

external stress was involved during or after the transformation, it was concluded that surface relief represents slip lines caused by the constraints of the surroundings. This indicates the existence of another lattice-invariant shear in the outer parts of partially-twinned Fe-26.4%Ni-0.24%C alloy martensites in addition to the well-known transformation twins localized around the midrib.

Present results also support the two-stage process of the partially-twinned martensite formation in Fe alloys which was first suggested by Patterson and Wayman [10].

References

1. N. D. H. ROSS and A. G. CROCKER, *Acta Met.* 18 (1970) 405.
2. A. F. ACTON and M. BEVIS, *Mater. Sci. Eng.* 5 (1969/70) 19.
3. G. THOMAS, *Met. Trans.* 2 (1971) 2373.

4. R. P. REED, *Acta Met.* 15 (1967) 1287.
5. T. N. DURLU, D. Phil. Thesis, Oxford University (1974)
6. E. O. FEARON, D. Phil. Thesis, Liverpool University (1972)
7. M. A. WECHSLER, D. S. LIEBERMAN and T. A. READ, *Trans. AIME* 197 (1953) 1503.
8. J. S. BOWLES and J. K. MACKENZIE, *Acta Met.* 2 (1954) 129.
9. D. P. DUNNE and C. M. WAYMAN, *Met. Trans.* 2 (1971) 2327.
10. P. L. PATTERSON and C. M. WAYMAN, *Acta Met.* 14 (1966) 347.

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Lamellar reorientations in low density polyethylene

One of the main deformation mechanisms in oriented polymers at room temperature is chain slip. This is true, for example, in polypropylene and high density polyethylene [1, 2] and in low density polyethylene [3]. Also, in polypropylene and high density polyethylene deformed in tension, the reorientation of the lamella surface has been found to be close to that expected when shearing of the material occurs as the result of chain slip [1].

In doubly-oriented low density polyethylene (LDPE) deformed in tension along the original draw axis chain slip and lamella separation occurred [3], the latter being attributed to straining of the interlamellar layers. This note provides further data on such an experiment and shows that the changes in the small-angle X-ray pattern are different from what may be expected if the lamellae are sheared by chain slip. It is shown that stretching of the interlamellar layers can account for the reorientation of the lamella surface.

Doubly-textured LDPE was prepared by drawing and rolling at room temperature and then annealing at 95.0° C. Test pieces were cut parallel to the draw (y) axis and strained in tension at 20 ± 3° C. X-rays showed that θ (angle between

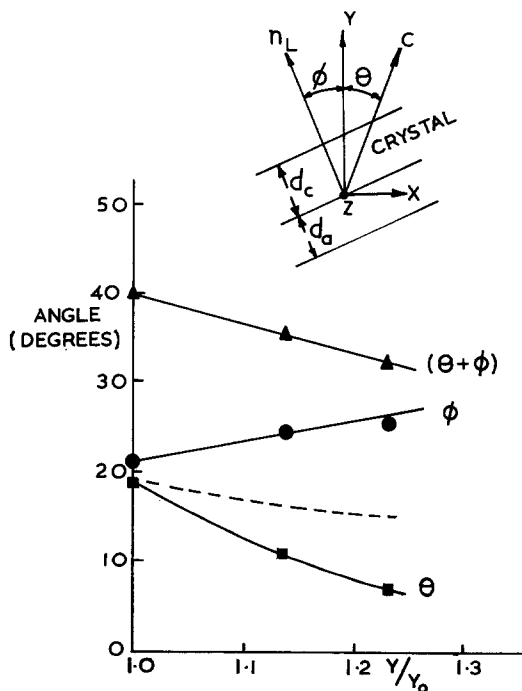


Figure 1 Angles θ , ϕ , and $\theta + \phi$ plotted against y/y_0 for doubly oriented LDPE at 20° C. θ and ϕ are defined in the schematic diagram in which n_L is the lamella normal, y is the tensile axis, and c is the chain axis. x is perpendicular to the rolling plane. The broken curve shows the variation of θ predicted by the equation $y/y_0 = \sin \theta_0 / \sin \theta$ (see text).

