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DISLOCATION MOBILITY MEASUREMENTS : AN ESSENTIAL TOOL FOR UNDERSTANDING THE ATOMIC AND ELECTRONIC CORE STRUCTURES OF DISLOCATIONS IN SEMICONDUCTORS

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Résumé - Des mesures de mobilité de dislocations dans des domaines de contrainte et de température encore inexplorés, ainsi que l'amélioration de la finesse des observations, ont permis de mettre en évidence des effets nouveaux. Ainsi les écarts à la loi de Schmid à forte contrainte et basse température, le changement d'énergie d'activation observé à haute température, l'influence des impuretés et des défauts ponctuels, de la longueur de dislocation, de l'illumination sur la vitesse des dislocations sont présentés et discutés en termes de germination et propagation des décrochements, de structure de coeur et de défauts de reconstruction. Quelques expériences nouvelles sont également suggérées.

Abstract - Thanks to the exploration of a wider range of stress and temperature, and to better instrumental resolution, new phenomena have been investigated and are reviewed here. Deviations to Schmid's law at high stresses and low temperatures, changes in the activation energy at high temperatures, influence on dislocation velocities of dislocation lengths, impurities and point defects and illumination are reported and discussed in terms of kink nucleation and migration, core structures and reconstruction defects. Some new experiments are also proposed.

1 - INTRODUCTION

The aim of this paper is to review the experimental results on dislocation mobility obtained since the Hünfeld Conference /1/ and to discuss them in terms of new theoretical approaches of the core structures.

The state of knowledge at that time could be summarized as follows :

(1) Velocities of screw and 60° dislocations were measured in Si, Ge and most of III-V compounds in a range of temperature between 0.45 and 0.65 T_m, and at resolved shear stresses between 5×10^{-5} and 10^{-3} G (G, shear modulus). These velocities were average values over travel distances of 10 to 100 µm obtained by double etching or X-ray topography on (half-)loops larger than 50 µm in diameter. A fair agreement was found between the different authors, at least in Si and Ge, and most of the data could be described by the phenomenological laws :

$$v \sim \exp(-Q(\tau)/kT) \quad ; \quad v \sim (\tau/\tau_0)^m$$

with an apparent activation energy about 1 to 2 eV, significantly stress dependent at low stresses, and a stress exponent m between 1 and 2, which could reach higher values at low stresses, mainly in Ge.
(2) A strong effect of electrically active impurities on dislocation mobility first observed by Patel /2/ was experimentally well established.
(3) There were indications that some treatments, such as annealing, could lead to

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changes in dislocation mobility, the origin of which remains obscure. Pinning was observed, as well as minor but detectable changes in velocities, including acceleration effects.
(4) TEM observations, mainly using the weak beam technique proved that moving dislocations were dissociated /4,5/. At high stresses and low temperatures, partial dislocations were shown to have different mobilities, which depend not only on their character (30° or 90°) but also on their position (leading or trailing) with respect to the sense of motion /6/. This predicts different velocities for the two types of 60° dislocations labelled 30/90 or 90/30 (with the leading partial quoted first). This difference has been detected -at the limit of resolution of the technique- at higher temperature and lower stress /3/.
(5) In the framework of the Peierls-Nabarro mechanism, different regimes of dislocation glide can be expected :

(i) Depending on the dislocation length (smaller or larger than a critical length X) kink pairs expand along the whole dislocation line, or annihilate with opposite kinks /7,8/.

(ii) Depending on the applied stress (smaller or larger than a critical stress τ_c determined by the stacking fault energy, elastic constants and orientation parameters) nucleation of double kinks on both partials is correlated or not /9/.

(iii) Localized obstacles to kink motion were considered by some theories to achieve a closer fit to experiments. Then, the temperature dependence of the obstacle spacing has to be considered and different regimes can be defined, depending on temperature and on the formation and binding energies, as well as diffusivity of these hypothetical obstacles, the nature of which still remains speculative /8,9/.

(iv) The doping effect is attributed by Hirsch /10/ to a change in the concentration of charged kinks.

Since the Hünfeld Conference, new results have been obtained, thanks to the extension of the investigated ranges of stress and temperature, and to improved experimental resolution. We shall describe in turn the specific behaviour of dislocations at temperatures above 0.75 T_m (§2.1) and at stresses higher than 10^{-3}G (§2.2), recents results on impurity effects (§2.3), new informations obtained by experiments at the scale of TEM (§2.4) and the influence of illumination on dislocation mobility (photoplastic effect) (§2.5). Conclusions and suggestions for further experiments and calculations are given at the end of the paper.

