On the electrical properties of Si-doped InGaP layers grown by low pressure-metalorganic vapor phase epitaxy

Roberto Jakomin a,1, Antonella Parisini a,3, Luciano Tarricone a, Massimo Longo a,2, Beatrice Fraboni b, Salvatore Vantaggio a

a CNISM, Dipartimento di Fisica, Università di Parma, Viale G.P. Usberti, 7/A, 43100, Parma, Italy
b Dipartimento di Fisica, Università di Bologna, Viale Berti Pichat 6/2, 40127 Bologna, Italy

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ABSTRACT

The electrical properties of undoped and silicon doped InGaP layers grown lattice matched on GaAs by low pressure metal–organic vapor phase epitaxy were investigated under different growth conditions. The possible presence of superlattice ordering was excluded by photoluminescence analysis. Undoped layers exhibited a background p-type contamination of the order of $10^{16}$ cm$^{-3}$; the role of possible carbon contamination is discussed. Capacitance–voltage and Hall investigation of Si-doped n-type layers evidenced a room temperature free electron density linearly increasing from $3.6 \times 10^{16}$ to $6 \times 10^{16}$ cm$^{-3}$ as a function of the Si precursor flow. The corresponding electron mobilities decreased from 1800 to 483 cm$^2$/V s. At lower temperatures, the conductivity and mobility of the n-doped samples showed a metallic like behavior, in some cases with values not consistent with a simple electronic transport into the conduction band, suggesting the presence of an additional parallel transport channel. Five main electron traps were identified by deep level transient spectroscopy and among them, two traps resulted to be dominant, one turned out to play a major role in the bulk, probably associated to the Si doping, and the other was active near the InGaP surface, ascribable to P-related defects.

1. Introduction

InGaP lattice matched to GaAs is considered a good alternative to the AlGaaS/GaAs system for the achievement of Al-free high quality devices [1,2]. High performance opto-electronic, photonic and high reliability electronic devices based on InGaP/GaAs have been fabricated [3–5]. However, lattice matched InGaP-on-GaAs layers can show, depending on the growth conditions and on the orientation of the substrate, a CuPtB-type ordering [6–8], which significantly influences their electrical and optical properties [9,10]. Optical measurements, in particular photoluminescence spectroscopy (PL), have been extensively employed to investigate these effects [11,12 and references therein]. Also the possible coexistence of such effects with strain has been carefully accounted for, to correctly evaluate the ordering parameter [13,14]. In addition, the InGaP/GaAs interface resulted to be problematic owing to intermixing effects [15] and the formation of charged dipoles, generating interface states and influencing the band offsets [16 and references therein]. Charge accumulation at such interface, induced by unknown defects and giving rise to a channel of parallel conduction, has even been pointed out by some authors [17,18]. Last, unstable surface defects, due, for example, to P desorption, can degrade the InGaP surface and cause irreproducible results [19].

It is therefore not surprising that a reliable characterization of InGaP-on-GaAs layers must take into account all the above aspects, and that the growth conditions have a significant influence on the electrically active defects. However, very little is generally reported about background impurities. Thus, in the present work we study the electrical properties of both undoped and Si–InGaP epitaxial layers, by focusing on the transport abnormalities induced by electrically active unintentional defects.

The InGaP layers were grown on GaAs by metal–organic vapor phase epitaxy (MOVPE) using the alternative metal–organic (MO) precursors tertiarybutylarsine (TBA) and tertiarybutylphosphine (TBP). In the paper, we will firstly discuss the assessment of the MOVPE growth of nominally undoped samples, to minimize ordering effects and lattice mismatch. Then electrical transport data of n-type InGaP layers will be analyzed. Despite mobility values at the state of the art, temperature dependent Hall effect measurements and the investigation of the deep levels reveal the presence of undesired defects at the surface, at the interface and into the bulk of the layers. These defects will be the object of discussion in the paper and the unusual p-type conductivity of the unintentionally doped InGaP will be commented on.
2. Experimental details

