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Deep Levels Due to Induced Dislocations in Vapor Phase Epitaxial GaAs

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In order to see electrical properties of dislocations in GaAs, high purity vapor phase epitaxial GaAs was deformed and induced deep energy levels were investigated by the DLTS (Deep Level Transient Spectroscoty). It was found that both the electron trap EL2 and the hole trap HL2 were induced very much and that Cu and Fe levels are also increased greatly by the deformation. However, no new energy level due to dislocation itself was detected. Types of the induced point defects seemed to depend on the direction of bending.

I Introduction

Undoped LEC (Liquid Encapsulated Czchralski) grown GaAs is getting great interests of GaAs IC researchers, because it becomes semi-insulating without Cr doping and because it can be grown much faster than by HB (Horizontal Bridgeman) method. Unfortunately, many dislocations such as more than 10^4 cm^{-2} are induced in LEC GaAs during the grouth and its density sometimes varies more than one order of magnitude from the center of the wafer to the periphery.

Active layers of the GaAs IC's must be prepared by direct ion-implantation into semi-insulating substrates. It has been found that threshould voltages of GaAs FET's fabricared by the ion implantation scatters very much and seems to depend on the grown-in dislocation densities[1-6]. These results suggest that the dislocations or the points defects related with dislocations electrically affect the activation of the implanted donors.

Purpose of this work is to make clear electrical properties of the dislocations in GaAs and what kinds of point defects and defect levels are induced by the plastic deformation of GaAs.

It was found that the electron trap EL2 and hole trap HL2 are introduced, and Cu and Fe levels increased very much by the deformation but no new deep level due to dislocation itself was found. Types of induced point defects seemed to depend on direction of the deformation.

II Experimental

Bulk GaAs generally contains as many point defects as more than 10^{16} cm^{-3}[7]. It is possible that induced dislocations interact with these grown-in point defects. Therefore, GaAs to be deformed was prepared by a vapor phase epitaxy[8].

Epitaxial layers were grown on (100) substrates. Carrier densities and thicknesses of the grown layers were generally 3-5x10^{15} cm^{-3} and about 10 μm, respectively. The deformation was performed by high purity carbon jig as shown in Fig. 1 in a hydrogen atmosphere. The carbon jig is a kind of three points bending jig[9] and was pushed at elevated temperatures with an constant power. Most of the deformations were carried out at 650°C for 5 min.

One wafer was devided into 5 parts as shown in Fig. 2(a) and the epitaxial layer was either expanded or compressed in either <011> or <0$\bar{1}$1>

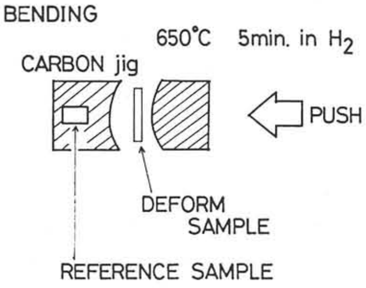

Fig.1, Schematic diagram of the bending jig.

direction as shown in Fig. 2(b). One chip of the wafer was used as a reference sample to see the effect of the same heat treatment on change of the deep levels and carrier densities. It was put on the carbon block of the bending jig in a face down way as shown in Fig. 1.

Deep levels were measured by an ordinary DLTS (Deep Level Transient Spectroscopy) method. Electron traps were measured by Al Schottky barriers. For the measurements of hole traps, Zn was diffused at 600°C, for 10 min. in a quartz ampoule. In calculation of the electron trap densities, effect of the deionized front region was taken into account[7]. Induced dislocaton desities were estimated from the EPD (Etch Pit Density) revealed by molten KOH. It ranged from $5x10^5$ cm^{-2} to $5x10^6$ cm^{-2}.

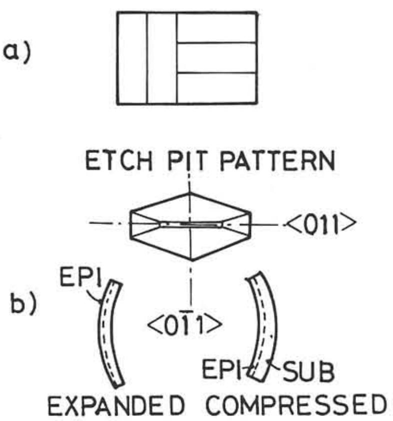

Fig.2, Schematic explanation of cutting way of the wafer (a), and bending direction of the wafer(b).