It will be noticed that this review only deals with dislocation glide. Cross-slip and climb processes have been excluded in spite of their interest in understanding some experimental data, though glide is primarily concerned at the scale of observations. It also deals mainly with Si, since most of recent experiments have been performed in this material.

It will also be noticed that some topics of interest (as the interaction of dislocations with impurities, the effect of core reconstruction defects or new results in compound semiconductors) are only briefly mentioned, since they will be treated in other lectures of the present conference.

2 - EXPERIMENTAL RESULTS AND DISCUSSION

2.1 - Dislocation mobility at high temperatures ($T > 0.7\ T_m$)

Velocities of 60° dislocations have been recently measured at temperatures up to 0.9 T_m in Si (4×10^{13}Pcm^{-3}, 2 MPa $\leq \tau \leq$ 45 MPa) by Farber and Nikitenko /11/, and in Ge ($\rho \gtrsim 15\ \Omega$.cm, 1 MPa $\leq \tau \leq$ 30 MPa) by Farber, Bondarenko and Nikitenko /12/, using pulse loading by four point bending and double etching. In both cases, a marked change in the temperature dependence of velocities was observed above a temperature of about 0.75 T_m (870 K for Ge, 1320 K for Si). Whereas, below this temperature, the results are in good agreement with litterature (Fig.1), above 0.75 T_m, the velocity can still be described by an Arrhênius law, but both the apparent activation energy Q ant the prefactor are changed. In Si, Q is increased up to 4 eV, but this is compensated by a very high prefactor (10^{12} to 10^{14} cm s^{-1}) so that the resulting velocities are larger than what can be obtained by extrapolation from lower temperatures. In Ge, the situation is quite different. The apparent activation enrgy, which depends on the stress at low stresses (< 10 MPa) below 0.75 T_m becomes stress independent above this temperature with a value of about 1.8 eV. The stress dependence, which exhibits a bend at ∿ 10 MPa at lower temperatures ($T < 0.75\ T_m$) can be described with a constant stress exponent ($m \sim 1.3$) in the whole investigated stress range (1 to 20 MPa) at 970 K. This leads, contrary to Si, to velocities lower than the values extrapolated from lower temperatures.

According to the authors, these effects are not explained by the present double kink theories. Apart from the fact that further confirmations of these results (including measurements of screw dislocation velocities) would be desirable, it seems that observations of the actual dislocation structure under stress in this temperature range could give decisive informations on the mechanism which controls dislocation motion. There are indeed some indications that dislocations under motion do not lie any more along <110> directions above 0.7 T_m /13/ and processes, like climb, might probably be taken into account.

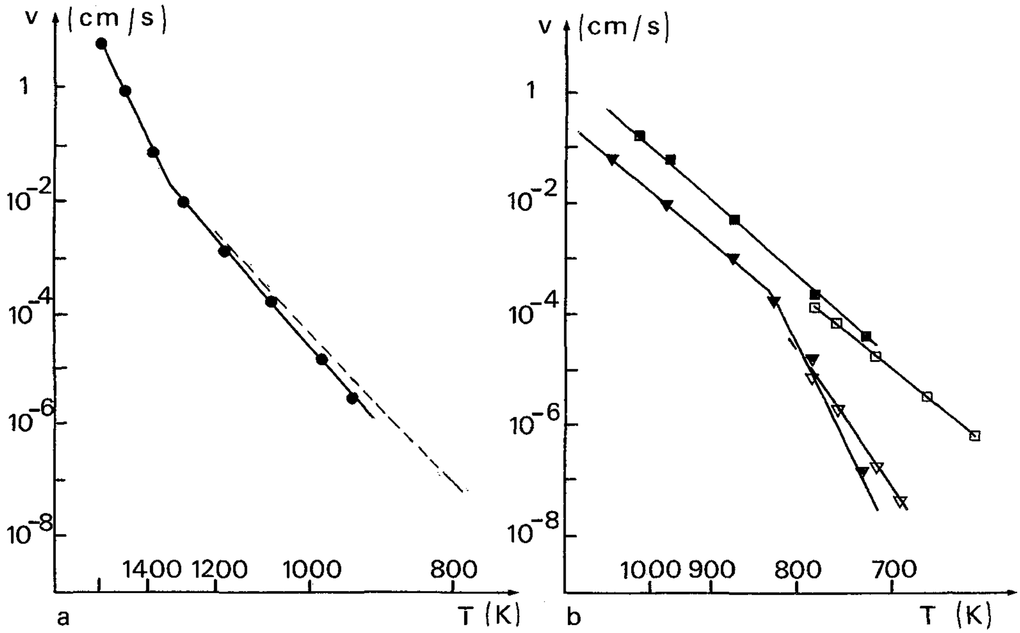

Fig.1 a) Velocities of 60°dislocations in Si ● Farber and Nikitenko/11/,---mean values from Ba Tu/14/, Fischer/15/, Sumino/19/ and George/3/. (τ = 10 MPa).
b) Velocities of 60°dislocations in Ge after Farber et al./12/ (● τ=20MPa, ▼ τ=5MPa) and Schaumburg/32/ (○ τ=20MPa, ▽ τ=6MPa).