The lattice matched In_{0.49}Ga_{0.51}P layers were grown on exact <100> ± 0.1° oriented 2 in. GaAs wafers by a horizontal, low pressure MOVPE reactor, equipped with a rotating substrate holder (gas foil rotation) to limit in-plane compositional non-homogeneities during the growth. This is particularly important, considering the large difference between the lattice constants of InP and GaP. n⁺- Type and semi-insulating substrates were used for capacitance and transport measurements, respectively. Metal–organic precursors, trimethylgallium (TMGa), trimethylindium, TBAIs and TBP, were used as sources for Ga, In, As, and P, respectively, while n-doping was achieved by using hydrogen-diluted disilane (SiH₄). Such precursors were employed to potentially guarantee a lower incorporation of unintentional impurities into the epitaxial layer, owing to their lower dissociation temperatures in comparison to the common hydrides, arsine (AsH₃) and phosphine (PH₃). The TMGa partial pressure was kept around 0.5 Pa. Unintentionally p-doped layers were obtained at a V/III precursor vapor pressure ratio (\( \Phi_{\text{V/III}} \)) ranging from 9 to 23, and by varying the growth temperature (\( T_\text{G} \)) between 550 and 600 °C. n-Type layers were grown at \( T_\text{G} = 600 °C \) and a molar fraction ratio between disilane and group III precursors (Si/III) from 10⁻³ to 10⁻³, with \( \Phi_{\text{V/III}} \) from 14 to 16. For every growth run the total pressure was 5 \( \times 10³ \) Pa and the growth rate \( \approx 2.2 \mu \text{m/h} \). The thickness of the InGaP layers was in the range of 0.5–5 µm, the thicker ones being grown to investigate the electrical properties in less doped InGaP. A nominally undoped GaAs buffer layer, 300 nm-thick, was systematically inserted between the GaAs substrates and the InGaP epilayer to improve the InGaP/GaAs interface quality. Considering the previously measured high resistivity of our MOVPE grown GaAs (p-type background impurity density lower than \( 10¹⁴ \text{cm}⁻³ \)) [20], such a buffer layer is not expected to influence the electrical properties of the overgrown InGaP layer even when the latter is unintentionally doped. No GaAsP interlayer was inserted at the InGaP-on-GaAs interface, since we demonstrated that such an interlayer degrades that interface [15]. Finally, a thin p⁻⁻GaAs cap layer (50 nm) was generally deposited on top of the nominally undoped samples to improve the ohmic contact properties. This cap layer was then selectively removed from the contact free area before performing the electrical measurements, by using the solution H₃PO₄:H₂O₂:H₂O (3:1:50).

The electrical properties were investigated by Hall effect measurements, carried out from 300 to 10 K, on square shaped samples by the van der Pauw method. Triangularly shaped Au–Zn ohmic contacts were deposited at the corners of nominally undoped InGaP samples and subsequently annealed at 430 °C for 60 s under H₂ atmosphere. An Au–Ge alloy was used for Si doped samples. Typical measurement uncertainty was 10%. Standard corrections due to contact size and to carrier depleted regions at the surface and the interface with substrate were evaluated before obtaining the electrical data. In particular, the free carrier density was considered reliable only when its value was self-consistent with the doping-dependent thickness of the depleted regions. Electrical properties were also investigated by capacitance–voltage (C–V) and deep level transient spectroscopy (DLTS) measurements. To this aim, large area Au–Ge ohmic contacts were prepared on the backside of the n⁻⁻GaAs substrate and Au Schottky barriers were made on top of the InGaP layers by standard photolithography (vertical configuration). In some cases the C–V measurements were also performed in samples grown on semi-insulating substrates, by using a planar ring-shaped ohmic metatllization formed around a Schottky dot contact (planar geometry). The DLTS system operated at a sample heating rate of 0.2 K/s, a temperature range from 78 to 400 K and an emission rate from 5 to 2 \( \times 10³ \text{s}⁻¹ \). Photoluminescence (PL) spectra were taken through a Fast Fourier Transform Spectrometer exploiting the 488 nm line of a 5 W Ar⁺ ion laser. The spectra were recorded as a function of the temperature (10–300 K) and at different power densities. A spectral resolution of 0.1 meV was typically used in the investigated spectral range.

