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On the Inhomogeneity of Plastic Deformation in Metals

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Summary. Microscopic inhomogeneity of plastic deformation in metals and alloys due to crystal anisotropy and the presence of crystals of different structures is described. The importance of the inhomogeneity in relation to the theory of the deformation of crystal aggregates is discussed. It is shown that macroscopic inhomogeneity is produced by common methods of deformation.

1. Introduction.

The physical explanation of mechanical properties such as ductility and strength of metals is a problem which has attracted the attention of scientists for many decades because of its great practical importance. Engineers and technologists devised methods by which the resistance of materials to deformation could be tested under what appeared to be the simplest conditions of stress. Physicists became interested in the problem about thirty years ago when it was possible to grow large metallic single crystals. These were subjected to the usual mechanical tests, and we now have a fairly complete knowledge of the behaviour of single crystals under tensile, compressive, torsional and reversed stresses at various temperatures and various rates of application of stress. A theory has been developed during the last fifteen years which satisfactorily explains the main features observed on single crystals, and even though some details are still under discussion, there cannot be any doubt as to the validity of its basis. The fundamental assumption is that slip in crystals occurs by the movement of a special type of lattice defect, the dislocation. In other words, slip starts in a restricted volume of the crystal and does not occur simultaneously along the whole slip plane.

Although slip is thus an inhomogeneous process, its macroscopically observable effects are in most cases homogeneous; e. g. the stress-strain curve obtained in a tensile or compressive test of a single crystal is essentially a smooth curve. The question then arises as to how the knowledge of the tensile curves of the single crystals helps to explain the behaviour of polycrystalline metals under stress. The tensile curve of a single crystal is, of course, obtained under homogeneous stress conditions producing quasi-homogeneous strain, and, if a crystal of a polycrystalline aggregate were also subjected to homogeneous stresses and deformed homogeneously, the behaviour of the polycrystal could be calculated. Such a calculation has in fact been carried out for the tensile curve of aluminium by G. I. TAYLOR[1]), and in spite of the assumption of homogeneous deformation of the crystals, the agreement between calculated and observed curves is very good. However the well known fact of increase in strength with decrease in crystal size cannot be explained by the theory. The inhomogeneity of deformation of the crystals in an aggregate has several causes, and in order to gain a greater knowledge of its details, several investigations have been carried out in our laboratory. Some results of these investigations will be described in this paper, grouped according to be the origin of the inhomogeneity.

2. Anisotropy of Elastic Limit.

The stress which has to be applied to start plastic deformation in a crystal depends on the relative orientation of the slip plane to the direction of stress. If the crystals in an aggregate were free to deform independently of each other, they would start deforming at different stresses. They are hindered in doing so because crystals of different orientations meet at the boundaries, and except in the case of deformation at elevated temperatures under stresses applied for long periods, the crystals always remain in contact at the boundaries and no gaps or holes are formed there. If we carry out a tensile test on a specimen consisting of large crystals with transverse boundaries, the central portions of the crystals are so far from the boundaries that they suffer no interference and deform almost in the same manner as free single crystals, i. e. to different extents and starting at different stresses. In the neighbourhood of the boundaries, on the other hand, there is a strong interaction between the crystals and the mode of deformation is changed.

This change has been investigated by HARGREAVES[2]) using flat aluminium specimens with a grain size of approximately 2 cm. A number of small indentations was placed on a line crossing the grain boundaries, these indentations serving as gauge marks and for the determination of the hardness which was found to be constant in any grain before deformation but varying from grain to grain. The specimens were then stretched in a tensile machine and the local deformation was determined from the distances between the gauge marks. There were wide differences in the amount of deformation from grain to grain ranging from 3 to 14 per cent for an overall elongation of the specimen of 5 per cent. Moreover the deformation was not uniform within one crystal, but was influenced by that of the neighbour. Hard crystals, i. e. those with a small average extension, inhibited the deformation in their neighbourhood and were themselves more heavily deformed near their boundaries than if they had been free. In large crystals these effects were observed up to 1.5 cm from the boundaries. The same conclusions were drawn from hardness measurements after deformation which showed the corresponding variations due to greater work hardening of the more heavily deformed regions, and from observations of density of slip lines and of asterism of Laue spots.

The result of the interaction between the crystals is that the extension is continuous across the boundary. This was to be expected[3]) since it is a consequence of the well established fact that the volume is not changed by plastic deformation.

