measurements (i.e. aged for 15 h at 300° C after being homogenised) was tested in tension at room temperature at an initial elongation rate of 2×10^{-3}/sec. The crss amounted to 1310 g/mm². In the Orowan model [10] it is assumed that dislocations bow out between particles and the crss is given to a first approximation by $\tau_0 = \tau_{om} + 2G\mathbf{b}/l$, where τ_{om} is the crss of the matrix [3], G is taken as the shear modulus of pure magnesium, **b** the Burgers' vector and l the mean precipitate spacing in the basal plane (estimated at 2.5×10^{-3} mm).

Though the (τ, γ) curve (fig. 5) conforms to alloys containing non-deforming particles for which equation 1 is valid, the above formula predicts a crss of 519 g/mm², which is much too low when compared with the experimental 1310 g/mm². At the moment it is not possible to decide whether this difference is caused by an increase in the shear modulus due to the presence of the precipitate or that the by-passing process [11], following the bowing of dislocations, is extremely difficult as can be expected for cph alloys. The rate of strain-hardening in the linear part of the (τ, γ) curve is of the same order of magnitude as with pure magnesium i.e. 1100 g/mm².

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Interlamellar Slip in Polyethylene

The purpose of this note is to comment upon the interpretation of the small-angle X-ray patterns in parts 2 and 3 of a study of orientation effects in polyethylene [1, 2]. In these papers, the authors propose that the crystalline regions of an oriented, branched polyethylene consist of stacks of lamellae. Changes in the wide-angle and small-angle X-ray patterns produced by annealing, compressing or extending an oriented specimen within certain limits are interpreted in terms of interlamellar slip. We shall define a precise model for a stack of lamellae and consider the effects of the rotation of the stack and of inter-lamellar slip within it on its diffraction pattern.

Fig. 1a shows the model we adopt. Each lamella is a single crystal and is surrounded by a layer of amorphous material of different electron density. Fig. 1b shows the Fourier transforms of the linear lattice formed by taking an identical point within each lamella, and of the lamella itself. The lattice transforms into a set of planes, intersecting the plane of the diffraction pattern in layer lines. These are spread out to a width of L^{-1} because of the finite length of the lattice, L. The lamella transforms into a spike normal to the plane of the lamella. The diffracted amplitude is given by the product of these transforms and therefore the diffraction pattern has the appearance shown by the shaded regions. The addition of the diffraction from the twin of the stack of lamellae shown in fig. 1a produces the four-point pattern shown in fig. 2a. This corresponds to the pattern obtained by annealing at 70° C after drawing and rolling in the $0y$ direction [1]. The material oriented in this way is therefore conceived as consisting of stacks of lamellae of the type shown in fig. 1a, and its twin as distributed in a matrix of amorphous material. The nature and extent of the amorphous material is not specified, but it must bound the stack by material of different electron density from that of a lamella.

This model is similar to the model adopted in [1] and [2], except that in [2] it is assumed that the lateral extent of the lamella (width W in fig. 1a) is large enough not to cause a spread in the diffraction pattern. The spread of the small-angle spots is, on this view, the result of variations in the orientations of the lamellae. To accord with

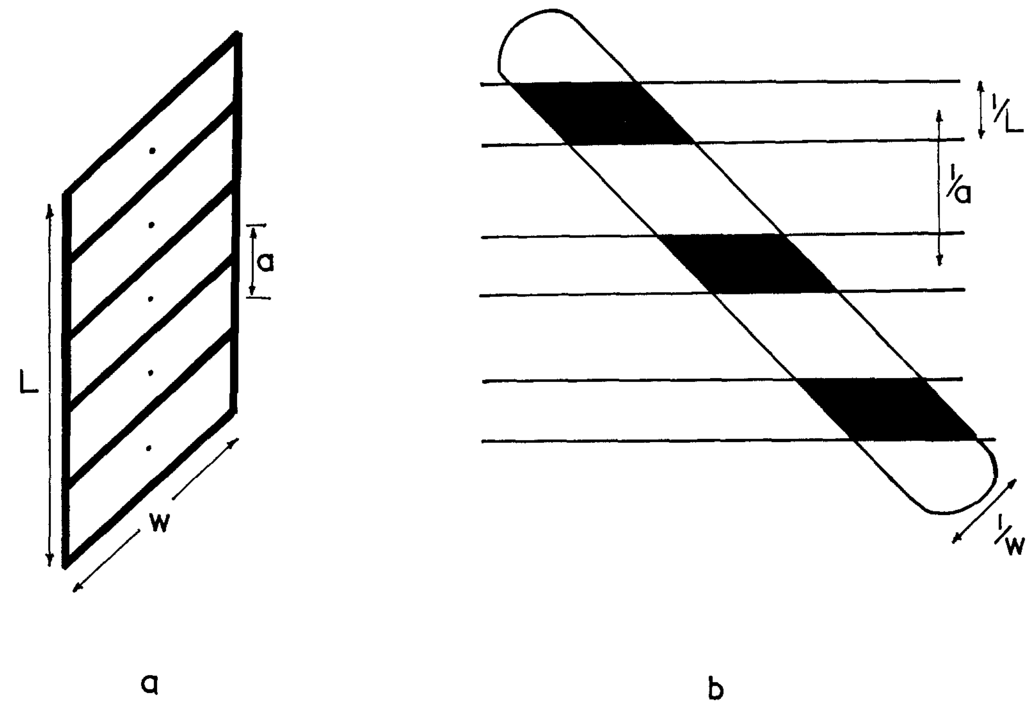

Figure 1

this view, the spike of width W^{-1} in fig. 1b must be replaced by a cone whose angle is the range of orientations of the lamellae in different stacks of lamellae. This will slightly change the shape of the diffraction spot, but will not materially change the diffraction pattern, nor will the conclusions of the following section be altered significantly.

