ABOUT THE NATURE OF DEFORMATION ANISOTROPY OF THE InP:Zn CRYSTALS

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ABSTRACT

The regularities of plastic deformation under the action of concentrated load in two crystallographic directions <100> and <110> of InP crystals doped with zinc, have been investigated. The study revealed that the crystal deformation in <110> direction is more sensible to the distribution of growth dislocations than in <100> one. The hypothesis about the nature of the observed anisotropy of plastic properties by indentation and by indentor motion is advanced.

1. INTRODUCTION

The InP crystal is a material widely used in semiconductor electronics. It is known that doping of a given compound with different impurities is the way to obtain materials with programmed electrophysical properties. The structure perfection, in particular, the density and distribution of growth dislocations play an important role. The question about the correlation between growth dislocation density and both mechanical and electrophysical properties has been little studied [1-3]. There is also no information about the mechanism of plastic deformation of InP crystals [3-5]. Therefore, the purpose of this paper is to investigate the regularities of the InP crystal deformation in connection to the growth dislocation distribution.

2. EXPERIMENTAL PROCEDURE

The InP crystals were grown using the liquid-encapsulated Czochralski (LEC) technique. The investigations were performed on the (001) face. The Vickers diamond pyramid was used for the crystal deformation (the indentation and scratching methods). Microhardness by indentation (H) and by scratching (H_s) was calculated using the usual formulas [6]. The microhardness measurements were produced in two orientations: 1) the indentor's diagonals d II <100> and 2) d II <110>.

3. RESULTS AND DISCUSSION

The data shown below in the table reveals some interconnection among the parameters.

Table Relationship different physical parameters N, H, n, μ in pure and doped with Zn InP crystals

Crystal	N, cm^{-2}	H, kg/mm^2	n, cm^{-3}	μ, cm^2/BC
n-InP	$(2 \div 3) \times 10^4$	360	$4{,}1 \times 10^{16}$	3900
p-InP:Zn	5×10^4	375	$2{,}7 \times 10^{18}$	50
p-InP:Zn	$2{,}1 \times 10^5$	410		

There is some clear correlation between the change of the growth dislocation density (N) and other properties of InP crystals (microhardness (H), concentration of bearers (n), electron mobility (μ). Please note that average values for N and H are mentioned in the table. However, it is known that the density of growth dislocations is spreads nonhomogeneously in the table. But it is known that the density

of growth dislocations is spreads nonhomogeneously in the diameter (D) of a crystal ingot. Therefore it is important to follow the change of these parameters inside of the ingot. To this purpose some sections on the (001) plane were cut out from the big ingot (fig.1a). The density of dislocations and respectively the microhardness H and H_s for <100> and <110> directions were evaluated on them. Received results are given in the fig.1,2. As it is seen from fig.1b, the curve of the dislocation density (N) has a concave form: the minimum N is in the center of piece, the maximum - on the edge of the surface.

Also a change of the Vickers and scratching microhardness (fig.2).

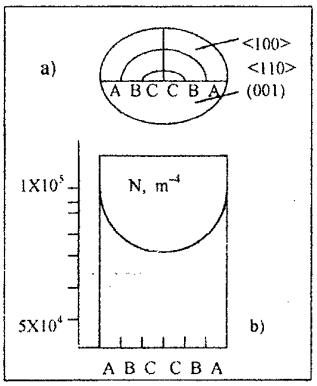

Fig.1 a) Schematic image of investigated regions of surface with D-20 mm; b) The distribution of the dislocation density on the (001) of the InP:Zn piece

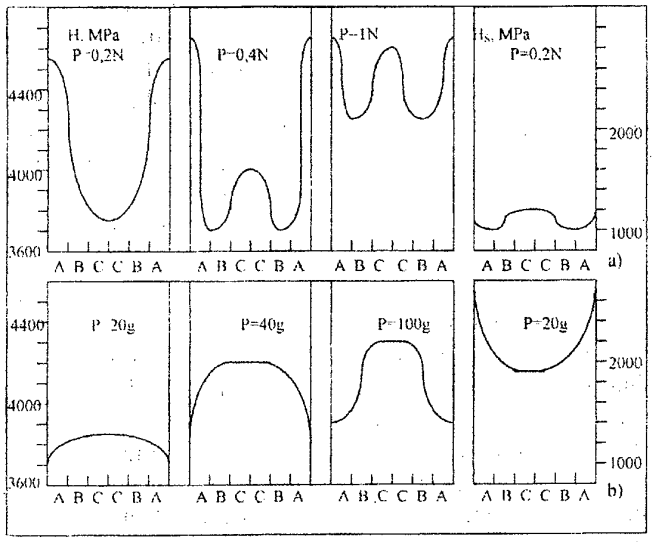

Fig.2 Microhardness change on the (001) plane of InP:Zn crystals by indentation (H) and by indentor motion (H_s) for two orientations <100> (a) and <110> (b) at measured removal from the cut center

It is important that curves have a complicated shape for both directions. As observed, the curves H(D) (dII<100>) have in general a concave form and the best correlation with N(D) takes place at P = 20 g; the curve H_s(D) has slightly a convex shape.