These conclusions were made from experiments carried out on the surface of the specimens. The objection has been raised that the surface behaviour is not representative of that in the interior of a specimen and that a crystal entirely surrounded by other crystals may behave differently. This objection has been met by OGILVIE[4]) using a recently developed etching technique[5]) by which the traces of slip planes are developed in the interior after the surface of the specimen has been removed electrolytically. He showed, in alpha brass with a grain size of approximately 2 mm, that different sets of slip planes were operating in various parts of the same crystal and that the extent of their operation also varied. Still more striking was the observation of twin bands which are a feature of annealed brass. The originally straight bands, including even those at a small angle to the direction of stress, were bent and compressed after the tensile deformation, indicating a complex stress system in a crystal in the interior of the specimen.

These investigations prove that the deformation of a crystal in an aggregate is inhomogeneous due to the anisotropy of the plastic properties of the crystals, and that this is the case for crystals in the interior, as well as those on the surface. This source of inhomogeneity is inherent in all crystal aggregates. The investigation of its details is however difficult experimentally, as the grain boundaries become visible only on etching, and the orientation which governs the plastic properties cannot be determined easily for small crystals.

These difficulties are overcome if an aggregate is used consisting of two types of crystals of widely different properties. Here the differences due to anisotropy are small compared with those due to the crystal structures and hard crystals are distinguishable from soft crystals after etching. The effect of the interaction thus becomes both greater and more easily observable.

3. Presence of Crystals of Different Structures.

In alloys containing more than one phase the crystals of the phases have different chemical compositions, crystal structures, yield points and work hardening characteristics. They do not start deforming simultaneously, and the extent to which they take part in the deformation depends on the relative volumes present, their distribution in the alloy and the slopes of their tensile curves which characterize work hardening. The deformation of two-phase alloys has been investigated by HONEYCOMBE and BOAS6) using copper-zinc alloys of approximately 40 atomic per cent zinc, and by CLAREBROUGH7) using silver-magnesium alloys of approximately 30 atomic per cent magnesium. In both these systems the alloys consisted of face-centred cubic alpha crystals and body-centred cubic beta crystals, which are much harder and contain a higher concentration of zinc or magnesium respectively than the alpha crystals. Both phases are ductile. Details of the deformation process were obtained directly by observation of the appearance of slip lines on the surface of polished specimes and from conclusions drawn from the recrystallization behaviour of the two phases on annealing. This was investigated microscopically and by means of *X*-rays. It should be emphasized that in general the relative deformations of crystals of different phases cannot be obtained from recrystallization data as crystals of different composition and structure behave differently on recrystallization. A further complication occurs in the alloys investigated owing to order-disorder transformations. It is found that after the same amount of deformation, a phase which is in a state of order before deformation starts recrystallizing at a lower temperature than when it is disordered before deformation. However if crystals of one phase only are considered or if parts of a crystal are compared, the assumption is justified that recrystallization starts in the most severely deformed crystals or regions of a crystal.

In both alloys the recrystallization of the beta phase starts at the boundary with alpha crystals. This shows again that the deformation of the hard phase is enhanced in the neighbourhood of the soft phase and is in agreement with the observation that slip lines always become visible first in the alpha crystals and sometimes continue for a short distance into the beta crystals.

The detailed investigation of the silver-magnesium alloys[7]) shows that the mode of recrystallization of the alpha phase is more complicated and varies with the proportions of the two phases. If there are only a few beta crystals dispersed in the alpha matrix the deformation of the alloy occurs almost entirely in the matrix. Moreover there is additional flow of the matrix around the hard particles which remain almost undeformed and as a result the recrystallization of the alpha phase starts near the alpha-beta boundaries.

If the beta phase is present in larger proportion and no deformation took place in the beta crystals, the alpha crystals between them would have to deform to a very large extent. Due to work hardening the stresses would be so high that fracture of the beta would occur. This has indeed been observed with non-metallic inclusions in steels. However in the alloys under consideration the beta crystals flow under the high stresses, and as the deformation is continuous across the boundary the beta crystals exert a strengthening effect on the alpha crystals. Therefore recrystallization begins near their centres, i. e. away from the alpha-beta boundaries. The deformation (per unit volume) of the alpha crystals is still greater than that of the beta crystals which however take part to an increasing extent in the deformation.

If both phases are present in nearly equal volume proportions the alpha phase is no longer the continuous matrix and alpha crystals become surrounded by beta crystals. Both phases then share the deformation to the same extent; in duplex brass the preferred orientations of the crystals of the two phases develop simultaneously and to the same degree[8]). The alpha crystals start recrystallizing at random, there is no preferred situation for the nuclei. These results indicate that here the homogeneity of deformation is considerably greater than expected. This mode of deformation and recrystallization continues to high proportions of the beta phase where the alpha crystals have the role of soft inclusions deforming together with the matrix.

