GENERAL INTRODUCTION

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Summary.

The necessity and difficulty of marrying together micro and macroscopic aspects in problems of plasticity are stressed and analysed on some examples. The importance of a mesoscopic scale is suggested.

Introduction.

When I was writing in 1953-4 my book on Dislocations, my friend B. Jaoul was composing, in the same lab., his own contribution to the mechanics of plasticity. It seemed to us at the time, and I still strongly believe, that the micro and macroscopic points of view from which these two books were starting should be ideally joined together.

This is indeed the spirit in which this special issue of the Journal of the Mechanical Behaviour of Materials has been composed. But the road towards this goal has been especially bumpy ; and we are still far from a perfect matching of these two approaches. This is what I want to stress, taking a few important examples :plastic deformation of pure metals, plastic relaxation of cracks, strain localisation in alloys.

I - Plastic deformation of pure metals.

Here, the microscopic description starts from a network of dislocations, either a simple Frank network or a polygonised mosaïc structure, purporting to describe a well recrystallised single crystal. When the stress is applied at moderate temperatures, the most mobile dislocations under the largest resolved shear stress start acting as Frank Read sources, and produce the first slip bands.

A historic case, already considered by G.I. Taylor before the last war, is that of metallic face centred cubic (FCC) single crystals under traction. One observes a complex succession of stages, characterised by different hardening rate and temperature dependences :

-easy glide by development of parallel slip bands on the most favourable slip system of the usual [110] (111) kind (stage I).

- stage II with strong and rather constant and temperature independant hardening rate, when crystal rotation under traction activates more slip systems.

- temperature dependent and decreasing hardening rate (stage III), connected with crossslip.

- finally new slip on non close packed planes can occur at the largest strains before striction.

This scheeme applies to many metals near room temperature. But it is further modified in two ways. At very low temperatures, stage III becomes too difficult and is replaced by the development of stacking faults and twinning. At high temperatures, stage III replaces stage II completely and further slip stages are replaced by climb.

The corresponding dislocations structures developed are very inhomogeneous on a microscopic scale :

- in stages I and II, slip is concentrated in narrow slip bands which remain a micron or more apart.

- in stages III and further, a polygonised structure develop by recombination of screws by crossslip. The final size of this structure decreases with increasing stress, but remains at least of the same order of magnitude.

Finally the mechanisms of hardening in each stage are understood qualitatively ; the stresses at which they appear and the activation energies involved are often analysed in details. The special role of the stacking fault energy involved in dislocation splitting is very clear. But the actual values of hardening rate in each stage cannot be predicted convincingly in any model. Thus in stage II so characteristic of FCC metals, the stress can be at least roughly related to the average density of dislocations, as if they were produced and cutting each other at random ; and the large hardening rate can be related to the production of Cottrell Lomer locks at the intersection of different slip bands. But no convincing model has been made of the actual value measured, of order

$$d\sigma/d\varepsilon \cong \mu/300.$$

Going from simple traction to more complex deformations does not, of course, simplify matters !

- successive tractions along different directions introduce latent hardening, often larger than normal hardening for simple traction. This is especially clear in stage I where the slip bands produced in latent hardening have to be crossed by the normal slip bands of the stage.

- but successive tractions along opposite directions soften the material considerably. In this Bauschinger effect, and especially in stage II, it is much easier to develop slip in a direction opposite to previous slip than to continue slip in the primitive direction. This is in keeping with the idea that in stage II dislocations of the slip bands are somewhat piled up on Cottrell Lomer locks.

- in plastic flexion or torsion, a minimal density of dislocations must be developped to preserve the crystalline continuity. But the dislocation densities actually developed are usually much larger and more complex.

The hardening produced is therefore highly anisotropic. The origin of this anisotropy is again complex, and no convincing microscopic model has been developed to describe it.

More complexities arise in polycrystals of these metals and other analyses, admittedly somewhat simpler sometimes, apply to other crystals.

But it seems clear that, compared with this wealth of different microscopic behaviours, the tentative macroscopic descriptions are rather poor and at best very schematic.

One can stress however the following points.

a - There are fundamentally two regimes of deformation, depending on whether the hardening is mostly thermally activated or mostly temperature independent.

An extreme regime at high temperatures and low stresses is a quasiviscous one, where the major part of hardening is temperature and time dependent. As in a viscous fluid, the stress necessary to develop straining depends first on strain rate $\dot{\varepsilon}$, and not so much on the integrated strain ε, or its history. Then, from

$$\dot{\varepsilon} \doteq f(\sigma, T),$$

one deduces

$$\sigma \doteq \phi(\dot{\varepsilon}, T).$$

But in most conditions, there is an amount of solid friction σ_i which appears in the corresponding equations :

$$\dot{\varepsilon} \doteq f'(\sigma - \sigma_i, T),$$

thus

$$\sigma \doteq \sigma_i + \phi'(\dot{\varepsilon}, T).$$

σ_i is dominant for instance in stages I and II of FCC metals. In this solid friction regime, hardness varies little with temperature and time ; quasistatic conditions ($\sigma \simeq \sigma_i$) prevail, except naturally for shock waves. This is also the regime of large hysteresis, such as shown in the Bauschinger effect : the hardness produced σ_i , as well as the corresponding dislocations content, depend on the plastic history of the material ; there is therefore no possibility of defining a potential strain energy only function of total stress or strain. Macroscopic models then necessarily refer to definite ways of straining.

b - The extreme variety of microscopic mechanisms and corresponding macroscopic behavious makes any attempt to schematise with simple phenomenological laws dependent on fairly narrow limits in physical conditions. Extrapolation outside these limits can be dangerous.

c - What is obviously missing in most cases is a good model at a mesoscopic scale, i.e. a scale between those of individual dislocations and of the crystal as a whole. It is clear now that the concept of average dislocation network with a constant density is not sufficient to describe stage II of FCC metals. The strong inhomogeneities observed at micron level (slip bands, polygonised structures) make it imperative to study in more details the structures and behaviour under stress and temperature of units of this kind. This has only recently started for polygonised boundaries in creep of FCC metals, using transmission electron microscopy ; very little has been done recently on slip bands of the same metals, except in the extreme case of fatigue where they obviously play a leading role.