A contrary situation occurs in the <110> direction: H_s(D) has a clearly concave form, analogous to the dependence N(D). The curves H(D), on the contrary, have a convex shape and the convexity increases with the growth of the indentor load.

The interconnection between the mechanical properties change and the dislocation distribution is also clearly followed by a change in the anisotropy coefficient K ($K = (H_{max}-H_{min}) / H_{min}$, %), fig.3.

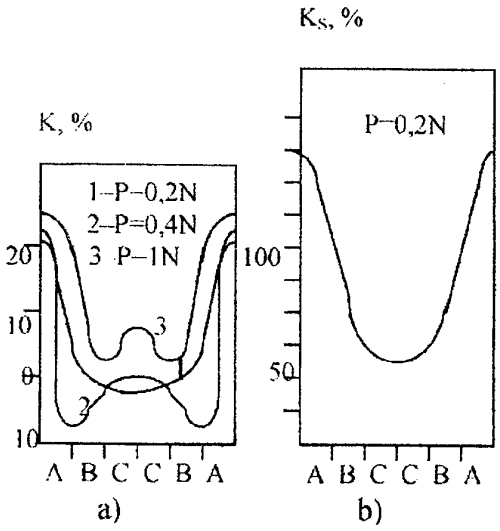

Fig.3 InP:Zn, plane (001). The change of anisotropy coefficient at indentation (a) and at indentor motion (b). Curves: $K = (H_{<100>} - H_{<110>} / H_{<110>})$ %, $K_s = (H_{s<110>} - H_{s<100>} / H_{s<100>})$ %

One can see that the curves K(D) and K_s(D), have generally a concave shape i.e. the anisotropy degree increases with the growth of the dislocation density (fig.3 and fig.1b).

It is known that one of the main causes of the dislocation production in crystals obtained from melt are the thermoelastic strains which appear due to non-uniform cooling /6/. The regions of most dislocation production (in our case characterised by parameter N) indicate high internal stresses present in the crystal. A regular radial increase of the growth dislocation density, the change of the microhardness and of the coefficient of microhardness anisotropy are obviously due to a like cause, and confirm the increase of the internal stress concentration in the periphery parts of the cut versus in its central one.

The above data also specify that microhardness in the <100> (fig.2a) and <110> (fig.2b) directions changes inadequately by the radial removal from the cut centre. The curves $H_{s<110>}$(D) and $H_{<100>}$(D) in general are synchronously changed with N(D), i.e. by deformation along these directions the material undergoes the hardening in the regions with high concentration of internal stresses. The curves $H_{s<100>}$((D) and $H_{<110>}$(D) are in contrast to this.

The inadequate behaviour of curves Hs<100> and Hs<110> at high temperature deformation (T>300K) was explained in the work /5/ by different contributions of two mechanisms of plastic deformation - sliding and twinning in the process of deformation by the indentor motion along <110> and <100> directions. It was shown that the sliding process brings more contribution in the first case, but the twinning - in second one.

If these factors play a certain role in the deformation of InP crystals at room temperature, then the <110> scratching direction must be more sensible to the internal stress change versus <100> direction, because the dislocations arising at indentation will meet by sliding the bigger resistance in regions with more high stress concentration. The twinning process of the contrary will be lightened in the overstressed region /7/. This is really noticed in the experiment (fig.2, H_s<110> and H_s<100>).

A contrary situation takes place in the case of the indentation - with the N(D) change better correlates <100> direction. It is known that the deformation maximum takes place in the center of Vickers indentation sides [6, 7]: namely for d II <100> maximum of plastic deformation noticed along <110> direction, but for d II <110> - along <100> one. Comparing these results, one may conclude, that the twinning process in responsible in case of indentation at d II <110> and slidind one - in case of

indentation at d II <100>. It is necessary to note that the above two factors, are not the only ones in the deformation of the InP crystals. However these factors will act in the region of high temperatures. However, at low temperatures in covalent crystals the dislocation mobility is small and the deformation at $T < 0,4\ T_{melt}$ (for InP T_{melt} = 1346 K) cannot be stipulated for the only dislocation mechanism. The mechanism of interstitial plasticity, i.e. mass transfer by the motion of point defects can play some role at relatively low temperatures [6, 8, 9]. The curves in fig.2 confirm this idea. Indeed the pointed out correlation between H(D) and N(D) exists only in general outline. In reality the situation is more complicated. There are maximum on the curves which are more pronounced by the increase of the indentor load. Presence of maximum in fig.2 curves indicates that the deformation of the InP single crystals at the room temperature is determined not only by the mentioned mechanisms - sliding and twinning. Probably a third mechanism - interstitial plasticity also exists. The contribution degree of this mechanism is the aim of a separate investigations.

4. CONCLUSIONS

1. The regularities of plastic deformation by indentation for two directions <100> and <110> ((RT) in dependence of growth dislocation density were studied.

2. It was noticed that the deformation along <110> directions is more sensible to the N change than along <100>.

3. The observed microhardness anisotropy was explained as a result of two deformation mechanisms - sliding and twinning.

4. The third mechanism probably, interstitial plasticity, takes place by indentation of the InP single crystals at room temperature as well as marked above two ones. Its role r will be investigated in the future.

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