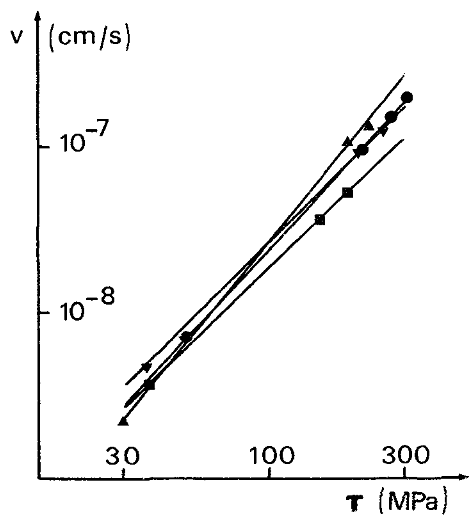

Fig.2 Velocities of 30/90 dislocations in swirl-free FZ Si measured at 693K in four different glide systems. Compression axis: [213]. ● PQ=(1$\bar{1}$1)[011], ■ PK=(1$\bar{1}$1)[110], ▼ UK=(111)[01$\bar{1}$], ▲ KQ=($\bar{1}$11)[101] ; after Alexander et al./16/.

2.2 - Dislocation mobility at high stresses ($\tau > 10^{-3}G$)

Alexander and coworkers investigated dislocation mobilities in Si in the high stress range using two complementary approaches. (i) Velocities of total dislocations were measured in swirl free floating zone Si /16/. (ii) From measurements of dissociation widths on dislocations frozen-in under load, relative mobilities of the various partial dislocations were also obtained /6,17/.

From hardness indentations, Kisielowski /16/ could develop by compression at 923 K half-loops belonging to different slip systems. Their velocities were measured at 693 K by compression and double etching. The results can be summarized as follows:

(i) The dependence of the velocity v on the *resolved* shear stress τ can be described by $v = v_0(\tau/\tau_0)^m$ in the range 30 MPa $\leq \tau \leq$ 300 MPa (see Fig.2)

(ii) The parameters v_0 and m depend not only on the type of the dislocation, but also, for a given dislocation type, on the glide system to which the dislocation belongs ($1.2 \leq m \leq 2.2$), which means that Schmid's law is not obeyed in Si ! This effect is more pronounced as stress increases, which explains that it was not detected by George /3/ who also observed loops of several glide systems, but at higher temperatures and lower stresses.

(iii) An empirical expression is given for the stress exponent :

$$m = m_0 + m_1 = - 7.2\ (1 - |\vec{b}\ .\ \vec{c}|)^2\ F_n/b\sigma + m_1$$

where m_1 depends only on the dislocation type (character and order of partials), and m_0 is calculated from the angle between the Burgers vector $\vec{b}$ and the compression axis $\vec{c}$ (unit vectors) and from the normal component Fn of the force acting on the dislocation, normalized by $b\sigma$ (σ=nominal compression stress). The values of m deduced from this formula agree fairly well with those measured at higher temperatures and lower stresses. Moreover, as pointed out by Alexander /18/, it is also possible, using the fitted values for v_0, m_0 and m_1, to predict that the differences between the velocities of the 30/90 and the 90/30 dislocations must be smaller in tension than in compression, which is in agreement with observations.

On the other hand, Weiß /18/ noticed that the dissociation widths of similar dislocations in different glide systems (in the same conditions of stress and temperature) could not be explained by mobility ratios which would depend only on the type of the partial dislocations. Assuming a different lattice resistance for the two partials and writing the balance of forces acting on them, one obtains a dissociation width :

$$d = \frac{d_0}{1-(f-\delta)\ \frac{\tau b}{2\gamma}}$$

where d_0 is the dissociation width at zero stress, γ the stacking fault energy, $f = F_2-F_1/F_2+F_1$, $\delta = R_2-R_1/R_2+R_1$, F_i and R_i being respectively the externally applied force and the total force (equal to the lattice resistance in the steady state) felt by the partial dislocation i. This analysis gave interesting indications on partial "mobilities", defined through a linear relationship between stress and velocity. This approach was improved by Heister /18/, who assumed that the experimental power law $v = v_0(\tau/\tau_0)^m$ followed by the total dislocation was also valid for partial dislocations. The total force on the total dislocation can then be written :