3. Results and discussion

3.1. Optimization of the undoped InGaP

It is well known that in unintentionally doped InGaP layers ordering effects induce a band gap reduction (BGR) and a split of the fourfold degenerate heavy–light hole valence band (VB) maximum [6,7]. Indeed, a maximum BGR value of about 490 meV has been predicted [7], and confirmed by an experimental value of 471 meV [11]. A mean order parameter \( \eta \) has been then defined, dependent on the fraction of the observed BGR with respect to its maximum value [7]. However, when the alloy band gap dependence on the In molar fraction deviates from the behavior predicted for unstrained, not ordered InGa₁₋ₓP, it is important to correctly estimate the effects due to both CuPt-B ordering and composition-induced lattice strain. Recently, Novak et al. [14] dealt with this problem, evaluating the final band gap behavior due to the interaction between the ordering effect and the mismatch-induced strain. Considering the above aspects, in this work the optimal growth conditions for nominally undoped InGaP layers were selected by plotting the band-to-band PL transition energy at \( T = 10 \text{ K} \) versus the In molar fraction estimated by X-ray diffraction (not shown here, see Ref. [20] for experimental details). The best compromise minimizing both ordering and strain effects was obtained for \( T_\text{G} = 600 °C \), \( \Phi_{\text{V/III}} = 15.63 \), and \( x = 0.485 \). In this case, the lattice mismatch \( (\Delta a/a = \text{variation of InGaP lattice parameter with respect to GaAs}) \) across the InGaP layer was measured by XRD and resulted in the range of \( 1.7 \times 10⁻⁵ < \Delta a/a < 2.5 \times 10⁻⁵ \), corresponding to an average In content of \( x_{\text{av}} = 0.485 ± 0.001 \) and indicating the lattice match condition.

The PL spectrum in Fig. 1 refers to the optimized undoped sample US (see Table 1) and shows a main peak located at \( E = 1.972 \text{ eV} \), corresponding to a negligible BGR. For this sample we investigated more deeply the possible presence of CuPt-B ordering. To support the absence of ordering in such sample, we analyzed the dependence of the main PL emission band on temperature and exciting photon flux. Generally, at temperatures in the range of a few tenths of K, a blue-shift of the main emission band, with increasing \( T \) or light power, is assumed to be a reliable proof of the presence of CuPt-B ordering [12,21]. Instead we observed only a red-shift, whose magnitude as a function of the laser beam power is evidenced in the inset of Fig. 1. Such red shift is fully ascribable to the thermal shrinkage of the energy gap. Thus the subsequent depositions of n-doped samples were performed by using the same optimal growth parameters.

![Fig. 1. T = 10 K: PL emission spectrum detected on an optimized, lattice matched, nominally undoped InGaP/GaAs heterostructure (sample US). The inset shows the red shift of the PL peak energy as a function of the exciting laser power at T = 12 K.](image-url)
A similar analysis on the doped samples would appear redundant, since, as previously demonstrated [6, 12], a high doping level limits the formation of the CuPt-B ordering. However, we also tested the doped samples, obtaining the confirmation that the doped InGaP samples here investigated do not exhibit any ordering effect.

