBE, As and P behave as donors on metal lattice sites, and as acceptors on Te lattice sites [9,11]. It is difficult to get conversion from n type to p type with these dopants at temperatures below 400°C because of the lack of availability of Te vacancies. Dislocation climb which is a thermally assisted process occurring at temperatures of 400°C and higher can generate Schottky pairs consisting of Te vacancies and Hg vacancies as shown below

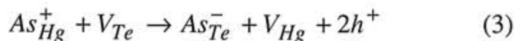
$$0 \rightarrow V_{Hg} + V_{Te}. \quad (1)$$

Once a pair of these defects is created, the Hg vapour quickly fills the Hg vacancies because of the metal rich conditions imposed around the crystal



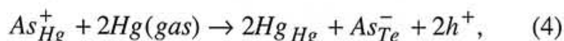
$$V_{Hg} + Hg(gas) \rightarrow Hg_{Hg}. \quad (2)$$

This in turn upsets the Schottky equilibrium and fresh pairs of Schottky defects are created as described by Eq. (1). With the creation of more Te vacancies by the climb process, arsenic atoms occupying metal lattice sites as donors (As^{+}_{Hg}) go on to occupy these vacant Te lattice sites, giving rise to arsenic acceptors (As^{-}_{Te})



$$As^{+}_{Hg} + V_{Te} \rightarrow As^{-}_{Te} + V_{Hg} + 2h^{+} \quad (3)$$

Combining Eqs. (1)–(3) as [(1(+2×(2)) + (3)], we get



$$As^{+}_{Hg} + 2Hg(gas) \rightarrow 2Hg_{Hg} + As^{-}_{Te} + 2h^{+}, \quad (4)$$

to give the final site transfer. Thus, a high concentration of arsenic occupying Te lattice sites, where it acts as an acceptor, builds up in the crystal and the crystal turns p-type and p-n junction forms.

Role of dislocations in the activation of group V acceptors in HgCdTe is evident from the following experimental facts. Single layer LPE HgCdTe material requires annealing at temperatures of at least 500°C under Hg saturated conditions to obtain p type material using As and P as dopants. At temperatures below 500°C, one does not get conversion from n type to p type. However, in heterostructures with group V dopant ion implantation, we could obtain conversion to p type at 450°C. With higher implant energies we could convert the material to p type at temperatures close to 400°C. These observations can be explained by assuming the occurrence of high density of dislocations at the heterostructure interfaces and under higher energy implant conditions. The high dislocation density may also reduce the activation energy for climb and result in the decrease of the group V acceptor dopant activation anneal temperature.

Too high a dislocation density or anneal temperature will result in degradation of signal to noise ratio of these devices. Clearly, a compromise needs to be arrived at between an allowable dislocation density in the HgCdTe heterostructures and the upper limit of temperature at which one can fabricate p-n junctions without giving rise to excessive compositional grading.

4. Origin of Shockley-Read centres in HgCdTe

Although structural defects such as dislocations, stacking faults and twins all act as Shockley-Read centres in limiting the minority carrier lifetime, we shall focus our attention on the point defects.

It is well established that the minority carrier lifetime in p-type $Hg_{1-x}Cd_xTe$ decreases with increase in Hg vacancy concentration [12]. This suggests that either the Hg vacancy or another native defect whose concentration is coupled to that of the Hg vacancy acts as the Shockley-Read centre in limiting the carrier lifetime below the radiative limit. Expression for the single level Shockley-Read limited minority carrier lifetime is given by

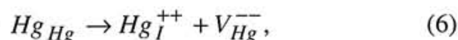
$$\tau_e = \tau_{no}[1 + \exp(E_F - E_t)/kT], \quad (5a)$$

where $\tau_{no} = 1/C_nN_t$, C_n is the capture cross section of electrons, N_t is the trap density, E_F is the Fermi energy and E_t is the trap energy. Examination of Eq. (5a) shows that when $E_F = E_t$,

$$t_e = 2t_{no} = 2/C_nN_t. \quad (5b)$$

This is independent of temperature, if all the traps are ionised. This means that in a temperature range where Hg vacancies are completely ionised and the Fermi level is pinned at the energy level of the Hg vacancies, a temperature independent carrier lifetime should result. However. experimental results in the literature indicate otherwise. This leads us to infer that Hg vacancies themselves are not the traps, but another native defect whose concentration is tied to the concentration of Hg vacancies are the traps. Indeed, examination of the way in which the crystals have been prepared to vary the Hg vacancy concentration, lends support to our hypothesis. $Hg_{1-x}Cd_xTe$ samples with different concentrations of Hg vacancies are all prepared under Hg saturated conditions with the crystals exposed to the highest temperature showing the highest Hg vacancy concentrations. From this, it appears that Hg interstitials, whose concentration tracks the concentration of Hg vacancies are the Shockley-Read centres.

