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► **To cite this version:**

P. Petroff. POINT DEFECTS AND DISLOCATION CLIMB IN III-V COMPOUNDS SEMICONDUCTORS. Journal de Physique Colloques, 1979, 40 (C6), pp.C6-201-C6-205. 10.1051/jphyscol:1979640 . jpa-00219056

HAL Id: jpa-00219056

<https://hal.science/jpa-00219056>

Submitted on 4 Feb 2008

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POINT DEFECTS AND DISLOCATION CLIMB IN III-V COMPOUNDS SEMICONDUCTORS

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Résumé.- Le phénomène de montée de dislocations dans les semiconducteurs du type III-V est ici examiné. Les domaines où l'information expérimentale est nécessaire à une meilleure compréhension sont discutés. Un modèle de montée de dislocations nécessitant la sursaturation de défauts ponctuels d'un seul élément est trouvé cohérent avec les principales observations expérimentales. La montée rapide des dislocations est attribuée à la diffusion accélérée des défauts ponctuels sous l'effet de la recombinaison. Finalement l'interaction entre le réseau de dislocations de montée et un niveau donneur profond suggère une association possible de ce centre donneur avec la source de défauts ponctuels nécessaire à la montée dans les structures $Ga_{1-x}Al_xAs$.

Abstract.- This paper reviews the low temperature dislocation climb process in III-V compounds semiconductors and points out areas in which more experimental information is needed to understand this complex problem. A dislocation climb model requiring the supersaturation of point defects of only one element of the compound is found to account for the main climb features. Rapid dislocation climb is attributed to recombination enhanced defect motion. Finally evidence of an interaction between the climb dislocations and a deep level donor center suggest that it might possibly be associated with the source of point defects needed for dislocation climb in $Ga_{1-x}Al_xAs$ structures.

Introduction.- Dislocation climb in semiconductors and more specifically for III-V compound semiconductors such as $Ga_{1-x}Al_xAs$ and GaP where low temperature dislocation climb has been found to play an important role on the devices properties /1-4/ has been a complex process to unravel. Probably this is because of the complex characteristics of the point defects (charge, impurity effects, clustering) in a compound where two elements have to be dealt with during climb. Furthermore, other effects such as dislocation dissociation, glide or shuffle core structures, and charge effects have to be taken into account for a complete picture of dislocation climb. This paper will discuss three features which are of importance in the climb process in III-V compounds : a) the crystallography and structural aspects of dislocation climb, b) the driving forces and c) the point defects problem. For space reasons, only the low temperature climb ($T \ll$ half the compound melting temperature) is considered in this paper.

1. Crystallography and structure of dislocation climb.- The crystallographic observations of dislocation climb in $Ga_{1-x}Al_xAs$ laser devices /1,2,3/ and GaP /4/ light emitting diodes have all been carried out by conventional transmission electron microscopy (TEM). The results of these TEM analysis are summarized in table I and an example of a climb network is given in figure 1. The important features of these dislocation networks are : a) the climb direc-

tions, b) the presence of small dislocation loops inside the main dipole and c) the dislocation network Burger's vector which is that of the dislocation from which they originate.

Table I :

Material	$Ga_{1-x}Al_xAs$	GaP
Burger's Vector	$\frac{a}{2} \langle 110 \rangle$	$\frac{a}{2} \langle 110 \rangle$
Climb Plane	{110}, {110}	$\langle 110 \rangle$
Climb direction	$\langle 100 \rangle \langle 110 \rangle$	$\langle 110 \rangle$
Dislocation Character	Edge, Mixed	Mixed
Type of Climb Network Produced	Edge Dipole Helical Dipole Dislocation loops	Helical Dipole
Climb Character	Interstitial Dipole Vacancy loops	Interstitial Dipole Vacancy loops

The nature of the giant helical dipoles has finally been established as being interstitial /3,4/ while the small dislocation loops are of a vacancy type