III Experimental Results and Discussions

3.1 Decrease of the free carrier density by the deformation

Figure 3 shows carrier density proifles of the deformed and the reference pn junction samples, which were obtained by C-V measurements. The dislocation density induced by the deformation was about 10^6 cm^{-2}. The carrier densities of the deformed sample are greatly reduced from the reference sample's value and depend on the measurement temperature, whereas the carrier density of the reference sample does not change very much by the mesurement temperature. These results indicate that many deep accepter type defects or impurities are induced by the deformation.

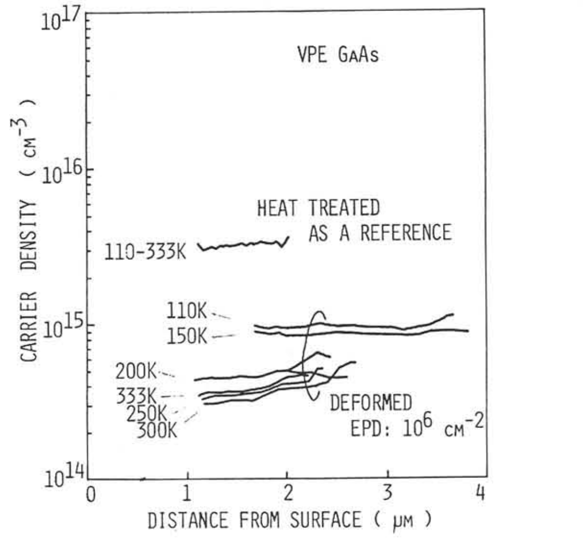

Fig.3, Change of the carrier density profile with temperature for the deformed and reference pn junction samples.

3.2 Induced deep levels by the deformation

Figure 4 shows the DLTS signals of pn junctions of the deformed samples and the reference sample. The epitaxial layer was expanded in <011> direction as shown in Fig.2. For the reference sample, only Cu and Fe levels are observed but no EL2 level was detected, probably due to the fact that EL2 density is decreased by the heat treatment[10]. By the deformation, the electron trap EL2 and the hole trap HL2, which is usually observed in LPE GaAs, appeared. In addition to that, the Cu and Fe levels are increased greatly.

Absolute deep level densities in the reference and the deformed samples are listed in Table 1. The electron trap EL2 was increased very much by the deformation. Its density was as high as $9.4x 10^{14}$ cm^{-3}. Such high desity of EL2 was never observed in VPE GaAs. Acturely no EL2 density was observed in the reference sample. Therefore, this large number of EL2 density must be introduced by the deformation.

As shown in Fig. 4, the hole trap HL2 is introduced by the deformation. Its concentration is about $6.9x10^{13}$ cm^{-3}. This level is always observed in LPE GaAs, and is thought to be due to a point defect related with As vacancy. The Cu density in the deformed sample is $2.4x10^{14}$ cm^{-3}, which is about 7 times larger than that in the reference sample. The Fe density is about 2 times larger than that in the reference sample. Cu generally diffuses in GaAs as an interstitial Cu and occupies Ga site to make deep accepter levels. So the substitutional Cu solubility greatly depend on Ga vacancy density[11]. Dislocation climb motion due to the deformation can emit many Ga vacancies and As vacancies. Therefore, it is reasonable to think that Cu and Fe solubility is increased very much by the deformation due to creation of Ga vacancy.

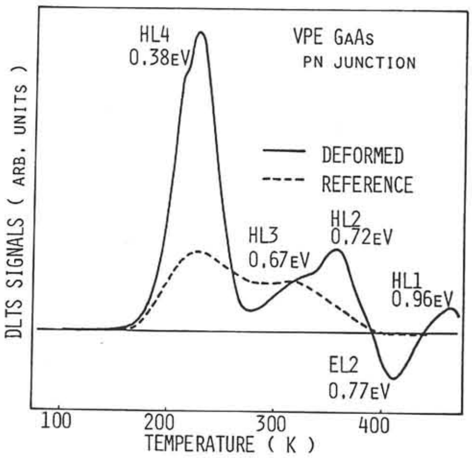

Fig.4, DLTS signals for the deformed and reference samples. The bending was in <011> direction.