The effect of rotating a stack of lamellae through an angle ϕ is to rotate its diffraction pattern (at high and low angles) through the same angle. If twin stacks rotate in opposite senses, the four-point pattern changes in the manner shown in fig. 2b. The reverse rotations are shown in fig. 2c. Interlamellar slip in the sense corresponding to compression in the $0y$ direction produces the effect shown in fig. 2d. Here, the linear lattice is rotated, and therefore the layer lines rotate, but the lamella and its diffraction spike remain in the same orientation. The effect of interlamellar slip in the sense corresponding to extension along $0y$ is shown in fig. 2e. Clearly the effect of slip is distinct from that of rotation. When slip is accompanied by rotation the effects of slip and rotation are combined. If the rotation of the high-angle diffraction pattern is observed to be the same as that of the low-angle pattern, evidence for slip is lacking. In [1], equal rotations at high and low angles occur during an anneal at 70 to 95° C, which is accompanied by a contraction of the specimen along $0y$, the change in small-angle pattern being similar to that shown in fig. 2b. The simplest explanation of this effect is that the stacks of lamellae rotate but that they do not deform by slip. This is contrary to the explanation given in [1], that the lamellae rotate as a consequence of interlamellar slip. In terms of our model, the rotation can be explained by supposing that the stacks of lamellae are forced to rotate by the deformation of the amorphous material surrounding them, without themselves undergoing internal slip. However, a simple rotation of the diffraction pattern might be reconciled with the occurrence of interlamellar slip on the basis of a different model. There is one type of model, in particular, for which slip will not affect the diffraction pattern. If the slip of one lamella over another draws in material, originally part of the amorphous region surrounding the stack, which can crystallise and thereby maintain the boundary of the lamella, and if the reverse transformation occurs at the opposite edge of the lamella, then the boundaries of the stack, and therefore its representative lattice, will remain unchanged by slip. If this behaviour obtains, the diffraction pattern can provide no