/4,5/. Unfortunately, there is presently no weak beam TEM contrast analysis available for these defects to determine possible dissociation such as those reported for glide dislocations in GaAs /6/ or GaP /7/. The fine features corresponding to small dislocation loops attached to the main dipole (See figure 1) are similar to those found during climb of dissociated dislocations in f.c.c. metals /8,9/ suggesting that the dislocations forming the dipole may indeed be dissociated into partials. As shown by studies of dissociated dislocation climb in metals /8/, the details of the climb model in this case will be considerably complicated by the dissociation. For this reason we consider a climb model only for undissociated dislocations and assume that in the case of dissociation, dislocation climb is taking place at the constricted parts of the dislocation.

behind the climbing dipole. This observation has been the motivation for a new dislocation climb model /5/ in compound semiconductors with the zinc blende structure. The compound nature of the lattice requires that two atomic species, instead of one, be absorbed or emitted to produce jog motion. Since it is rather unlikely that point defects of the two species are present in equal supersaturation in these compounds, a supersaturation of one type of point defects is postulated. For representing the model we choose the Ga interstitial, $(Ga)_i$, as the excess point defect. The $(Ga)_i$, shown in figure 4A has attached to the dislocation core in figure 4B and an As vacancy at the dislocation core $(V)_{As}$, is then created in the process. The formation of an As vacancy, $(V)_{As}$, is then involved to provide the As atom which completes the climb as shown in figure 4C. The $(V)_{As}$ has moved so

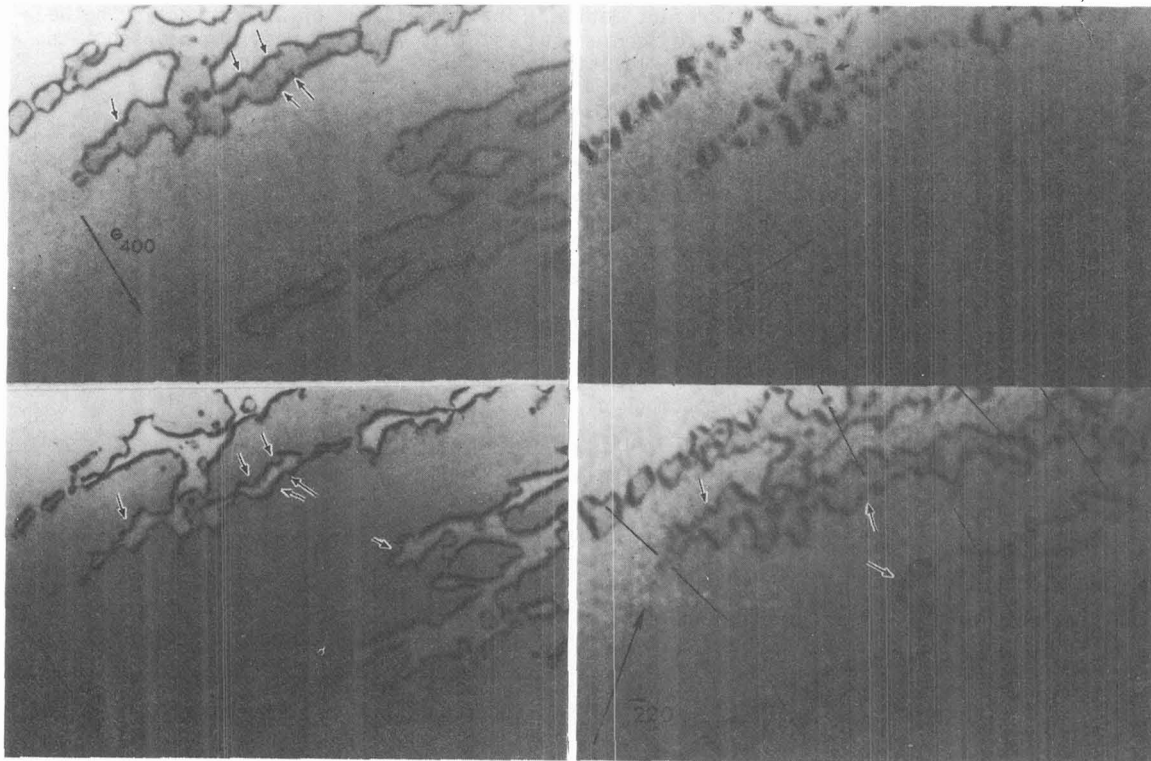


Fig. 1 : Transmission electron micrographs of a dislocation network produced by climb during carrier injection and degradation of a $Ga_{1-x}Al_xAs$ (DH) laser device. Arrows point out the small loops attached to the main dislocations. The Bragg vectors are indicated for each micrograph.