Table 1, Trap densities of the deformed and reference samples.

	Levels	Et(eV)		Reference	Deformed
	EPD (cm^{-2})			$2.0X10^4$	$1.4X10^6$
PN JUNCTION			Nd-Na	33.0	9.4
	Levels	Et(eV)		($X10^{14}cm^{-3}$)	
	HL1	0.96	Cr	<0.01	0.21
	HL2	0.72		<0.01	0.69
	HL3	0.67	Fe	0.25	0.63
	HL4	0.38	Cu	0.35	2.4
	EL2	0.77		0.01	0.62
SCHOTTKY			Nd-Na	69.0	9.8
	Levels	Et(eV)			
	EL2	0.77		<0.01	9.4

*Reference : heat treated on the same condition

3.3 Dependence of the trap density on the dislocation density

It is difficult to bend GaAs wafer uniformly, so the curvature of the deformed GaAs sample changes depending on the position of a wafer. The measured EPDs of the sample varied from about $5x10^5$ cm^{-2} at the edge of the wafer to about $5x10^6$ cm^{-2} at the center.

To see dependence of trap densities on the EPD the trap densities are plotted against the EPD in Fig.5, together with the net donor density (Nd-Na). The net donor density decreases with increase of the EPD, but tend to saturate at higher EPD values. The HL4 (Cu) density increases monotonically with increase of the EPD, whereas the HL3 (Fe), HL2 (LPE-B), and HL1 (Cr) densities have the maximum at around EPD of 10^6 cm^{-2} and decreases again for higher EPDs. Dependence of the electron trap EL2 density on the EPD was also obtained on a Schottky barrier sample. It also has a maximum and decreases again at a higher EPD. These results suggest that type of the induced defects is different at a higher EPDs. In the other words, when the deformation is too much or the deformation speed is too fast, the induced dislocation networks become so complicated and different types of point defects might be introduced.

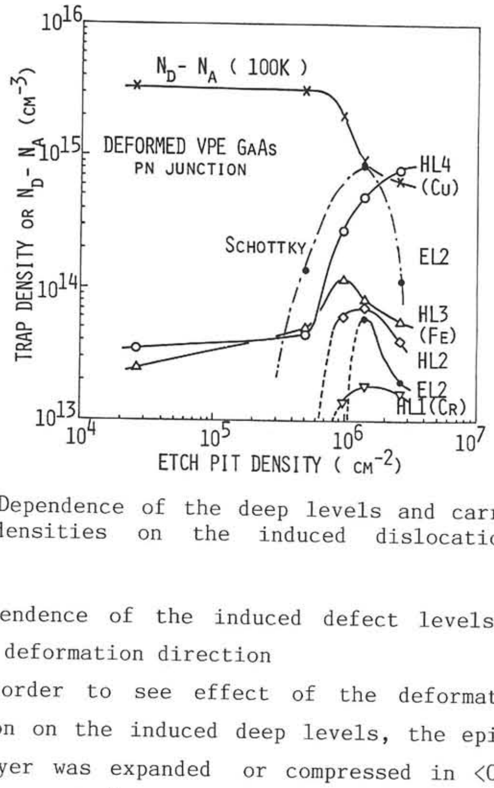

Fig.5, Dependence of the deep levels and carrier densities on the induced dislocations.

3.4 Dependence of the induced defect levels on the deformation direction

In order to see effect of the deformation direction on the induced deep levels, the epitaxial layer was expanded or compressed in <0$\bar{1}$1> direction at 650°C as shown in Fig. 2. The EPD was about $5x10^6$ cm^{-2}. The DLTS signals measured on the pn junction samples are shown in Fig. 6. A remarkable difference of the induced deep levels was observed for the expanded and compressed samples. Only the hole trap HL2 ($1x10^{15}$ cm^{-3}) was observed in the expanded sample. This hole trap HL2 is observed in all LPE GaAs and is never observed in VPE GaAs. Since LPE GaAs is grown in Ga rich condition, the hole trap HL2 is thought to be due to As vacancy related point defects, such as As single vacancy or their complexes with some interstitials.