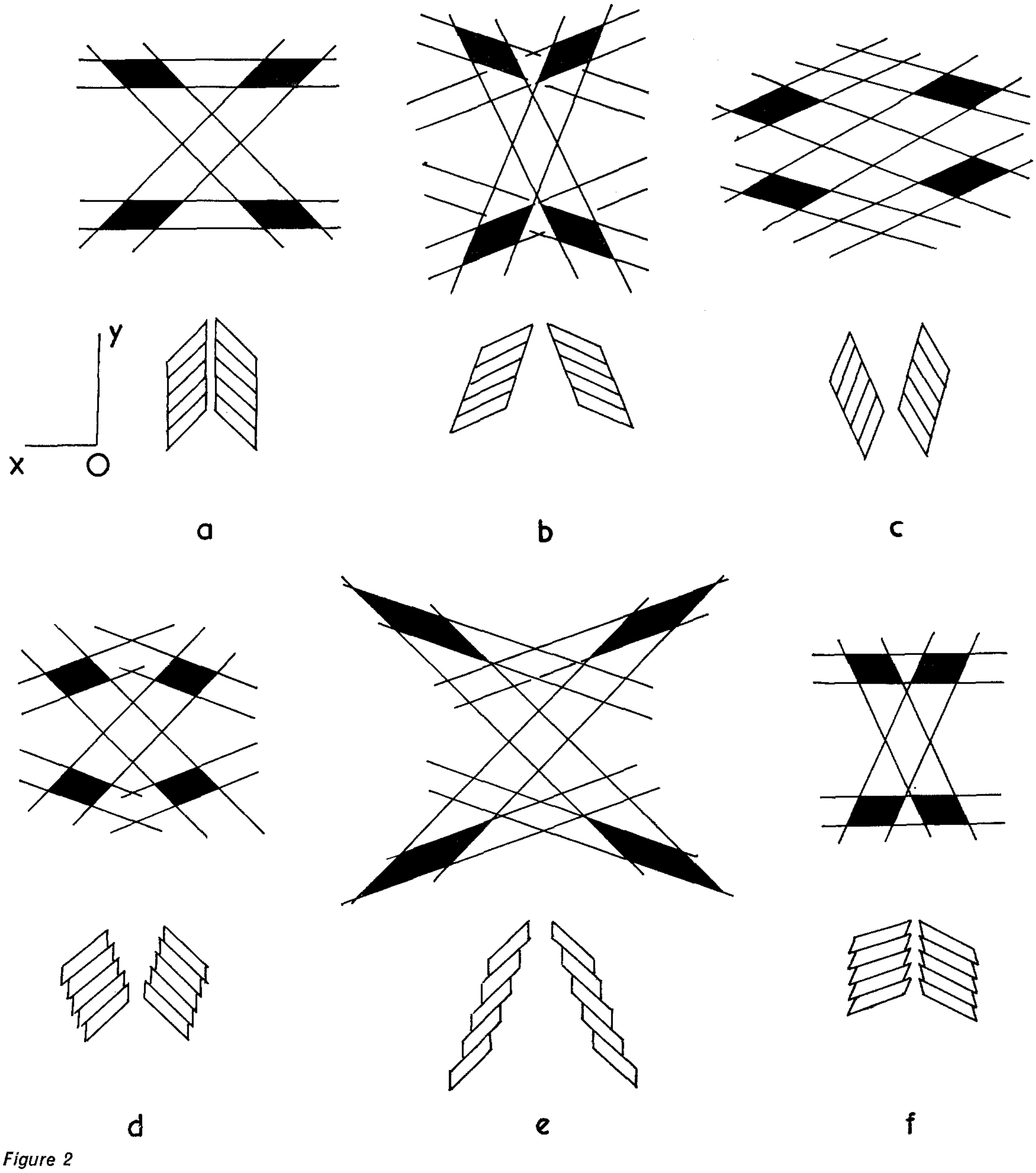

Figure 2

evidence for or against interlamellar slip.

If it is supposed that a stack of lamellae does in fact deform by interlamellar slip then it can be expected that the slip will be accompanied by a rotation, which is enforced by the constraint of the surrounding material. In order to predict the amount of rotation, the nature of the constraint must be known. The rotation will not necessarily be the same as that accompanying the uniaxial deformation of a metal single crystal by slip. Furthermore this rotation itself depends upon the mode of testing. In a tensile test the constraint is that the grips must remain on the line which initially joins them. This leads to the equation [3]

$$\frac{l_2}{l_1} = \frac{\cos \phi_1}{\cos \phi_2} \tag{1}$$

where l_1 and l_2 are the lengths of the crystal before and after extension and ϕ_1 and ϕ_2 the corresponding angles between the tensile axis and the slip plane normal. In a compression test the rotation is not merely the reverse of that in a tensile test, because the constraint, which is that the planes parallel to the compression plate keep their initial orientation, is different. This constraint leads to the equation [3]

$$\frac{l_2}{l_1} = \frac{\sin \phi_2}{\sin \phi_1} \quad (2)$$

where l_2 and l_1 are the lengths before and after compression, respectively. Until the nature of the interaction between the individual stack of lamellae and the material which surrounds it is understood, it is impossible to specify the amount of rotation which must accompany the extension or contraction of the stack of lamellae by interlamellar slip. In [1] and [2] it is assumed that the constraints are such that equation 1 holds. The effect of slip under this constraint upon the diffraction pattern from our model is shown in fig. 2f. Of the diffraction patterns published in [2], only one appears, by rough measurement, to show a change of this type.

Interlamellar slip is an attractive mechanism for the deformation of crystalline polymers, and it is reasonable to postulate its occurrence. However, when the diffraction patterns observed in [1] and [2] are interpreted in terms of a model of discrete stacks of lamellae, they give little support to the postulate that the stacks deform by interlamellar slip. An alternative interpretation of the diffraction patterns might well give better support for interlamellar slip in polyethylene under the conditions described in [1] and [2]. The present uncertainty emphasises the need for further knowledge of the structure of the material, so that the diffraction patterns can be interpreted unambiguously and the relationship between interlamellar slip and lamellar rotation can be predicted.

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Book Reviews

Advances in Materials Research, Volume 3

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Pp 417 (John Wiley, 1969) 182*s*

With this third volume, the character of the series "Advances in Materials Research" appears to have changed considerably. Hitherto, the accent has been on experimental methods. Volume I (1967) was specifically entitled "Experimental Methods of Materials Research", and dealt with diffraction and microscopy, Mössbauer effect, mechanical methods, calorimetry and diffusion. Volume 2 (1968) was again largely experimental, and the central theme was microplasticity. The new volume is predominantly theoretical, and has no longer any pretension to homogeneity.

If we are to judge by its first article, it would appear that this book is aimed at a readership completely different from the one for which the first two volumes were written. The article is "The Continuum Theory of Dislocations" by T. Mura of Northwestern University, and is an account of the work of the author and others in developing rigorous mathematical methods for solving plastic strain problems. Many of us who feel at home with Cottrell or Lothe and Hirth must, I feel sure, throw in the sponge at this point. Excellent though the article may be, one feels that in this series it is sadly misplaced; it fails to fulfil the editor's purpose as he presents it in his Introduction.

The second article, "Fatigue Hardening in Face-Centred Cubic Metals", by R. L. Segall of the University of Warwick is, in complete contrast, ideally suited to the series, and is a readable survey of how direct observation of dislocations has aided our understanding of