TABLE I Variation of the long spacing (d) values with y/y_0 . d_y/d_{y_0} is the ratio of the long spacings measured along y

y/y_0	$d(\text{\AA})$	d_y/d_{y_0}
1.0	152	1.0
1.14	167	1.13
1.22	177	1.19

chain and tensile axes) and ϕ (angle between lamella normal and tensile axis) varied with extension as shown in Fig. 1. The angle ($\theta + \phi$) changed during deformation and is evidence of chain slip within the lamellae. The molecular axes remained in the xy -plane of the sample, indicating that there was plane strain on the molecular scale within the crystals. Also, there was little deviation from plane strain in the sample as a whole, since the strain along the z -axis was small in comparison with the strains along the y - or x -axes, as shown in Fig. 2 (the x -, y -, z -axes are defined in Fig. 1). Long spacing (d) values as a function of extension are given in Table I.

If the material deformed by chain slip alone, the long spacing would be changed by the factor $\cos(\theta + \phi)/\cos(\theta_0 + \phi_0)$ where θ_0 and ϕ_0 are the initial values of θ and ϕ . This predicts that the value of d should increase from 152 Å prior to deformation to 167 Å at $y/y_0 = 1.22$. In fact the observed value of d at $y/y_0 = 1.22$, was 177 Å. Therefore, the chain slip which occurs cannot account for all of the increase in long spacing and

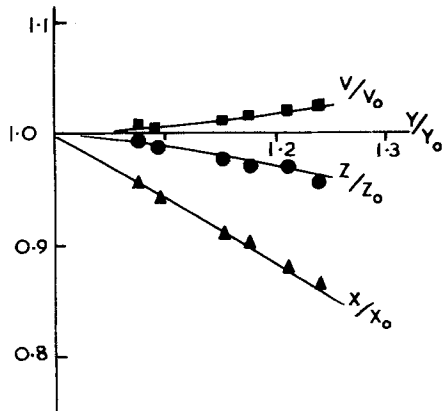


Figure 2 x/x_0 , z/z_0 and v/v_0 plotted against y/y_0 . The x , y and z directions are defined in Fig. 1. v_0 and v are the volume of the sample before and after deformation.

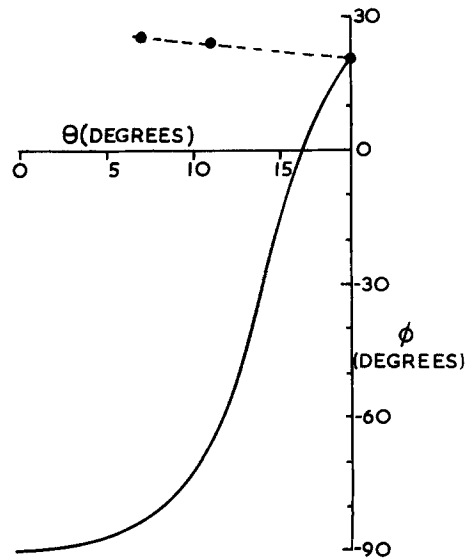


Figure 3 Predicted orientation (ϕ) of lamella normals (full curve) according to Equations 1. The observed ϕ values are shown by the broken line.

so does not occur homogeneously through the material. In support of this is the fact that the changes in molecular orientation (θ) with extension are more rapid (see Fig. 1) than is predicted for uniform chain slip, which should follow the relation: $y/y_0 = \sin \theta_0 / \sin \theta$ in a plane strain, which is approximately the case here.

It may be expected that the chain slip which occurs would change ϕ in the opposite rotational sense to that which is observed: the equations connecting θ and ϕ with the shear γ_c for chain slip in tension are

$$\gamma_c = \cot \theta - \cot \theta_0$$

$$\text{and } \gamma_c = \tan(\theta_0 + \phi_0) - \tan(\theta + \phi) \tag{1}$$

The orientation of the lamellae as predicted by Equations 1 is given in Fig. 3, showing that n_L would rotate through the tensile axis, in contrast to the observed change in ϕ which is away from the tensile axis. Therefore, it appears that because continuity of the oriented molecules through the sample does not exist, the constraints on the molecules of a tensile test do not apply.