On the other hand, when the epitaxial layer was compressed, the electron trap EL2 was greatly induced ($\sim 1x10^{15}$ cm^{-3}) but the hole trap HL2 was not induced very much ($1.5x10^{14}$ cm^{-3}). Furthermore a lot of HL4 (Cu: $9x10^{14}$ cm^{-3}) and HL3 (Fe: $5.7x10^{14}$ cm^{-3}) are observed. These results might seem to be very complicated, but if one takes into consideration the facts that the electron trap EL2 is a Ga vacancy related defect and that both Cu and Fe occupy Ga site to make deep acceptors, the results becomes to be very clear. When the epitaxial layer is compressed, climb motion of dislocations probably emits much more Ga vacancy than As vacancy and if there are many interstitial Cu and Fe in GaAs or on GaAs surface, it is reasonable to think that many Cu and Fe are incorporated in Ga sites in addition to formation of Ga vacancies.

In the case of bending in <011> direction of Fig.2, difference of induced deep levels between the expanded and compressed deformations was not so clear. Climb motion of dislocations in this case probably emits both of Ga vacancies and As vacancies, and both of the electron trap EL2 (therefore, the Cu and Fe levels, too) and the hole trap HL2 are induced in almost the same amount.

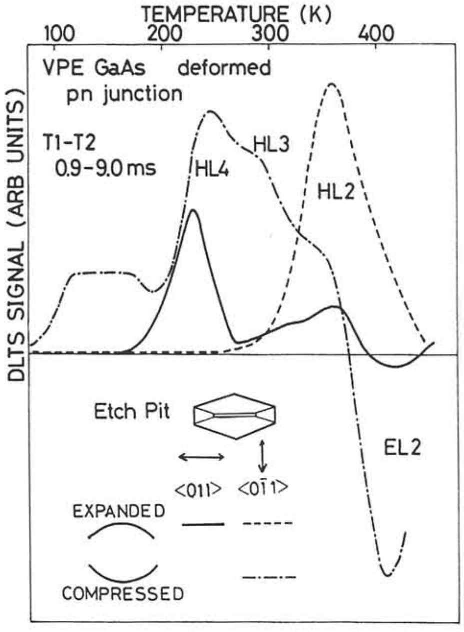

Fig.6, DLTS signals of the expanded and compressed samples in <011> direction.

IV Conclusions

High purity vapor phase epitaxial GaAs was deformed to investigate deep energy levels of dislocations and point defects induced by the deformation. It was found that both the Ga vacancy related electron trap EL2 and the As vacancy related hole trap HL2 were induced very much, but no new level due to dislocation itself was induced. Introduction of these point defects seemed to depend on the direction of bending, probably due to difference of emission of point defect by the climb motion of dislocations. The Cu and Fe levels were also increased very much. This can be understood by the increase of the solubility of these impurities due to increased Ga vacancies.

Acknowledgements

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References

1) Y.Nanishi, H.Yamazaki, T.Mizutani and S.Miyazawa: Proc. Int. Symp. GaAs and Related Compounds Japan,1981,p7.
2) Y.Nanishi, S.Ishida, T.Honda, H.Yamazaki and S.Miyazawa: Japan. J. Appl.Phys. 21 (1982) L335
3) S.Miyazawa and Y.Nanishi: Proc. 14th Conf. Solid State Devices, Tokyo,1982,p525.
4) Y.Nanishi, S.Ishida and S.Miyazawa: Japan. J. Appl. Phys. 22 (983) L54.
5) T.Honda, Y.Ishida, S.Miyazawa, H.Yamazaki and Y.Nanishi: Japan. J. Appl. Phys. 22 (1983) L270
6) S.Miyazawa, Y.Ishii, S.Ishida and Y.Nanishi; Appl. Phys. Lett. 43 (1983) 853
7) F.Hasegawa, Y.Iwata, N.Yamamoto and Y.Nannichi: Japan. J. Appl. Phys. 22 (1983) L502.
8) A.Koukitsu, H.Seki and M.Fujimoto: Japan. J. Appl. Phys. 15 (1976) 1591
9) S.K.Choi, M.Mihara and T.Ninomiya: Japan. J. Appl. Phys. 16 (1977) 737.
10) F.Hasegawa, N.Yamamoto and Y.Nannichi: Appl. Phys. Lett. 44 (1984) Aug.15th, to be published
11) F.Hasegawa: J. Appl. Phys. 45 (1974) 1944.