3.2. Conductivity and mobility measurements

3.2.1. Undoped InGaP

Epitaxial InGaP layers generally show intrinsic n-type conductivity, almost independent of the growth temperature [22–27], with rare exceptions, where background acceptors are reported [27]. Notably, in our nominally undoped samples, a p-type background doping of a few $10^{16}$ cm$^{-3}$ was obtained by Hall effect measurements. These densities were also confirmed by C–V measurements, which showed nearly flat profiles, compatible with a uniform background impurity density (see Table 1). To investigate on the nature of this unusual background doping, we extended the Hall measurements from fractions of Kelvin degree to 600 K. In Fig. 2a and b the typical temperature dependences of the hole density and mobility are respectively shown for a representative undoped InGaP layer (sample U44 of Table 1). At about 400 K a change of conductivity was reproducibly observed; by lowering the temperature, the hole density decreases, owing to the hole freezing into acceptor impurities, with a slope consistent with a thermal ionization energy around 40 meV, as expected for an effective mass state ($E_{\text{mea}}$ ~57 meV [28]). Therefore, these data appeared dominated by the occupancy variation of a background shallow acceptor. However, a hump appears in the Hall data at a temperature of about $T = 150$ K, evidenced in Fig. 2 by arrows. The same feature was visible in most of the unintentionally doped samples, independently of their thickness. In addition, below 150 K, an unusually high value (around 5, nearly temperature independent) of the van der Pauw corrective function was generally required to obtain the transport data, especially in thin layers, as for the appearance of inhomogeneities. Weak variations of the Hall data with aging were also detected, but no magneto-resistance or photoconductivity effects, nor hysteretic behaviors. Most of the surface/interface defects reported for nominally undoped InGaP are electron traps, generally present in layers having n-type conductivity [19, 29–32]. On the other hand, hole traps have been widely investigated in irradiated p-type layers [33], but only a main hole trap, 0.34 eV above the VB edge, has been reported in “as grown” p-type layers [34]. Since the hole density in our undoped InGaP is basically independent of the layer thickness, we believe that the background p-type conductivity cannot be ascribed to a spatially localized trap at the surface of the layer, or at the interface with the substrate. Further, we have to exclude the possibility that such background doping can be introduced by a contamination of the MOVPE environment or the metal–organic sources. In fact, nominally undoped GaAs and InP, grown in the same MOVPE reactor and with the same metal–organic sources, were characterized by a low impurity incorporation [20, 35]. We conclude that the defect which causes the p-type conductivity in our samples might be related to C incorporation, which can introduce shallow acceptor levels in InGaP [36]. As a matter of fact, a high C concentration in InGaP has also been shown to induce compensation, owing to the amphoteric behavior of C [23] and we recall the possibility to easily obtain a high p-type C self-doping in GaAs layers by properly varying the growth conditions [20].

3.2.2. Si-doped InGaP

In our Si-doped InGaP layers, a nearly linear increase of the $n_{\text{hi}}$ electron density versus the Si$_2$H$_6$ vapor pressure was observed from $3.6 \times 10^{18}$ to $6 \times 10^{18}$ cm$^{-3}$. The RT Hall density values were also confirmed by C–V measurements (Fig. 3). The corresponding $n_{\text{hi}}$ at RT ranges from 1800 to 483 cm$^3$/V s, exhibiting a dependence on the electron density which approaches a $n_{\text{hi}}^{1/3}$ power law. These results are in agreement with the typical values reported for InGaP grown by MOVPE, which generally exhibit higher $n_{\text{hi}}$ than those obtained

### Table 1
Electrical data for nominally undoped and Si-doped samples discussed in the work. The Si/III value is referred to the molar fraction ratio between disilane and group III precursors in the vapor phase. In the Si-InGaP layers grown on semi-insulating substrates (acronym S.I. in the IV column) the C–V measurements were taken on a second piece of the samples used for Hall measurements by using a planar geometry (see the experimental section).

<table>
<thead>
<tr>
<th>Sample</th>
<th>Si/III ratio</th>
<th>Growth temp. (°C)</th>
<th>Type of substrate</th>
<th>Layer thickness (nm)</th>
<th>RT carrier density (cm$^{-3}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>U5</td>
<td>0</td>
<td>600</td>
<td>S.I.</td>
<td>550</td>
<td>$3.4 \times 10^{16}$ (Hall)</td>
</tr>
<tr>
<td>U44</td>
<td>0</td>
<td>580</td>
<td>S.I.</td>
<td>5000</td>
<td>$6 \times 10^{15}$ (CV)</td>
</tr>
<tr>
<td>D1</td>
<td>$1.80 \times 10^{-8}$</td>
<td>600</td>
<td>S.I.</td>
<td>1000</td>
<td>$3.6 \times 10^{16}$ (Hall)</td>
</tr>
<tr>
<td>D2</td>
<td>$1.05 \times 10^{-4}$</td>
<td>600</td>
<td>S.I.</td>
<td>1000</td>
<td>$3.8 \times 10^{17}$ (Hall)</td>
</tr>
<tr>
<td>D4</td>
<td>$1.26 \times 10^{-3}$</td>
<td>600</td>
<td>S.L.</td>
<td>500</td>
<td>$4.1 \times 10^{18}$ (Hall)</td>
</tr>
<tr>
<td>D10</td>
<td>$4.97 \times 10^{-5}$</td>
<td>600</td>
<td>n$^{-}$-Type</td>
<td>1000</td>
<td>$\approx 4 \times 10^{18}$ (CV)</td>
</tr>
</tbody>
</table>

* Depending on the frequency.
by molecular beam epitaxy (MBE) with comparable free electron density [37].

