problem presented by the fully twisted ribbons in Fig. 1, where the distance of closest approach between two lamellae twisting in phase is always determined by the ribbon width.

Such a half-twist has been observed by Geil [1] in some branched polyethylene spherulites, and by Keller and Machin [8] in crystallization of some stretched polyethylene melts. The results from the above ion etched linear polyethylene spherulites are consistent with other SEM observations made in this laboratory of unetched deeply ion-etched, and HNO_3-etched samples and on spherulites deformed to 50 to 100% elongations, which will be reported in full at a later date.

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Comments on "Constant stress creep and constant true strain-rate tensile tests of the superplastic alloy PbSn"

Recently, Baudelet and Suery [1] have reported results on true stress – true strain curves for lead–tin eutectic alloy (Pb–61.9 wt %Sn) in a superplastic condition which they believe conflict with earlier results which we have reported [2]. We reported strain hardening which was approximately a stress increase of 20% for an engineering strain of 100% (ϵ = 69%) in as-rolled lead–tin eutectic [2]. Baudelet and Suery [1] tested lead–tin eutectic which had been prepared by extrusion and then annealed "about 15 min at 130°C". They found strain independent flow, i.e., no strain hardening or strain softening. On the basis of this, they criticized our result, indicating that a possible source of error in our experiment was uncertainty in the true gauge length due to deformation in the grips [1]. In our earlier paper, we emphasized the importance of determining the actual length of the specimen which underwent deformation in creating the condition of constant applied strain-rate [2]. Our specimen was designed to provide minimal "feed-in" of material from the fillet to the gauge of the specimen. We have measured the volume of material fed into the gauge length by measuring the lateral and longitudinal profiles of the specimen near the gauge and found that this material effectively increases the gauge length by less than 2%. Baudelet and Suery [1] have calculated that a grip contribution of 20% to the initial gauge length is necessary to produce a false strain hardening effect in their material.

The explanation for this discrepancy in the results arises from differences in the conditions of the starting material used in our work [2] and that done by Baudelet and Suery [1]. Fig. 1 shows the true stress – true strain curves for lead–tin eutectic in various starting conditions.

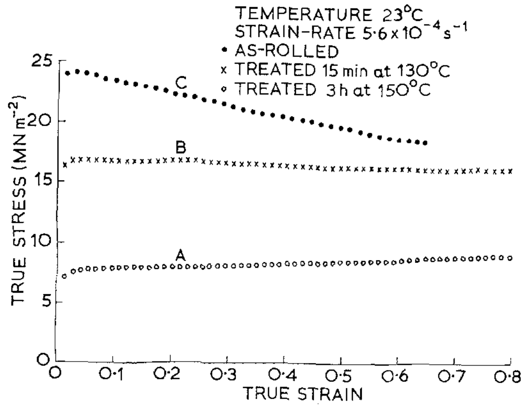

Figure 1

Curve "A" for as-rolled material tested immediately after preparation using a constant imposed strain-rate does indeed show strain hardening. Curve "B" for material produced by rolling and then annealing for 15 min (exactly) at 130°C, the condition most similar to the material used by Baudelet and Suery [1], and tested in our constant imposed strain rate device [2], shows flow which is very nearly strain independent. Curve "C" for material produced by rolling and then annealing for 3 h at 150°C shows pronounced strain softening. The condition of the material thus greatly affects both the character of the true stress – true strain behaviour and the level of the flow stress which is observed. We conclude that the difference between our results [2] and those of Baudelet and Suery [1] are explained entirely by the slight difference between the microstructure of the starting materials, and not by the form of the grips used in the two experiments. Subtle changes in the microstructure can significantly affect the mechanical properties observed, and therefore the characterization of the material is important.

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Growth of n-alkane crystals on graphite and on carbon-fibre surfaces

The growth of polymers, for which *n*-alkanes are models, on solid surfaces is a useful method of examining structure and interactions at solid/solid interfaces of relevance to adhesive systems and composite materials.

Fig. 1a shows a photograph, taken using incident polarized light, of crystals of *n*-hexatriacontane ($C_{36}H_{74}$) grown from amyl acetate solution on the basal plane of freshly cleaved Ticonderoga graphite. Three important features are evident: (i) the lozenge-shaped crystals are growing on an edge face; (ii) crystals grow independently without contact with one another; and (iii) crystals are oriented in three directions; the directions adopted by the largest visible dimension of the crystals are the $\langle 10\bar{1}0\rangle$ directions of hexagonal graphite as shown in Fig. 1b, and are the directions of the twin-boundaries used as crystallographic indicators [1, 2]. It is also noted that cleavage steps on graphite surfaces run in $\langle 10\bar{1}0\rangle$ and $\langle 11\bar{2}0\rangle$ directions [3]. Crystals grown from the melt showed similar orientation, but the mass in which they grew obscured details. The structures of several alkanes have been described by Keller [4], who found that the structure of $C_{36}H_{74}$ crystals depends to some extent on their method of preparation. In particular, those grown from petroleum ether are monoclinic when formed in suspension, and are orthorhombic when formed on evaporation of the solvent. Growth of the crystals, of which Fig. 1a is typical, was difficult to obtain by cooling the amyl acetate solution, and they were usually grown by evaporation. Also, it was found that the crystals much preferred to grow on edges of graphite flakes rather than on the basal plane. The question arises as to whether the parallel stacking of the hydrocarbon chains is perpendicular to (in which case the habit may be orthorhombic), or oblique to (when the habit is monoclinic or triclinic), the major faces of the crystals. Characterization of the crystals grown on graphite is obviously required, but on the supposition that the crystals are orthorhombic with parallel chains as in