The small dislocation loops observed inside the climb network are characteristic of dislocation climb produced by optical or current or electron beam carrier injection /5,10/. The examples in figures 2,3 show the dislocation loops confined to the inside of the main climb structure, suggesting that they represent a by-product of the point defects left

as to compensate the local stresses at the dislocation (Fig. 4D) and the process may be repeated. Thus a supersaturation of only one type of point defects may suffice for dislocation climb, providing that a trail of point defects on the other sublattice is left behind the moving dislocation. This simple model can easily be adapted to climb by jog motion with

identical results. A thermodynamical justification for this model raises the question of driving force for the dislocation climb.

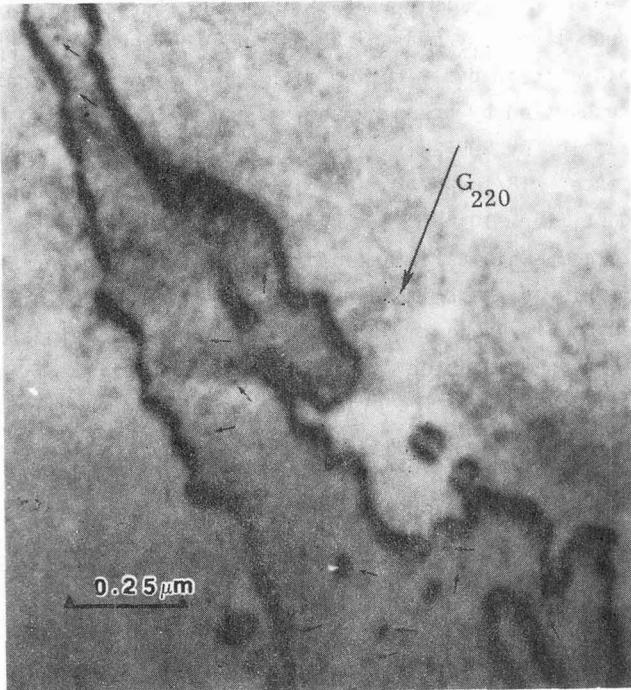


Fig. 2 : Transmission electron micrograph of dislocations produced by climb during current carrier injection in a $Ga_{1-x}Al_xAs$ laser device. Secondary dislocation loops inside the main dipole are shown by arrows.

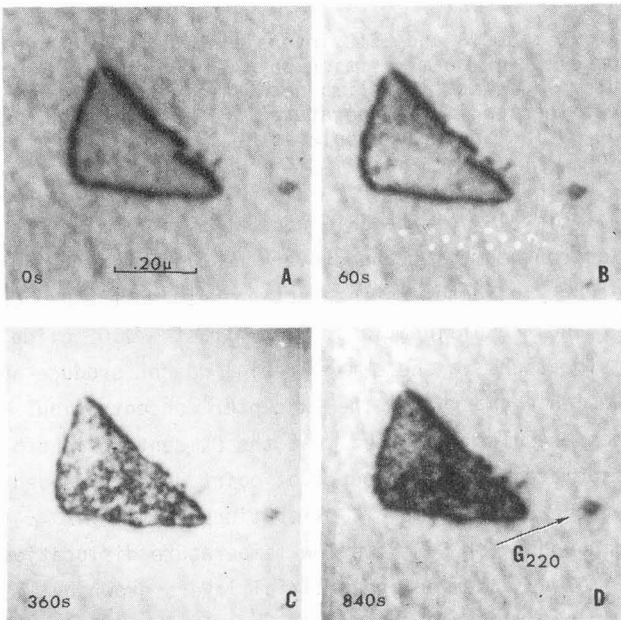


Fig. 3 : A. Transmission electron micrograph of a dislocation loop produced during carrier injection and degradation of a $Ga_{1-x}Al_xAs$ laser device. B,C,D: secondary defect structures produced by electron beam stimulated carrier injection. The times in the sequence of micrographs are indicated.

