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THE first and major point raised by Bagley, Rudin and Schreiber is that when they compare my results[1] for the onset of flow instability in cone-and-plate viscometers with those for the onset of flow instability in capillary extruders they observe discrepancies. For example, the critical shear stress in the first type of instrument is independent of molecular weight and dependent on geometry, whereas in the second type it is dependent on molecular weight and independent of geometry. They have chosen the conclusion that what I have observed and measured is some other phenomenon than the one observed in capillaries. While this may be a correct conclusion it certainly does not follow from their argument about effects of 'geometry', and I think it is an incorrect conclusion for the following reasons.

An increase in the radius of a capillary tube is, in a sense, a change of geometry, but it is not the same sort of change as that between a capillary and a cone-and-plate viscometer: and, because it has been observed that the shear stress at the onset of melt fracture is independent of tube radius, it does not follow that the stress is independent of other sorts of geometrical changes. In the theory I give for fracture I take into consideration the total elastic energy in the sheared liquid. This is the integration of elastic energy through all variations in shear field and throughout the volume of liquid. For the cone-and-plate viscometer, in which the rate of shear is everywhere constant, this integration is very simple. Since the final expression contains certain dimensions of the cone–plate system, geometrical factors are important, and I stated this clearly in the final sentence of the article. A similar calculation for a capillary tube extruder cannot possibly contain the same geometrical factors, and a facile transference of critical shear stress data from one geometry to another is therefore not permissible. On the other hand, a theory which, when applied to geometrically completely different shearing devices, gives results in agreement with experiment will probably be correct. The application of the total elastic energy concept to the case of a capillary tube does in fact lead to the prediction that the shear stress at the onset of flow instability for silicone fluids and other 'power law' liquids is independent of tube radius. This and similar calculations are to be published.

The second major point concerns the approach to fracture through the elastic energy concept. I agree that the concept is a direct descendant of Griffith's theory[2], but it is considerably modified to avoid a crack length in the final expression and to find application to liquids. Griffith's own discussion of liquid fracture is more to do with cavitation than fracture, and the method of calculating the elastic energy from the components of stress was not available to him. A puzzling feature of the analysis given by Bagley, Rudin and Schreiber is the failure to introduce a fracture criterion. There seems to be no reason why their final expression, equation (4), should not apply at any pair of values of shear stress and rate of shear and not just at the values at melt fracture. In fact the whole analysis appears to comprise a set of dimensionally consistent identities, and agreement with experiment, as they state, is to be expected.

A further point should not go unanswered. Bagley, Rudin and Schreiber, using a dubious argument about entanglements, conclude that some of the silicone fluids I studied 'would not be expected' to have non-Newtonian properties and 'should not' have elastic properties. However, these theoretical predictions are not borne out by my own measurements on silicones in the 10^2–2×10^3 poise range or by the measurements of J. J. Benbow and E. R. Howells[3] in the 10^2–3×10^5 poise range, of A. J. Barlow and J. Lamb[4] in the 8–$2 \cdot 6 \times 10^2$ poise range, of A. J. Barlow and G. Harrison[5] in the 1–10^3 poise range, and of C. C. Currie and B. F. Smith[6] in the 30–10^3 poise range.

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'Hook Effect' in the Analysis of X-ray Diffraction Line Profiles

IN the analysis of X-ray diffraction line profiles by the Warren and Averbach method[1] a hook has often been observed in the particle size coefficients (A_L^P) versus distance (L) plots. In a recent work[2] on the analysis of X-ray diffraction line profiles due to cold-worked silver obtained by Geiger counter X-ray diffractometer, no such hook was observed, while in another work[3] on the analysis of X-ray diffraction line profiles of cold-worked molybdenum by a photographic method, the hook was quite pronounced. In order to investigate as to why the hook is produced in the photographic method and not in the diffractometer method, the 111 and 222 reflexions of cold-worked aluminium were examined by photographic as well as diffractometer method.

The aluminium examined was spectroscopically pure and was supplied by Johnson Matthey and Co., Ltd., London. Spectroscopic examination revealed that aluminium rod contained an estimated amount of 0·0002 per cent of silicon, 0·001 per cent of magnesium, 0·0005 per cent of iron and copper and less than 0·0005 per cent of tin, cadmium, sodium and calcium. Filings from an aluminium rod were allowed to pass through a sieve of 325 mesh/in.[2] and the samples for the Geiger counter diffractometer were prepared in the form of flat-faced briquettes with a binder. The X-ray diffraction patterns of the cold-worked and annealed materials were measured with a Geiger counter diffractometer. To avoid the asymmetric broadening and shift of the peaks by rate meter recording the line shapes were recorded manually and fixed-count measurements were made of the diffracted intensity, at intervals in 2θ varying from 0·5° in the background to 0·01° in the lines using an electronic scaling unit and an EIT electronic counter. For a photographic method, an 11·46 cm diameter camera was used. All the films were processed under the same conditions. The line intensity measurements were carried out by an Adam Hilger non-recording type of microphotometer. The intensity scale was calibrated through suitably prepared standard wedges.

The Fourier coefficients for 111 and 222 reflexions were determined by Beevers–Lipson strips. For comparison of the Fourier coefficients of the line profiles determined by two methods, the coefficients of pure X-ray diffraction line profiles were determined by eliminating the geometrical broadening by the method of Stokes[4]. For eliminating the instrumental broadening the line profiles from fully annealed specimens were used as standard.