Carbon Doping of InAlAs Layers Grown by Metalorganic Vapor Phase Epitaxy

M. L. P. Ribeiro, B. Yavich[*], C. V. B. Tribuzy, and P. L. Souza

Laboratório de Semicondutores, Centro de Estudos em Telecomunicações,
Pontifícia Universidade Católica do Rio de Janeiro,
Rua Marquês de São Vicente 225, Rio de Janeiro, 22453-900 RJ, Brazil

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Carbon doped AlInAs layers grown by low pressure metalorganic vapor phase epitaxy are investigated as a function of growth temperature. Photoluminescence spectra show a significant drop in peak intensity as the growth temperature is reduced, due to the incorporation of non-radiative defects. It is shown that the C doping is not related to the deterioration of the optical properties of the layers. On the other hand, both the net hole concentration and the conductivity increase as the growth temperature is reduced due to a more efficient C incorporation and a lower incorporation of the passivating H atoms.

I Introduction

InAlAs doped with acceptors has many applications in electronic and optoelectronic devices. The *p*-type layer of a *pin* amplitude modulator structure, when InGaAs/InAlAs multiple quantum well (MQW) form the intrinsic region, is a *p*-type InAlAs layer. Normally, Zn is used as the *p*-type dopant and gives rise to excellent quality layers and high controllable doping levels. The disadvantage of this impurity is its large diffusivity [1]. Large diffusion coefficients imply in including a thick intrinsic undoped InAlAs layer between the MQW region and the doped layer to avoid Zn diffusion into the MQW region. This leads to a thicker total intrinsic region and therefore larger applied reverse voltages. Moreover, if one intends to use *p*-type delta-doped layers in the MQW structures [2] one should avoid Zn because the atomic diffusion will give rise to broad doping profiles. The alternative would be to use C, which is an acceptor by substituting As atoms.

Carbon has recently received much of attention as an intentional *p*-type dopant for InAlAs grown by metalorganic vapor phase epitaxy (MOVPE) due to its low diffusion coefficient [3] and high achievable doping level (over 10^{19} cm^{-3}), which is attractive for fabricating a range of high-speed electronic and optoeletronic devices [4]. However, to obtain high doping levels one should use low growth temperatures (T_g), which, on the other hand, are undesirable because they favor defect incorporation. Therefore, knowledge of the limitations of C doping of InAlAs is fundamental for device design.

In this work we present results of a systematic investigation of transport and optical properties of C doped InAlAs layers grown at different temperatures. It is shown that for growth temperatures at which the transport properties of the layers are adequate, the optical characteristics are unappropriate.

II Experimental details

The lattice-matched InAlAs epitaxial layers were grown in a horizontal MOVPE Aixtron 200 reactor at 100 mbar at temperatures between 530°C and 650°C. Hydrogen was used as the carrier gas. The precursors are trimethylgallium (TMGa), trimethylaluminium (TMAl), arsine (AsH_3), tetrabromide (CBr_4) for Ga, Al, As and C respectively. Hall measurements were carried out using a Bio-Rad HL 5500 fixed magnetic field system in the standard Van der Pauw geometry at 300 K. X-ray measurements were made with a Bede Scientific QC2a x-ray diffractometer to determine the alloy composition. Secondary ion mass spectrometry (SIMS) was used to evaluate the C atomic concentration, [C]. The photoluminescence (PL) measurements were performed with the 514 nm line of an Ar$^+$ laser for excitation. The signal was dispersed by a 250 mm monochromator and detected by a nitrogen cooled Ge photodetector.

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[*]with the A. F. Ioffe Physico-Technical Institute, Saint-Petersburg, Russia.

III Results and discussion

C doped InAlAs layers were grown with Tg varying between 530°C and 650°C. The growth temperature normally used for InAlAs is around 630 °C. However, we find that if such high temperatures are used, no matter how high the IV/V flux ratio is, the layers are n-type and not p-type. Three different factors may contribute to this effect. First, C is an amphoteric dopant and its probability of occupying a site III or V is most likely temperature dependent. However, such an effect should be very weak [5] Second, it is well known that low T_g leads to a significant incorporation of defects, which can act as hole traps or compensating species. Finally, H atoms are incorporated with the C atoms, passivating them. Heat-treatments can be used to remove the H atoms and re-establish the C electrical activity.

Fig. 1 shows the two dimensional net free charge concentration, $[c]$ as a function of T_g for samples grown using the same CBr_4 flux. It is clear that the p doping level increases as T_g is reduced. One can also see in Fig. 1 that for T_g ¿ 575°C, the samples are n-type. After growth they were subjected to a heat-treatment for 15 minutes under forming gas flow. In all cases, the free hole concentration, $[p]$, increased with the annealing temperature, as shown in Figs 2a and 2b for samples grown at 650°C and 530°C, respectively, in agreement with results published by other authors[6, 7].