The temperature dependence of the free carrier density, measured by Hall effect, is shown in Fig. 4. \( n_H \) varies very weakly as a function of the temperature. In the less doped sample \((N_d - N_A = 3.6 \times 10^{16} \text{ cm}^{-3} \text{ at RT, lowest curve in Fig. 4a, with } N_d = \text{donor density}, N_A = \text{acceptor density})\) the thermal activation energy for electrons into the CB, estimated by the \( n_H \) data just below RT, results in about 7.5 meV. Such a value is lower than the theoretical predictions for the 1 s level of a shallow donor in InGaP [28], consistently with an intermediate doping level. Indeed, this value decreases to 4.6 meV by further increasing the impurity density \((N_d - N_A = 1.7 \times 10^{17} \text{ cm}^{-3} \text{ at RT, middle curve in Fig. 4a), becoming negligible at over a few } 10^{18} \text{ cm}^{-3} \) (highest curve in Fig. 4a). Moreover, in all samples, \( n_H \) and \( \mu_H \) become temperature independent, as it usually happens for a degenerate electron gas, when the temperature is decreased (Fig. 4a and b).

This effect occurs even in the less doped sample (D1), where the electron density is comparable with the critical value required for the Mott transition (about \( 8 \times 10^{16} \text{ cm}^{-3} \) for the n-type InGaP by assuming an effective electron mass \( m = 0.109 \text{ } m_0 \) and a static relative dielectric constant \( \varepsilon = 11.74 \)). However, in such a case, the self-consistent fit of \( n_H \) and \( \mu_H \) versus temperature fails when the transport is described by a single valley model. Preliminary low-field magneto-resistance measurements versus temperature on the D1 sample revealed a peak at about 60 K (Fig. 5), suggesting the presence of a parallel conduction channel. This channel seems to play a dominant role at low temperatures and can account for the high mobility values (upper curve in Fig. 4b), characteristic of a two dimensional electron gas (2DEG), which may be formed by charge accumulation at the InGaP surface or at the InGaP/GaAs interface. The presence of mixed conduction effects could be the reason for the discrepancy between the C–V and Hall density values of Table 1.

On the contrary, in more heavily doped samples, the doping level overcomes the critical Mott’s density and no appreciable magneto-resistance signal was detected in the investigated temperature range. Thus, for these samples it remains unclear whether the metallic-like conductivity behavior is related to a parallel conduction mechanism or simply to a degenerate 3D electron gas. Moreover, it should be remarked that no significant persistent photoconductivity effect (PPC) was observed in such samples.

A charge accumulation at the n-InGaP/GaAs heterointerface has been already observed in samples grown by As and P hydrides [38,39]. For this reason, we suggest a similar explanation also for the parallel conduction channel hypothesized for our samples. We remind that the formation of a 2DEG at the InGaP/GaAs interface has been evidenced even in structures having an n-type background conductivity [17,18,40]. In those samples, unexplained persistent photoconductivity effects also appeared at low temperature. Differently, in our unintentionally doped samples no evidence of 2DEG was found even when a 5 nm single GaAs quantum well was embedded between two InGaP cladding barriers. 

Fig. 3. RT carrier concentration data obtained by C–V measurements on different Si–InGaP samples as a function of the Si/III vapor pressure ratio. The triangular symbols connected by a short vertical dotted line referred to the same sample (D3, see Table 1).