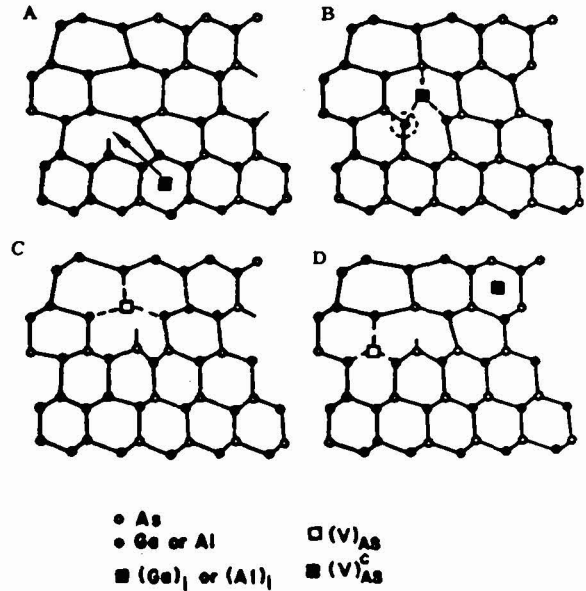


Fig. 4 : Dislocation climb model for an undissociated dislocation in the zinc blende lattice projected normal to the $(1\bar{1}0)$ plane. The 60° dislocation with $b = 1/2 \langle 110 \rangle$ is shown in 4A. 4B-4D schematic of the dislocation climb.

2. Driving force during low temperature dislocation climb.- To follow the example of dislocation climb in GaAs structures, we note that since the crystals were grown under Ga-rich conditions, it is reasonable to assume that $(Ga)_i$, and /or $(V)_{As}$ are the dominant defects at equilibrium near the growth temperatures and that some of them have been quenched to low temperature.

In the case of low temperature climb, the driving force is the lower chemical potential of the vacancy relative to the interstitial under the supersaturation condition. The driving force for climb is

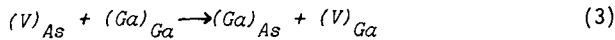
$$F \approx \frac{kTb}{V} \ln \left(\frac{[(Ga)_i]_0}{[(Ga)_i]} \frac{[(V)_{As}]}{[(V)_{As}]_0} \right) \quad (1)$$

in which b is the dislocation Burger's vector and V is the As or Ga atomic volume. The $[]$ and $[]_0$ indicate respectively the quenched-in and equilibrium concentrations at low temperatures for the respective point defects. In the derivation of equation (1), it was assumed that the reaction



should produce a lowering of the crystal's free energy. In fact this will be the case if we note that the formation energy of $(Ga)_i$ is much larger than that of $(V)_{As}$ because of the larger volume of the former $/11/$. It follows that the supersaturation ratio of

interstitials $[(Ga)_i] / [(Ga)_i]_0$ is larger than that for vacancies $[(V)_{As}] / [(V)_{As}]_0$ and thus the reaction (2) is favored. The excess vacancies will eventually form small clusters. Since the formation energy of antisite defects $(Ga)_{As}$, is not expected to be large in the vacancy rich regions, a possible reaction might be



in which the Ga on an As site defect, $(Ga)_{As}$, is produced. The reactions (2) and (3) provide the two types of defects $(V)_{Ga}$ and $(V)_{As}$ needed for the formation of intrinsic dislocation loops of the type shown in figures 2 and 3.

If dislocation climb occurs by jog motion, its velocity is controlled by the diffusion of $(Ga)_i$ to the jogs. The jog diffusivity, assuming no pipe diffusion, is then

$$D_j \propto b^2 \exp \left(- \frac{E^*}{kT} \right) \quad (4)$$

where E^* , is the effective activation energy for point defect diffusion under carrier injection conditions. Point defects in compound semiconductors usually behave as recombination centers and are associated with deep level traps. Carrier recombination at such centers has been shown to rapidly enhance point defect diffusion at low temperature /13, 14/. Under recombination enhanced diffusion conditions $E^* \cong E_T - E_g$ where E_T is the thermal diffusion energy and E_g is the bandgap energy. From the series of micrographs in figure 3, it is clear that carrier injection stimulates recombination enhanced diffusion of point defects and promotes point defect clustering. In general, the rapid dislocation climb in $Ga_{1-x}Al_xAs$ and GaP may be attributed to this effect.