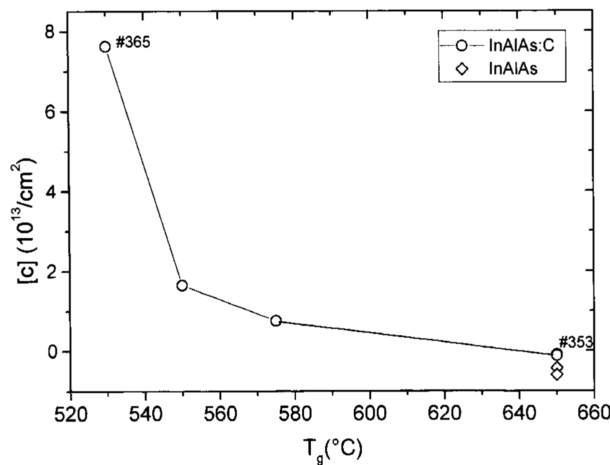

Figure 1. Net free carrier concentration as a function of growth temperature. The circles refer doped material, while the diamonds correspond to undoped layers.

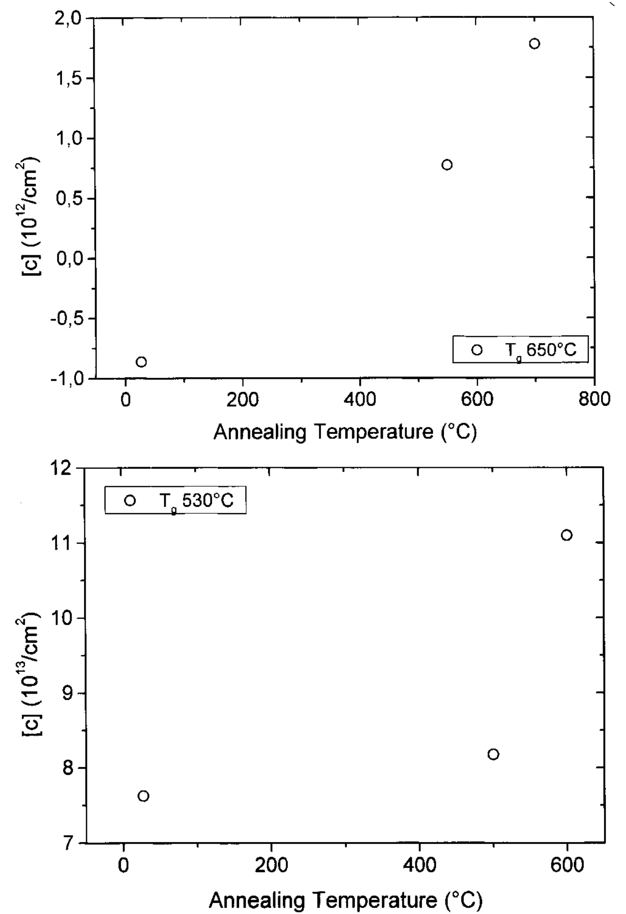

Figure 2. Net free carrier concentration as a function of anneling temperature for samples grown at a) 650°C and at b) 530°C.

SIMS measurements were performed in order to determine $[C]$ and to compare it with $[c]$. Table 1 shows the results of $[C]$, $[c]$ and the electrical activity efficiency of the C atoms for samples grown at three different temperatures. As Tg is reduced, not only the electrical activity increases but also the C atomic incorporation probability does. The electrical activity efficiency is defined as the ratio of $[c]$ and $[C]$. The improved sticking coefficient for C at low Tg is attributed to the fact that AsH_3 is only partially cracked at low temperatures favoring the incorporation of other species in the V sub-lattice. In fact, at low temperatures also oxygen is more effectively incorporated as a donor occupying sites in the V sub-lattice. [8]

Table 1 Results of [C], [c] and the electrical activity efficiency of the C atoms for samples grown at three different temperatures.

Sample	T_g (°C)	[C] (cm^{-3})	[c](cm^{-3})	Electrical activity efficiency (%)
#365	530	5×10^{18}	1.3×10^{18}	26
#363	575	1.3×10^{18}	2×10^{17}	15
#353	650	5×10^{16}	-1.44×10^{16}	0

Fig. 3 shows an increase in the conductivity as T_g is reduced due to the increase in [p]. One concludes that despite the more important incorporation of defects at low T_g, in particular of oxygen [8], the conductivity is improved. Thus, C doped InAlAs layers grown at temperatures of the order of 550°C and subjected to a post-growth anneal are adequate, in terms of their transport properties, for application in electronic devices such as high speed transistors, as has already been discussed by other authors [4, 9]. However, for optical and optoelectronic devices it is crucial that good optical quality is also achieved.