$$R = \tau b = \tau_0\ b \left(\frac{v}{v_0}\right)^{1/m} = R_1 + R_2 = R_{0_1} \left(\frac{v}{v_{0_1}}\right)^{1/m_1} + R_{02} \left(\frac{v}{v_{02}}\right)^{1/m_2}$$

which leads necessarily to $m = m_1 = m_2$ for consistency. Using the experimental values of m, and fitting two parameters for each total dislocation, most of the splitting widths for five glide systems could be found within 10 % from the experimental values. However, as pointed out by Alexander himself, the expression $v = v_0(\tau/\tau_0)^m \exp(-Q/kT)$ has not direct physical meaning. It seems then difficult to find a physical support for the statement $m=m_1=m_2$, all the more is a part of m probably accounts for reduction by the applied shear stress of the energy to be thermally actived (which is usually described by a linear prefactor m=1, although this is only valid at low stresses $\sigma b^3 << kT$), the second part being related to a modification of the prefactor (via changes in details of the core structure) by the non-shear components of the stress tensor. The satisfactory agreement between the model and the experiment is perhaps an indication that both partials might be concerned in a similar way by the two effects. It must be noticed however that the present description is based on the steady state assumption, although some scatter can be observed in measurements of dissociation widths.

2.3 - Influence of impurities (and point defects)

The interactions between dislocations and *specific impurities* intentionally introduced in controlled concentrations will be treated by Sumino in the present conference. The best studied case for non doping impurities is that of Oxygen in Si. The main results of Sumino and coworkers /19/ concerning dislocation mobilities are the following : dislocations at rest are readily pinned by oxygen atoms, but the velocities of both screw and 60° dislocations are similar in floating zone (FZ) (O $\leq 10^{16}$ cm^{-3}) and in Czochralski (CZ) Si (O $\simeq 4\times10^{17}$ cm^{-3}) at 3 MPa $\leq \tau \leq$ 30 MPa and 875 K $\leq T \leq$ 1045 K. In addition, moving dislocations are gradually blocked in CZ Si if the shear stress is decreased below 3 MPa.

Curious effects are noticed even in the "purest" FZ material ; they are more critical as far as velocity measurements are concerned, and could well be attributed to *residual impurities*. A first example is given by Bondarenko et al./20/ who observed "traces" left in the slip planes behind moving dislocations in Si. Such traces were detected by TEM behind fast dislocations (v 10^{-3} cm.s^{-1}) generated at identations (load 5×10^{-2} to 10^{-1}N, 298K $\leq T \leq$ 973K). They appeared at the original surface, but could not be explained by surface steps nor by an oxide layer which should protect the surface against slip penetration. Behind dislocations with slower velocities (4 point bending, 10MPa $\leq \tau \leq$ 40MPa, 823 K $\leq T \leq$ 1023 K), traces could not be seen by TEM, but were revealed by Sirtl etch and observed by SEM and microinterferometry. They were resolved as ridges, whose height was the same for every individual dislocation in a given crystal. In both cases traces were more intense and stable in heavily doped material, though they could be observed in weakly doped FZ Si ($\sim 10^{15}$ cm^{-3}).

Bondarenko et al. /21/ also looked for a possible influence of prestraining conditions (temperature, dislocation velocities and distance moved to bring the dislocations to their starting positions) on dislocation mobility. As in /19/, these treatments (as well as annealing under zero stress or modifications of the cooling rate of crystals) influenced mainly the starting stress τ_{st}, i.e the minimum stress required to put dislocations in motion, but did not affect dislocation velocities at $\tau > \tau_{st}$.

Another example is given by Alexander et al. /16/, who observed etch mounds instead of etch pits at dislocations produced at 923 K in swirl-free FZ Si. These mounds appeared only at some dislocations and seemed to be in limited number, so that their proportion decreased when the dislocation density was increased. They were attributed to atoms of unknown nature diffusing from the bulk to the dislocation, then enhancing their resistance to etching, although fewer mounds were observed in conventional FZ Si, and even fewer in CZ Si. They were not observed at 693 K, but a measurable slowing down of 60° dislocations during motion was detected (screws were not investigated in this respect). Fig.3 shows the velocity of dislocations as a function of the distance they have moved. The decrease in velocity is observed as far as the motion can be followed, and is stronger for 90/30 than for 30/90 dislocations. However, this effect is small enough for the above mentioned results (averaged over the first 100 µm) to hold (deviations to Schmid's law for instance).