3. Point defects in III-V compounds semiconductors.

The origin of the point defects taking part in the climb process has remained one of the major unknowns. Experimental observations /10/ would indicate that it is "grown-in" the material. It has also been proposed /15/ that the necessary defects originate at the dislocation line itself, by recombination enhanced emission of vacancies from the dislocation core. A major difficulty of this model is that it does not explain several observations of interrupted dislocation climb in devices under injection conditions /10/. It also does not account for the absence of dislocation climb for dislocations such as those shown in figures 1, 2, and 3 when carrier injection is stimulated by the electron beam of the microscope. In an

attempt to resolve this problem in $Ga_{1-x}Al_xAs$, dislocation climb networks have been analyzed by scanning deep level transient spectroscopy (SDLTS) /16,17/. This analytical technique allows for the characterization of deep levels by measuring their energy levels and displays their distribution with a high degree of spatial resolution by scanning the electron beam over the sample. By SDLTS it was found that a donor deep level center /18/, the DX center, which is the most abundant (10^{17} - $10^{18} cm^{-3}$) center in $Ga_{1-x}Al_xAs$ (DH) structures, exhibits a marked decrease in its concentration over the climb dislocation network /19/. This change is shown in figure 5 where the SDLTS signal from the DX center is displayed as a function of the probe position. The electron beam induced current (EBIC) is also shown in figure 5 for the same scanning line.

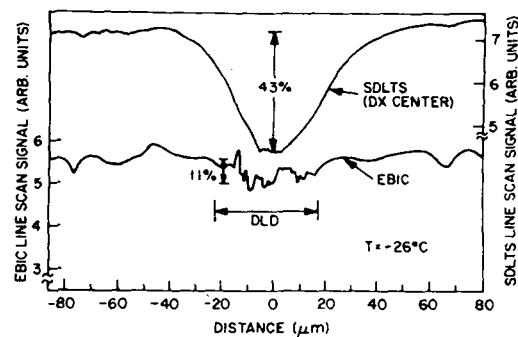


Fig. 5 : SDLTS and EBIC intensities as a function of the electron beam position on a $Ga_{1-x}Al_xAs$ (DH) structure. A dislocation climb network is indicated by DLD. The specimen temperature $T = -26^\circ C$ is such that the SDLTS line scan displays the DX center distribution.

Analysis /19/ of the EBIC and SDLTS signals as the electron beam is scanned over the climb dislocation network indicates an interaction between the DX center and the dislocations. By contrast, $\langle 110 \rangle$ glide dislocations in the same material do not produce appreciable changes in the DX center concentration. These results suggested that the DX center might be associated with the source of point defects needed for the rapid climb of dislocations in $Ga_{1-x}Al_xAs$ compounds. For GaP, the low temperature dislocation climb was observed in epitaxial layers grown by liquid phase epitaxy from Ga rich solutions. There also it is likely that a defect complex involving $(Ga)_i$ is involved and moves under recombination enhanced diffusion to promote dislocation climb. However, this defect complex has not yet been identified.

4. Conclusions.- This review of the low temperature dislocation climb process in III-V compound semiconductors points out areas in which more experimental information is needed. A detailed analysis of the dislocation climb structures should establish the role of the dislocation dissociations in the climb process. Further characterization of the origin and nature of the point defects involved during climb are also badly needed. The dislocation climb model discussed here requires the supersaturation of point defects of only one element of the compound and accounts well for some of the main dislocation climb features. The rapid dislocation climb at low temperature can be attributed to the recombination enhanced diffusion of point defects. Finally, evidence of an interaction between the dislocation network and a deep level donor center (the DX center) suggests that it might possibly be associated with the point defects required for dislocation climb in $Ga_{1-x}Al_xAs$ structures.

Acknowledgements.- The author wishes to thank L.C. Kimerling, R.L. Hartman, D.W. Johnston, D.V. Lang, R. Logan and M.B. Panish for their collaboration and stimulating discussions on this subject. The experimental assistance of R. Logan, A. Savage, A. Williams and H.G. White is also gratefully acknowledged.

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