We propose that the changes in ϕ result from deformation of the interlamellar layers. This mechanism has apparently not been previously

used to explain results on lamellar reorientations. Let small strains ϵ_{ax} and ϵ_{ay} occur in the interlamellar region along the x - and y -directions. As a result of this deformation a plane surface in the interlamellar region with normal parallel to n_L would be rotated towards the tensile axis by an amount $\Delta\phi$, given by {4}

$$\Delta\phi = (\epsilon_{ay} - \epsilon_{ax}) \sin\phi \cos\phi \quad (2)$$

The lamellar surfaces (or, as could be the case, the surfaces of crystal blocks which form the lamella [5]) may follow this deformation and thus n_L would rotate by amount $\Delta\phi$ away from the tensile axis

The stretching of the interlamellar regions may be found by writing the extension ratio of the lamella structure as:

$$\begin{aligned} \frac{d_y}{d_{y_0}} &= \frac{y_a}{y_{a_0} + y_{c_0}} + \frac{y_c}{y_{a_0} + y_{c_0}} \\ &= \frac{y_a/y_{a_0}}{1 + y_{c_0}/y_{a_0}} + \frac{y_c/y_{c_0}}{1 + y_{a_0}/y_{c_0}} \end{aligned} \quad (3)$$

where y_a and y_c are the thicknesses of the amorphous and crystalline regions measured along the tensile axis; y_{a_0} and y_{c_0} are the corresponding thicknesses before deformation. Putting the crystallinity $\chi = y_{c_0}/y_0$ into Equation 3 and rearranging give

$$\frac{y_a}{y_{a_0}} = \frac{\frac{d_y}{d_{y_0}} - \chi \frac{y_c}{y_{c_0}}}{1 - \chi} \quad (4)$$

Values of y_a/y_{a_0} were calculated from Equation 4: y_c/y_{c_0} was obtained from the geometry of the lamella, and the value of χ was taken to be 0.55, as the crystallinity of LDPE is normally in the range 0.5 to 0.6

The dimensional changes of the interlamellar region along the χ -direction were taken to be the same as for the lamellae stack: given by

$$\chi_a/\chi_{a_0} = \frac{\cos\theta}{\cos\theta_0} \quad (5)$$

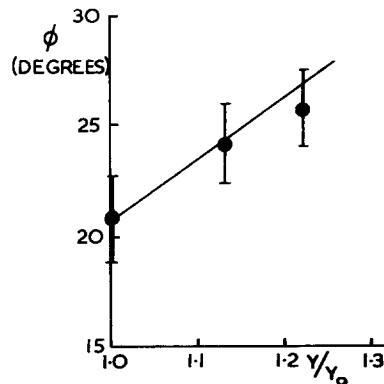


Figure 4 Angle ϕ plotted against y/y_0 . The full line shows the values predicted by Equation 2.

Using Equations 4 and 5, the quantities $(\epsilon_{ay} - \epsilon_{ax})$ were found for increments in y/y_0 of value 0.01 up to $y/y_0 = 1.22$. Values of $\Delta\phi$ were calculated using Equation 2, the new total ϕ being obtained after each increment. Values of ϕ calculated in this way are compared with the experimental values of ϕ in Fig. 4 and there is agreement within the experimental error of $\pm 2^\circ$. This conclusion would also hold if crystallinities of 0.5 or 0.6 had been assumed.

References

1. D. M. SHINOZAKI and G. W. GROVES, *J. Mater. Sci.* 8 (1973) 1012.
2. R. J. YOUNG, P. B. BOWDEN, J. M. RITCHIE and J. G. RIDER, *ibid.* 8 (1973) 23.
3. A. KELLER and D. P. POPE, *ibid.* 6 (1971) 453.
4. S. P. TIMOSHENKO and J. M. GERE, "Mechanics of Materials" (Van Nostrand Reinhold Co., New York 1972).
5. A. PETERLIN, *J. Mater. Sci.* 6 (1971) 490.

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