It could be noticed here that some of these effects attributed to impurities might perhaps be explained as well by considering intrinsic point defects produced by the deformation /22/ or those core defects predicted by Jones /23/ as a result of reconstruction of atomic bonds in the core of partial dislocations.

2.4 - Small scale observation of dislocation glide

The resolution of double etching techniques or X-ray topography is about 2.5 µm. It was thus quite appealing to attempt observations of dislocation glide at the much finer scale obtained by TEM. In situ observations have been performed mainly in Si, using either a high temperature straining stage /24,25/ or simply by annealing dislocation configurations out of equilibrium (e.g. single partial motion under the stacking fault tension in widely dissociated dislocations /26,27/). In these experiments, although an accurate control of temperature is difficult, a confidence range of $\sim$ 10 K can be achieved after calibration of the heating stage by means of a suitable phase transition. The local stress can be measured using the curvature of dislocations at the bend between two adjacent <110> segments (line tension anisotropy should be taken into account, and special care is needed at very ligh stresses, where the order of magnitude of the radius of curvature becomes closer to the dissociation width (H. Gottschalk, private communication).

Straight dislocation segments are observed to glide smoothly at 793 K $\leq$ T $\leq$ 888 K, 70 MPa $\leq \tau \leq$ 550 MPa /25/ in agreement with /24/. The velocities show reasonable agreement with data extrapolated from lower stresses, at least for dislocations longer than $\sim$ 0.4 μm. This is also true for unpinned partial dislocations at 693 K, τ = 280 MPa /27/.

Clear evidence for a length effect on dislocation mobilities was also found. Louchet observed that dislocations shorter than a critical length X had a velocity which increased with their length L, although the exact v(L) dependence was not established (Fig.4). X seems to depend slightly on stress and temperature (X $\simeq$ 0.4 m at $\tau \simeq$ 90 MPa, T $\simeq$ 873 K, X $\simeq$ 0.1 μm at $\tau \simeq$ 550 MPa, T $\simeq$ 813 K). Consistently, Hirsch et al. /27/ observed the length effect for short dislocation segments (L < 0.2 μm at $\tau \simeq$ 280 MPa, T $\simeq$ 593 K). These values of X are well below the resolution of X-ray and etching techniques and are not larger in FZ Si than in CZ.

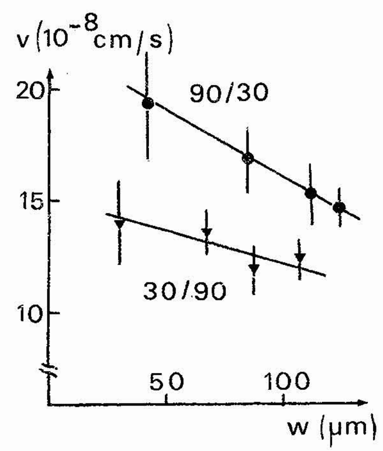

Fig.3 Slowing down of moving 60° dislocations in swirl-free FZ Si. W: distance moved. After Alexander et al. /16/.

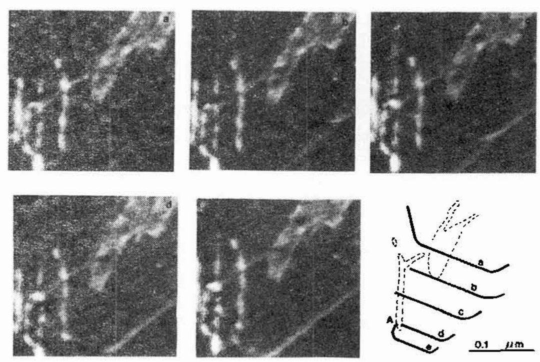

Fig.4 A dislocation moves from positions (a) to (d) at constant velocity. After pinning in (d) on obstacle A, it gets shorter and slows down (d) to (e) (T=813K, τ=550MPa, 0.40sec from one picture to the next). After Louchet /25/.

Intersections of mobile dislocations with the forest have also been observed (/25/ and Louchet, unpublished results). The single jogs thus formed are associated to large visible cusps, which can either straighten and readily catch up the straight part of the dislocation, or persist and follow the dislocation motion. In most cases, jogs move conservatively (i.e. they are dragged along straight lines parallel to the Burgers vector), but non conservative motion is also observed between 770 K and 870 K.