The interaction between the crystals of different hardness influences their relative deformations and thus their recrystallization behaviour but it does not affect the crystallographic mechanism of slip. This is shown by the observation that after heavy deformation, the preferred orientations of each phase in duplex brass, are the same as those of the two phases deformed separately[8]).

Effects similar to those of varying proportions of the phases can be expected also in single phase alloys if crystals of orientations which give rise to high elastic limits are distributed in increasing numbers. The case of the alloy in which the two phases are present in equal proportion, corresponds to that of random orientation of the crystals. In such an aggregate consisting of crystals of similar size but different elastic limits each crystal has to follow the general deformation, the interaction at the boundaries remains highly localized and there is not much additional deformation and work hardening near the boundaries. This view is supported by the observation⁹) that the polycrystalline tensile curve of cubic metals lies between the tensile curves of single crystals of different orientations. It also explains the good agreement between the experimental tensile curve and that calculated by TAYLOR. It thus seems that the effects of the inhomogeneity of deformation investigated in aggregates of large crystal size and discussed in section 2 of this paper are not so marked if aggregates of smaller grain size are considered. In order to take the grain size effect into account, it is thought that Taylor's theory may be altered, without changing its structure, by the addition of a term incorporating the inhomogeneity of deformation. These considerations hold, of course, only for cubic metals. In hexagonal metals possessing only one slip plane, the type of deformation is entirely different and the inhomogeneity has more serious effects as exemplified by the fact that the work hardening curve of the polycrystalline material is much steeper than the steepest of curves observed on single crystals⁹).

4. The Deformation Process.

The inhomogeneities discussed had their origin in the interaction between different crystals and are thus of a microscopic nature. However another type of inhomogeneity is produced by common methods of deformation. This occurs on a macroscopic scale affecting a large number of crystals and is a consequence of friction between forming tool and metal. The frictional forces exerted by the tools during such processes as drawing, extrusion and rolling, depend on their shape, i. e. die angle, diameter of the rolls etc. It is well known that the outer layers of the metal are deformed differently from the centre so that internal stresses are set up and the hardness and preferred orientation vary from surface to core¹⁰). These changes occur in single-phase metals and alloys as well as in two-phase alloys⁸). The usual methods of investigation have been to observe the flow of various layers in a composite specimen of two similar materials, or the change in shape of a square network ruled on one of the inner surfaces of a split specimen. COLLINS and HONEYCOMBE[11]) have investigated the inhomogeneity produced during wire drawing and rolling of brass of small grain size by observing the recrystallization behaviour. As mentioned above, the most heavily deformed regions of a single phase alloy specimen can be assumed to recrystallize first; the method consists of annealing the deformed material in such a way that these parts have just started to recrystallize. This method has an advantage over the two other methods mentioned as the observations are made on the actual material and not on composite models.

In wire drawing the deformation of the outer layers depends on the friction between die and metal and therefore on the condition of the die and its lubrication. Using worn steel dies and a heavy grease as lubricant, the additional deformation at the surface is very marked particularly after small reductions in area of the wire. In wires drawn approximately 10 per cent, recrystallization occurs at the surface but not at all at or near its centre. With increasing reduction nucleation occurs also in intermediate layers still leaving the centre of the wire the last part to recrystallize. Reductions of 70—90 per cent are needed to reduce the surface effect; recrystallization then starts half-way between surface and centre. This shows that localized deformation due to frictional forces becomes less significant the greater the total deformation.

Using new tungsten carbide dies and stearic acid as lubricant the surface effect is absent and even after small deformations recrystallization never starts at the wire surface. However at all deformations recrystallization is always more advanced in the outer zones and never starts at the centre.

The differences in behaviour of the various layers in rolled sheet are similar to those observed in wire drawn through worn steel dies. After small reductions in thickness recrystallization starts in a thin surface layer leaving the rest of the specimen almost unrecrystallized. This effect becomes less pronounced as the deformation is increased until at high deformations no zone of preferential recrystallization can be found and nuclei appear at random throughout the specimen.

These conclusions are not fundamentally new. However they are obtained with a technique less open to criticism than some of those previously used and agree with the conclusions drawn from measurements of the variation of hardness across the specimen. They also show clearly the role of friction between the tool and metal.

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