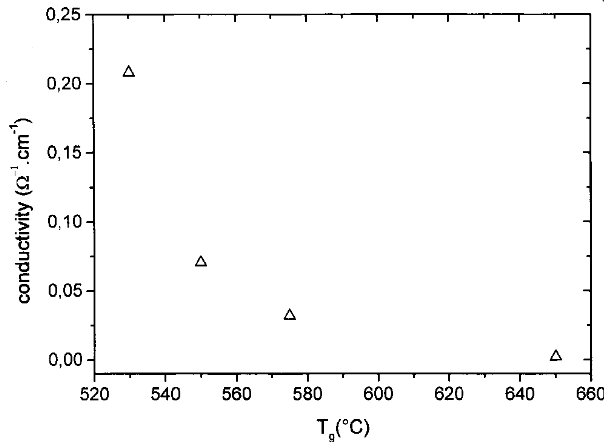

Figure 3. Conductivity as a function of growth temperature.

We observe that as Tg decreases the optical quality of the layers gradually deteriorates. The near-edge optical transition decreases its intensity as T_g is reduced from 650°C and it is no longer observed when the growth temperature reaches 575°C. This effect is independent of the presence of C impurities, since the same result is observed for the undoped reference samples. As shown in Fig. 1, only when Tg is lowered to 575 ^{o}C the as-grown samples are p-type. In order to better understand this effect, one can think that lowering Tg, is the equivalent of decreasing the V/III ratio [10]. This is because at lower temperatures the cracking of the AsH_3 is less efficient, as mentioned before. Hence, at low Tg, more defects are incorporated, as well as more C atoms and less passivating H atoms. As observed, the more efficient incorporation of C or its improved electrical activity are not related to the reduced PL intensity. Therefore, one concludes that some of the incorporated defects are very efficient non-radiative centers. The nature of such center is still under investigation.

The effect of the doping in the near-edge emission can be observed in Fig. 4, where the 30 K PL spectra for an undoped and a doped sample, both grown at 650°C, are shown. In the case of the doped sample, the PL peak energy is 15 meV below that of the undoped reference sample. PL measurements as a function of excitation power (P) show that the peak energy blue-shifts as P increases for the doped sample, as expected from a donor-acceptor (D-A) type transition. Since this does not occur for the reference sample, one can conclude that the D-A transition is limited by the presence of C acceptors. The unintentionally incorporated donors compensate the acceptors. One should remember that these samples are n-type. The donors are most likely oxygen which are usually incorporated in Al containing material.

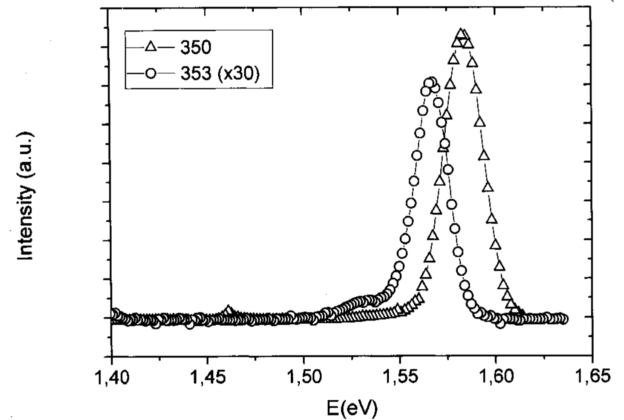

Figure 4. 30 K photoluminescence spectra of a doped (circles) and an undoped (triangles) sample.

IV Conclusion

C doped InAlAs layers have been studied as a function of Tg. It was shown that the [p] and the conductivity increase as Tg decreases due to a more efficient C incorporation as well as a lower incorporation of passivating H atoms. Annealing the samples increases the electrical activity due to the removal of H atoms. It has been also observed that as Tg decreases, the intensity of the near-edge PL emission is reduced and for Tg ¡ 575°C, it is no longer observed. This effect is attributed to the incorporation of non-radiative defects at low Tg. It is suggested that TBAs, as As source, could be an alternative for AsH_3 to obtain good optical quality C doped AlInAs, since its cracking temperature is lower that that of AsH_3[11], and consequently the low Tg would not lead to a significant incorporation of defects.

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