Pinning by strong but invisible obstacles and subsequent unpinning were also observed on 90° partials, but not on 30° partials /27/. These latter results are in agreement with those of Gottschalk /26/ and of Desseaux and Bourret /28/. It is clear however that weak obstacles giving rise to cusps smaller than the image width (50 Å) are not detected by these techniques.

The transition mentioned above, between a length dependent regime (L<X) and a length independent one (L > X) can be attributed to obstacles to kink motion, resulting in a mean free path of kinks X/2. This can be obtained either by localized obstacles (with an average separation X along the dilocation line) or by a high migration energy W_m for kinks (secondary Peierls potential). In this latter case, the mean free path is directly related to kink-kink collisions. Most of the known localized obstacles (jogs, radiation induced defects,...) can be ruled out for several reasons (visible cusps, similar behaviour for FZ and CZ Si), and there is reasonable evidence that kink motion is controlled by a high migration energy. Under this assumption, Hirsch et al. /27/ and Louchet /25/ could estimate kink velocities and migration energy of kinks, using Hirth and Lothe theory /7/. W_m is found about 1.2 to 1.3 eV. The total activation energy for dislocation motion has also been estimated through Hirth and Lothe's expressions of velocities to $F_{dk}^*+W_m \sim 2$eV in the regime L<X /27/, where F_{dk} is the saddle point energy for double kink nucleation. This leads to $F_{dk}^* \sim 0.7$ to 0.8 eV. However, the theoretical prefactor for dislocation velocities in the regime L>X differs significantly from the experimental one, as discussed below, and this is likely to be true for L<X also. It seems then more reasonable to derive F_{dk}^* from the experimental activation energy measured in the regime L>X. One obtains then : $F_{dk}^*/2 + W_m \sim 1.9$ eV and hence $F_{dk}^* \sim 1.2$ to 1.4 eV. The single kink energy Fk can be derived from $F_{dk}^* \simeq 2F_k - \sqrt{G\tau b^6/2\pi}$ /7/. The stress dependent term is not negligible at the high stresses involved in TEM work, and one finds $F_k \sim 0.9$ to 1 eV.

Another argument in favour of a larger W_m might be given by the experimental prefactor which is found to be about 35 times larger than the theoretical value $\nu_D \tau b^4/kT$ expected from Hirth and Lothe's theory, but 10 times smaller than the expected value for localized obstacles with a separation X. Such a discrepancy could be attributed to an entropy factor exp(S/k) in the former case (with $S \sim 3.6$ k), which may arise partly from an elastic contribution, but mainly from a core contribution. This explanation cannot stand in the second case, since exp(S/k)>1. (There is, however, some doubt on the validity of the theory at high stresses).

A different theoretical approach to dislocation mobilities has been given recently by Heggie and Jones /23/, in which "antiphase" reconstruction defects of the dislocation core (solitons) are assumed to help both kink nucleation and migration, since it seems easier to transfer an existing dangling bond than to break a bond. The consequences of this approach are given by Jones in this conference. However, the form of Hirth and Lothe's equations is unchanged, and so are the conclusions, except that W_m and F_k have somewhat different meaning.

2.5 - Enhancement of dislocation mobility by optical excitation

The effect of illumination on dislocation velocities (photoplastic effect) has been demonstrated recently in Si by Küsters and Alexander /29/, using similar techniques as in /16/. Straight dislocation half-loops were created by prestraining (T=923K, σ=63 MPa) in swirl-free FZ Si (2×10^{12} Bcm^{-3}, carrier lifetime at 300 K:5ms), and moved at high stresses and at temperatures low enough to obtain a noticeable change in the electrical carrier density by illumination. During deformation, one part of the crystal was illuminated by a Nd-YAG laser ($h\nu$=1.17 eV, P=1 $W.cm^{-2}$). The observed dislocations were situated in the range of 50 µm below the surface, in which most of the incident light was absorbed. Screws and 30/90 dislocations were observed to move faster under illumination, whereas the velocity of 90/30 dislocations was not affected significantly. As shown in Fig.5, the apparent activation energy for the motion of 30/90 dislocations at $\tau \simeq 300$ MPa is 1.82 eV in the dark, and is reduced by an amount of 0.68 eV under illumination. The Arrhénius plot shows that the prefactor is also reduced and that the effect vanished at $\sim$710 K. In the case of screw dislocations, the decrease of the activation enrgy is about 0.56 eV. A careful control of temperatures in the illuminated area showed that they were not increased by more than $\sim$2 K, and this cannot account for the magnitude of the observed effect (in addition, an increase of temperature would have also increased the velocity of 90/30 dislocations !). In order to compare this situation with the doping effect, a n-doped specimen (5.5×10^{16} Pcm^{-3}) was deformed in the same conditions without illumination. It emerged that the velocities of both types of 60° dislocations were increased, in clear contrast with the photoplastic effect. A behaviour similar to the photoplastic effect has also been found by Maeda and coworkers /30/ in GaAs under electron irradiation.