The concentration of Hg interstitials is orders of magnitude lower than that of the Fermi level pinning Hg vacancies. This explains the temperature dependence of the minority carrier lifetime. The concentrations of Hg interstitials and Hg vacancies are coupled via the Frenkel reaction



$$Hg_{Hg} \rightarrow Hg^{++}_{I} + V^{--}_{Hg}, \quad (6)$$

where a neutral Hg atom occupying a Hg lattice site goes into an interstitial site acting as a double donor, leaving behind a vacant lattice site of Hg which acts as a double acceptor. The mass action constant for the Frenkel reaction described by Eq. (6) is given by

$$K^{++}_{F} = [Hg^{++}_{I}][V^{--}_{Hg}]/[Hg_{Hg}], \quad (7)$$

where [ ] indicates site fraction, equal to number of defects per cm^3/total number of lattice sites per cm^3. $[Hg_{Hg}]$ is approximately unity since most of the Hg lattice sites are occupied. Therefore

$$K^{++}_{F} = [Hg^{++}_{I}][V^{--}_{Hg}] \quad (8)$$

It should be noted that K^{++}_{F} is dependent on temperature and not on partial pressure of Hg. Thus, at a fixed tem-

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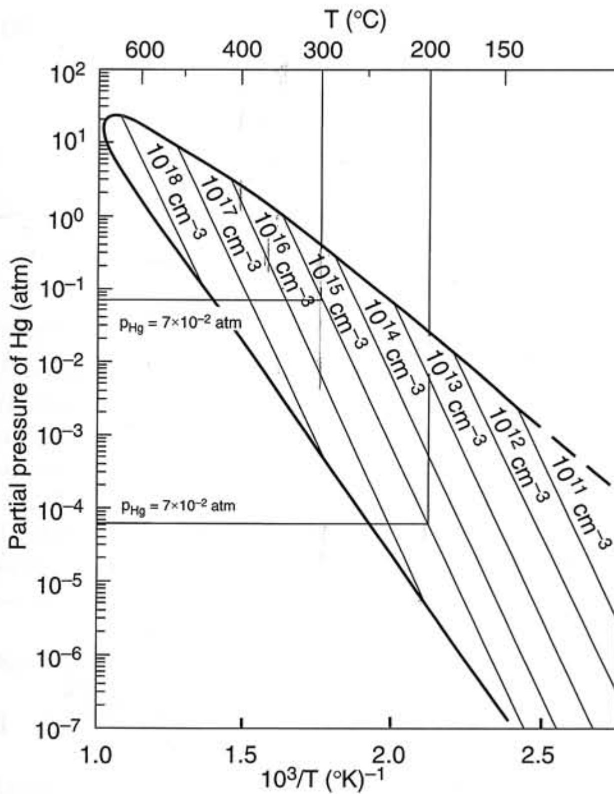

Fig. 4. Iso-hole concentration plot of HgCdTe indicating possibility of obtaining identical hole concentration for anneals at two different temperatures and partial pressure of Hg.

perature the value of K_F^{++} is fixed, and when the partial pressure of Hg is varied, $[Hg_I^{++}]$ and $[V_{Hg}^{--}]$ change in such a way that their product remains constant. Thus, in samples prepared at higher temperatures under Hg saturated conditions, K_F^{++}, $[Hg_I^{++}]$ and $[V_{Hg}^{--}]$ will all be higher. Based on this argument, it is easy to see that $Hg_{1-x}Cd_xTe$ samples could be prepared at different temperatures and partial pressures of Hg with the same Hg vacancy concentration, but with different Hg interstitial concentrations; samples prepared at the lowest temperature and lowest Hg pressure will have the lowest concentration of Hg interstitials. For example, a Hg vacancy concentration of 10^{15} cm^{-3} can be obtained with an anneal at 300°C and 7×10^{-2} atmosphere Hg partial pressure, or with an anneal at 200°C and 6×10^{-5} atmosphere Hg partial pressure (Fig. 4); however, the concentration of Hg interstitial donors with the lower temperature anneal will be smaller. If Hg interstitials are the Shockley-Read centres, the minority carrier lifetime in samples prepared at the lower temperature and lower Hg pressure will be higher than that in samples prepared at higher temperatures, even though the Hg vacancy concentration in both samples is the same.

5. Summary

Issues relating to VLWIR HgCdTe detectors with high detectivity, and uniformity have been reviewed. Dislocation climb assisted secondary reaction resulting in a pair of Schottky defects is proposed to be the operative mechanism entailed in the transfer of the group V elements from Hg lattice sites onto Te lattice sites and giving rise to n to p conversion. Origin of native point defect related Shockley-Read centres, and methods to minimise them have been examined.

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