Fig. 4. Hall electron density (a) and Hall electron mobility (b) versus temperature for differently Si-doped InGaP layers: ● D1 (Si/III: \( 1.8 \times 10^{-5} \)), ○ D3 (Si/III: \( 1.05 \times 10^{-4} \)), ■ D4 (Si/III: \( 1.26 \times 10^{-3} \)).

Fig. 5. Physical magneto-resistance \( \Delta \rho/\rho \) measured in sample D1 as a function of the temperature: ● in dark by heating the sample; ○ in dark by cooling the sample; ■ under light, \( \rho(B) \) is the resistance measured under a magnetic field, \( B = 8 \text{ kg} \).
3.3. C-V and DLTS measurements

C-V measurements were performed at RT on several n-type samples, either through planar or vertical geometry. Data scattering was observed at low Si/III ratios, as shown in Fig. 3; moreover, a dependence of the C-V characteristics on the frequency of the a.c. signal modulating the reverse bias and/or a deviation from flat profiles expected for a homogeneous net doping level, were sometimes observed. However, no appreciable differences in the C-V behaviors were observed if a thin p+ -GaAs cap layer was deposited on the Si-InGaP and then removed before forming the Schottky barrier (samples not discussed in this work).

Detailed capacitive measurements were carried out on sample D10 (see Table 1) in order to investigate the spatial charge and deep levels in the Si-doped InGaP. The C-V density profiles, recorded for different modulation frequencies of the applied reverse bias, are reported in Fig. 6. The C-V profile gives the depth dependence of the net electrically-active impurity density from the surface. For an n-type material it can be written as $N(x) = (N_d - N_a) + N_{val}(x) - N_{ac}(x)$, by taking into account the occupancy of donor- and acceptor-like traps, $N_{val}(x)$ and $N_{ac}(x)$, and by supposing a uniform distribution of shallow impurities, $N_d$ and $N_a$. In Fig. 6 shows that beyond a depth of about 200 nm from the surface, the net donor density is almost constant ($\approx 10^{16}$ cm$^{-3}$) and independent of the modulation frequency of the bias. The peculiarity of the present result is given by the frequency dependent enhancement of the net donor density profile near the surface. It highlights a relevant contribution of the surface traps to the dynamic response. Such contribution adds to that of the bulk deep levels, whose activation energies measured from the bottom of the CB have been identified in DLTS analyses of InGaP epilayers by Yoon et al. The peak broadening has been attributed to a distribution of traps resulting from the random nature of alloy bonding. The authors concluded that the dominant traps in the bulk are B and C. Traps B and C convolve to form a broad peak and appear uniformly distributed in the layer.

Deep levels with activation energies similar to those of traps B and C have been previously reported for Si doped InGaP grown by MBE and associated to donor-related defects [32,41,42]. In particular, some authors ascribed these defects to a DX-like center. Another similar structure having an activation energy of 0.39 eV below the CB has been identified in DLTS analyses of InGaP epilayers by Yoon et al. The peak broadening has been attributed to a distribution of traps resulting from the random nature of alloy bonding. The authors excluded a possible identification of such traps with a DX center [43]. A deep level similar to our trap D, localized near the surface, has been reported by Huang et al. in epilayers grown under low P$_2$ pressure (V/III ratio of 4) and attributed to phosphorous vacancies and allowed us to change the depth of the investigated volume. We can conclude that the dominant traps in the bulk are B and E. Traps B and C convolve to form a broad peak and appear uniformly distributed in the layer. Trap E is more pronounced closer to the surface, where trap D also clearly appears, suggesting a surface localization of this defect.

Deep levels with activation energies similar to those of traps B and C have been previously reported for Si doped InGaP grown by MBE and associated to donor-related defects [32,41,42]. In particular, some authors ascribed these defects to a DX-like center. Another similar structure having an activation energy of 0.39 eV below the CB has been identified in DLTS analyses of InGaP epilayers by Yoon et al. The peak broadening has been attributed to a distribution of traps resulting from the random nature of alloy bonding. The authors excluded a possible identification of such traps with a DX center [43]. A deep level similar to our trap D, localized near the surface, has been reported by Huang et al. in epilayers grown under low P$_2$ pressure (V/III ratio of 4) and attributed to phosphorous vacancies and