The photoplastic effect was attributed to non radiative electron-hole recombinations at dislocations, which are known to act as very active recombination centers at these temperatures. The vibrational energy released by recombination can help either double kink nucleation or kink migration processes, since associated energies are expected to be larger than the observed energy reduction. However, if the recombination can occur at localized defects as kinks /31/ it might be more difficult on a nucleating double kink. This argument is in favour of an enhancement of the kink velocity, except if double kink nucleation occurs at localized defects such as solitons, as suggested by Heggie and Jones. A careful investigation of any change in the mean free path X of kinks under illumination should be quite instructive in this respect.

3 - CONCLUSIONS AND SUGGESTIONS FOR FURTHER PROGRESS

From the set of results reported above, it seems obvious that the conclusiveness of an experiment in often submitted to a careful control of a lot of parameters, some of which had sometimes been forgotten in the past. It is also clear that several effects of interest can be detected only if velocities are measured with an accuracy better than $\sim$15 %. Hence, a first requirement is a good control of temperature and stress which have a drastic influence on velocities. For instance in Si, an uncertainty $\Delta v/v \sim 20$ % in velocities can arise from a $\Delta T \sim 4K$ or $\Delta\tau/\tau \sim 10$ % (and better control of stress cannot be easily achieved !). In addition, and this is mainly true at high stresses, results from different authors cannot be successfully compared, unless the glide systems and all the components of the stress tensor are known. Nevertheless, experiments in which T or τ cannot be accurately determined can give valuable results (e.g. relative values of mobilities of different types of dislocations), provided that fluctuations do not occured at the scale of observations.

A second point which has not been mentioned before and has to be controlled is the influence of the free surface, for at least two reasons. (i) Image forces may enhance or retard kink nucleation at the surface (geometrical effect). (ii) The surface may act as a sink (or a source) of impurities or point defects. This latter effect could help in achieving more quickly a steady state regime through pipe diffusion towards the surface of impurities scavenged by dislocations. In contrast, internal loops would accumulate more and more impurities. This might explain the different behaviour of internal loops and half-loops observed by Milevskii and Smolskii /33/, though George /3,34/ did not notice such an effect at the same stresses and temperatures (but probably in a silicon crystal of different purity).

As regards geometrical surface effects, it is possible to get rid of them, even if a surface technique such as etching is used for measurements. This was convincingly proved by Kisielowski /16/ and Küsters /29/ who were able (i) to determine the orientation of the emerging dislocations from the shape of the pits, and (ii) to detect bends of dislocations below the surface, to measure their depths and to obtain bulk velocities by successive light polishing and re-etching of the surface. On the other hand, we believe that the asymmetry of dislocation mobility in Si and Ge reported by Nikitenko et al. /35/ is essentially a geometrical surface effect, as demonstrated by the shape of the pits, although it is also clear the different "states" of impurities in the dislocation core are certainly involved in their experiment.

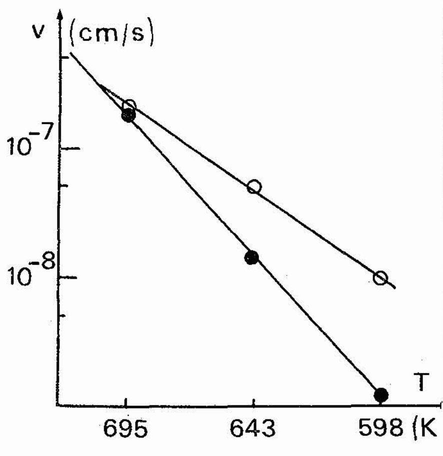

Fig.5 Velocities of 30/90 dislocations with (o)and without (•) illumination in swirl-free SZ Si. τ=294MPa, compression axis [213], glide system (111)[011]. After Küster /29/.