2 - Plastic relaxation of cracks.

The study of crack propagation offers a good example of the progressive and necessary adjustement between micro and macroscopic descriptions.

Cracks were first analysed in brittle materials such as glass, using only elastic distorsion models. This lead to the famous Griffith criterion for the propagation under stress σ of a crack of length L :

$$L \geq K \sigma^{-2}$$

where K is a factor related to the elastic constants and surface tension of the medium and to the geometry of crack and sample.

In less brittle materials, it was known that larger stresses were necessary. It was observed that, during its propagation and at rest, the tip of a crack was plastically relaxed. The energy of plastic relaxation should then be added to surface tension in the value of K, increasing it by a large factor indeed. A detailed analysis of such relaxation was started, in terms of dislocation loops emitted from the crack front or of dislocation sources activated in its neighbourhood.

But, under industrial pressure connected with the failures of the Comets planes, a shortcut was developed by analysing the plastic relaxation in terms of a macroscopic plasticity model. This required the knowledge of phenomenological constants (such as K) which were fitted with experiment. A fairly accurate prediction of which of the preexisting cracks should grow to failure was thus established for given materials of various forms and crack contents, used under specified conditions. However further studies have made clear, as expected, that no safe extrapolation could be made to other conditions of temperature, speed or environmental atmosphere or a fortiori to other materials. A return to more microscopic studies has been under way more recently. Instabilities of plastic relaxation mechanisms around a moving crack are, in parallel, studied for some polymeric materials where the viscous relaxation is hopefully better mastered. The two connected questions

- under what conditions does a moving crack start to relax plastically ?
- under what conditions does a stationary and plastically relaxed crack start moving again ?

require indeed, in metallic materials, both a microscopic study of the nucleation and growth of plasticity and, again, a mesoscopic study of possible instabilities.

3 - Strain localisation in alloys.

In solid solutions, a localisation of slip on a macroscopic scale can arise under specific conditions of concentration, temperature T and strain rate $\dot{\varepsilon}$: this is the wellknown blue brittleness of steel and the Portevin-Lechatelier effect of aluminium base alloys.

Blue brittleness arises as a sharp peak in the stress strain curve near the macroyield limit, followed by a plateau of constant stress and inhomogeneous strain. A Luders band of deformation thus nucleates and then propagates through the polycrystalline sample, separating two regions : the one with little strain is eaten up progressively to make room for the one with the total strain corresponding to the end of the plateau.

Portevin Lechatelier effect looks essentially like a repeated yield peak+ plateau effect, more or less well defined depending on the testing condition (T, $\dot{\varepsilon}$, rigidity of the testing machine) ; it also corresponds to the propagation of more or less well defined deformation bands across the (polycrystalline) sample. Repeated yield effect can also be observed in steel under some conditions.

Studies at two levels are obviously required :

- macroscopic studies of the nucleation and growth of deformation bands.
- microscopic studies to explain when such instability is observed.

Initial studies concentrated on the second type of question : how could pinning of dislocations by solute atoms explain the yield point and its eventual repetition ? It can be noted that, at the time, the same pinning by C or N in Fe involved in the development of yield points in steels was also incorrectly invoked to explain the temperature variations of yield point ; it is now known that a pure lattice friction on screws dominates the second effect. It can also be noted that the lower and upper limits in a (T, $\dot{\varepsilon}^{-1}$) diagram for observing repeated yield were correctly attributed respectively to different conditions :

- below the lower limit, solute atoms are too slow (even with the help of vacancies created by straining, in Al base alloys) to recapture dislocations generated in propagating deformation bands during the aging time of the plateau.
- above the upper limit, solute atoms are too few along the dislocations or, more usually, too mobile to act as pinning centers.

More recently, such phenomena were considered in a phenomenological way as instabilities in steady state regimes. As in other examples such as hydrodynamical instabilities in Rayleigh Benard rolls of fluids under thermal gradients, a purely time dependent model equation was considered, neglecting spatial variations. Although this is perhaps sufficient to describe approximately the high (T, $\dot{\varepsilon}^{-1}$) limit, where a continuous transition probably occurs, it is certainly not sufficient in the lower (T, $\dot{\varepsilon}^{-1}$) regime, where deformation bands are well defined and separated in space.

To describe whole range of (T, $\dot{\varepsilon}^{-1}$) where the instabilities are observed, it is clear that one must introduce spatial inhomogeneities, and also that one is far from a small perturbation regime. To study the lower limit in (T, $\dot{\varepsilon}^{-1}$), the older point of view of characterising the nucleation and propagation of separate deformation bands would be more appropriate. It is for instance fairly clear that this range should be sensitive to the type of stress applied, to the form of the sample, to regions of stress concentrations near the grips. No systematic work has been done recently along these lines.

Conclusions.

The three examples chosen show fairly conclusively that to master the plastic properties of crystals, a double difficulty must be first surmounted : the behaviour of individual dislocations as well as the properties of more complex collections of such dislocations on a mesoscopic scale must be understood in some details. The use of macroscopic equations of continuum mechanics can then be a guide to study inhomogeneities and instabilities. But it is mostly the mesoscopic scale of slip bands, polygonised walls and deformation bands which needs more systematic studies of each individual entity.