![Fig. 6. Net donor concentration profile obtained in sample D10 by C-V measurements performed at different a.c. sampling frequencies, where $N(x) = (N_d - N_a) - N_{ac}(x)$, $N_d$ = shallow donor concentration, $N_a$ = background acceptor concentration and $N_t$ = deep electron trap concentration.](image)

![Fig. 7. Arrhenius plot related to the main traps detected by DLTS in sample D10. The activation energy is calculated with respect to the conduction band.](image)

![Fig. 8. DLTS spectra recorded in sample D10 at different reverse ($V_{bias}$) and filling ($V_{fill}$) biases: curve (1): $V_{bias} = -10.4$ V and $V_{fill} = 7$ V; curve (2): $V_{bias} = -10.4$ V and $V_{fill} = 3.8$ V; curve (3): $V_{bias} = -3$ V and $V_{fill} = 5.3$ V.](image)
related complexes [44]. According to Paloura et al., the lattice mismatch can also induce a trap with similar features [32]. Sometimes the latter two traps – related to phosphorous vacancies and to lattice mismatch – can appear convoluted; however, Zhu et al. resolved the two contributions by high resolution DLTS [45]. Finally, InGaP epilayers grown under high P$_2$ pressure (V/III ratio between 10 and 17) have been reported with a high deep level concentration, with an activation energy of 0.82 eV, similar to our trap E [41]; a dependence of this deep level on the annealing suggested its attribution to phosphorous antisites [41]. However, our trap E can neither be identified with the deep level of comparable activation energy revealed by Paloura et al. [32], because it is suppressed by Si doping, nor with the traps discussed in Refs. [29–31], which are localized at the GaAs/InGaP interface, in proximity to extended defects. Noteworthy, no electron trap similar to our deep level A has been reported in the literature. Moreover, our investigation did not evidence the electron trap at 0.26 eV below the CB, which is revealed by some authors and attributed to P deficiency, this result being consistent with the use of high V–III ratio [41,44]. In Table 2 we give an overview of the traps mentioned above and in Section 3.2.1, and the possible identification of some of them with the main traps detected in our n-type samples. From this comparison, our main bulk traps B and C may be related to the Si doping. However, the absence of any PPC effect contrasts with the attribution of the defects to a DX like trap. The identification of an E bulk trap seems more appropriate. In fact, it can be associated to an antisite defect of P, favored by the high V/III ratios used. It is worth considering that both P-vacancy and P-antisite defects may be present in sample D10, respectively as a result of the absence of a cap-layer and of the TBP overpressure present during the growth. Thus, an enhancement of the E trap at the surface is justified, while the attribution of the deep level D to P-vacancy related defects is reliable and can explain the surface localization of this trap. To suggest a final remark, we could also suppose that either trap E or D plays a role in giving the peculiar feature exhibited by the C–V profile (Fig. 6), being both detected at the surface.

4. Conclusions

The electrical properties of InGaP/GaAs layers were studied, in samples grown by MOVPE employing the TBAs and TBP metal–organic precursors for the V-group elements. The possible presence of a superlattice ordering was excluded by photoluminescence analysis, also in the nominally undoped samples, which exhibit an unusual acceptor background density of $10^{16}$ cm$^{-3}$. The origin of such p-type conductivity remains not fully explained, although C incorporation appears most probably responsible of such behavior. In the n-type layers, the RT free electron density linearly increases from $3.6 \times 10^{16}$ to $6 \times 10^{18}$ cm$^{-3}$ as a function of the Si/III ratio, while the electron mobility decreases from 1800 to 483 cm$^2$/V s. All samples exhibited a low temperature metallic-like behavior, consistent with the existence of a parallel conduction channel, tentatively attributed to charge accumulation at the interface with the substrate. However, the mobility data in the Si-doped InGaP layers are comparable with the best results reported in the literature. The activation energy and capture cross section of five main electron traps were identified; two of them resulted to be preferentially located near the surface. The comparison with data reported in the literature suggested their identification as P-related defects. The main bulk electron traps may be associated to the Si-doping, although the absence of any persistent photoconductivity effect discourages the attribution of such deep level to a DX center.

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