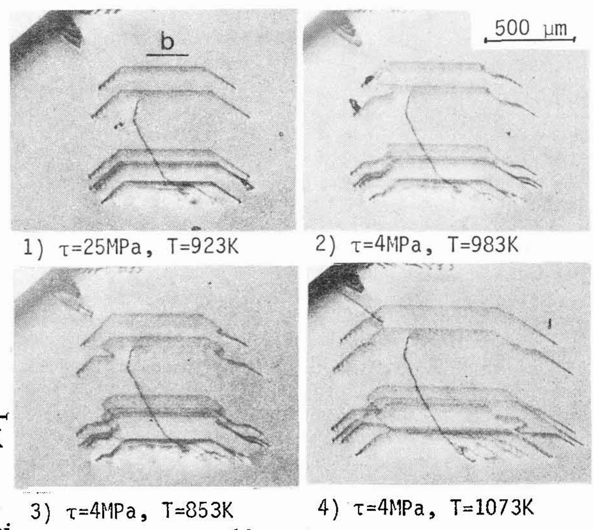

Fig.6 FZ Si (10^{14}Pcm^{-3}). At low stress, 60° dislocations have been pinned (except at the surface from 1)to 2)), while the screw segment moved, creating free 60° segments. This lead to nearly complete unpinning from 3)to 4).X-ray topography (George, unpublished).

Third, it is necessary to have a good knowledge of tne whole thermomechanical history of the crystal and of the investigated dislocations. This history starts from the growth conditions of the crystal, but the conditions of dislocation loop expansion are of importance, and this is complicated by the fact that in most cases the length of mobile dislocations increases continuously. It seems likely that different deformation conditions lead to somewhat different core structures of dislocations. The question then arises of the possible evolution of the starting core structure when the dislocation length is increased under different conditions. This might be correlated to the still opened /28/ glide vs shuffle alternative, and also to the existence and concentration of reconstruction defects (solitons, vancancy soliton /23/ or impurity soliton complexes).

These last two requirements for an "ideal" experiment are in favour of in situ observations in the bulk, which can be performed by X-ray topography and in several cases by TEM. X-ray topography has suffered until recently of very long exposure times. Nearly real-time recordings are now made possible by high power sources /36,19/, and in particular by synchrotron radiation. Existing high temperature straining stages /37/ do not allow an accurate control of temperature but they should be improved in this respect when new "dedicated" synchrotrons with smaller source size are available. X-ray contrast is also very sensitive to stress gradients (elastic strains), and this is an useful way of controlling them. Short wavelength transmission topography would also allow observations of crystals thick enough to be deformed in compression (at least in Si). Fig. 6 shows an example of the information that can be obtained from bulk observations : pinning of the emerging 60° dislocations has been partly released at the surface (possibly by out diffusion) and by the motion of new segments created by the non-pinned screw segment.

As regards TEM, a further step could be achieved by in situ straining at the resolution of weak beam techniques, in particular for observations of the motion of dissociated dislocationsin the neighbourhood of the transition between correlated and uncorrelated nucleation of double kinks, in connection with high resolution observations. In such experiments care must be taken to find the best compromise between reasonable penetration and minimization of radiation damage, which will depend on temperature. Radiation defects can behave as obstacles to kink motion, thus hindering dislocation movement, and this is more likely to happen at low temperatures, when diffusion is more difficult. On the other hand, if temperature is low enough, the recombination of radiation induced electron-hole pairs can increase significantly dislocation velocities, as reported §2.5, and this is perhaps the reason why the velocities measured by Hirsch et al. /27/ by in situ annealing at 693 K are found to be larger than extrapolation of measurements performed at higher temperatures /3,25/.

This leads to the relation between electrical properties and core structures of dislocations, which should be investigated on the same dislocations, or at least on dislocations which have the same history than those used for mobility measurements. This has already been attempted in /21,32/. Thanks to new microscopic techniques (CL,EBIC,SDLTS...) it may be possible to set up experiments which should alternate measurements of mobility and other physical properties of the same dislocation. A nice exemple has been given by Maeda et al. /30/ who coupled in situ straining experiments and CL SEM observations. An interesting point to investigate should be to compare the electrical response of curved vs straight dislocations by changing the shape of a given dislocation by stress jumps at a constant temperature.

The general agreement for a high value of the migration energy of kinks, as well as the idea that solitons could influence dislocation motion through an enhancement of nucleation rates and/or migration velocities of kinks (if checked by experiments), might be applied to existing theories, and in particular to the behaviour of dissociated dislocations in the two regimes of uncorrelated and correlated motion of partials. This could perhaps help in understanding the particular features of dissociated dislocation motion and the deviations to Schmid's law observed at high stresses. More accurate calculations of core structures in saddle point configuration should be also highly desirable in this respect (as well as for a better description of doping and photoplastic effects), and might perhaps help in improving the present knowledge of potentials. A deeper investigation of nucleation, equilibrium concentration and stability of solitons should also be of interest.

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