Using  $V^*$ , the values of O for these two flow laws are 121.2 and 106.2 kcal/mole for Post and for Carter and Ave'Lallemant, respectively. The equation for the extrapolation of these mechanical data is, following use of appropriate conversion factors,

$$\dot{\epsilon} = A \exp - \left(\frac{Q + 13.4 \times 2.39 \times 10^{-2}P}{RT \times 10^{-3}}\right) \sigma'' = A \exp - \left(\frac{Q + 0.32P}{RT \times 10^{-3}}\right) \sigma''$$
(9)

where Q is expressed in kilocalories per mole, P and  $\sigma$  are in kilobars, and the various material constants of Eq. 8, in addition to the new values of Q, are to be inserted appropriately.

Figure 1 shows the variation in differential stress at an assumed constant strain rate of  $10^{-14}$  sec<sup>-1</sup> for both the dry and damp flow laws (dashed curves) under ocean basins (Fig. 1a) and continents (Fig. 1b). The geotherms used in the extrapolations are the most recent ones of Mercier (5), based on pyroxene geothermometry and geobarometry. The solid curves in Fig. 1 are also obtained from Mercier's geotherms and from determinations of recrystallized grain sizes, which are functions of stress only, in xenoliths based on the new data of Ross et al. (9). On the basis of the analysis of Ave'Lallemant et al. (20), the solid curves should be regarded as average estimates of variations of stress with depth under the continents and oceans. It is evident from Fig. 1A that extrapolation of Post's data to a strain rate of 10<sup>-14</sup>  $sec^{-1}$  fits the measured oceanic stress data very well, whereas the results of Carter and Ave'Lallemant do not. However, under cratons (Fig. 1b), extrapolation of the latter flow equation provides the better fit, at least at depths above about 175 km. A similar observation by Mercier *et al*. (7) led them to suggest that the upper mantle beneath continents contains small but significant quantities of water, and more than beneath oceans, a suggestion supported by the presence of hydrous phases in many mantle xenoliths from beneath continents.

Following the procedure of Mercier et al. (7), we have attempted to estimate variations of strain rate and viscosity with depth, assuming that the measured stresses are correct and inserting these, as a function of depth, into Eq. 9. Strain rates and equivalent viscosities ( $\eta =$  $\sigma/3\dot{\epsilon}$ ) are then calculated as a function of depth (Fig. 2). The mechanical results of Carter and Ave'Lallemant can be ignored for the upper mantle beneath SCIENCE, VOL. 203, 19 JANUARY 1979

ocean basins (solid lines), and only Post's results need be considered. These results indicate a rapid increase in  $\dot{\epsilon}$  with depth to a value of  $10^{-14.25}$  sec<sup>-1</sup> at 60 km (near the lid of the low-velocity zone) and then a slight decrease and leveling at about  $10^{-14.75}$  sec<sup>-1</sup> at depths below 100 km. The value of  $\eta$  decreases to about 10<sup>21.5</sup> poises at 60 km and remains nearly constant thereafter. Under continents, the values of  $\dot{\epsilon}$  and  $\eta$  calculated from Carter and Ave'Lallemant's results seem more reasonable to depths of about 150 km, below which, perhaps because of dehydration, Post's data may be the more reliable.

The analysis given above indicates the importance of the determination of  $V^*$ for olivine. Stress generally decreases with depth in the depth interval for which observations are available, and the strain rate generally increases and viscosity decreases except under ocean basins. We believe that these conclusions are physically reasonable and should be incorporated as constraints into future attempts to model thermal convection and plate motions.

JOHN V. Ross Department of Geological Sciences, University of British Columbia, Vancouver, Canada V6T 1W5

HANS G. AVE'LALLEMANT Department of Geology, Rice University, Houston, Texas 77001

NEVILLE L. CARTER\*

Department of Earth and Space Sciences, State University of New York, Stony Brook 11794

### **References and Notes**

- 1. R. J. O'Connell, Tectonophysics 38, 119 (1977).
- 2. C. G. Sammis, J. C. Smith, G. Schubert, C. A. Yuen, J. Geophys. Res. 82, 3747 (1977).
- 3. N. L Carter, Rev. Geophys. Space Phys. 14, 301 (1976)
- 301 (1976).
   J.-C. C. Mercier and N. L. Carter, J. Geophys. Res. 80, 3349 (1975); J.-C. C. Mercier, Am. Min-eral. 61, 603 (1976).
   J.-C. C. Mercier, J. Geophys. Res., in press.
   R. L. Post, Jr., Tectonophysics 42, 75 (1977).

- R. L. Post, Jr., Tectonophysics 42, 75 (1977).
   J.-C. C. Mercier, D. A. Anderson, N. L. Carter, Pure Appl. Geophys. 115, 199 (1977).
   R. J. Twiss, *ibid.* 115, 188 (1977).
   J. V. Ross, H. G. Ave'Lallemant, N. L. Carter, J. Geophys. Res., in press.
   R. L. Stocker and M. F. Ashby, Rev. Geophys. Space Phys. 11, 391 (1973); C. Froidevaux and G. Schubert, J. Geophys. Res. 80, 2553 (1975).
   N. L. Carter and H. G. Ave'Lallemant, Geol. Soc. Am. Bull. 80, 2181 (1970); C. B. Raleigh and S. H. Kirby, Mineral. Soc. Am. Spec. Pap. 3 (1970), p. 113; S. H. Kirby and C. B. Raleigh, Tectonophysics 19, 165 (1973).
- Tectonophysics 19, 165 (1973).
   D. L. Kohlstedt and C. Goetze, J. Geophys. Res. 79, 2045 (1974).
- Res. 79, 2045 (19/4).
  13. O. D. Shirby and M. T. Simnad, Trans. Am. Soc. Met. 54, 227 (1961); P. G. Shewmon, Dif-fusion in Solids (McGraw-Hill, New York, 1963); J. Weertman, Rev. Geophys. Space Phys. 145 (1970)
- J. Kubat and M. Rigdahl, *Mater. Sci. Eng.* 24, 223 (1976); *Phys. Status Solidi A* 35, 173 (1976);
   E. H. Rutter *et al.*, in preparation.
   J. J. Gilman and W. G. Johnston, *J. Appl. Phys.* 20 (1976)
- J. Gilman and W. G. Johnston, J. Appl. Phys. 30, 129 (1959).
   H. W. Green et al., in Experimental and Natural Rock Deformation, P. Paulitsch, Ed. (Springer, New York, 1970), p. 272.
   J. D. Blacic, thesis, University of California, Los Angeles (1971); J. M. Edmonds and M. S. Paterson Contrib. Mingral Paterol 30, 141
- Paterson, Contrib. Mineral. Petrol. 30, 141 (1971)
- (1971).
   (1971).
   (1) C. Getting and G. C. Kennedy, J. Appl. Phys.
   (1) 41, 4552 (1971); H. K. Mao, P. M. Bell, J. L. England, Carnegie Inst. Washington Yearb. 70, 18. 281 (1970).
- (1970).
   J. Meisner, Carnegie Inst. Washington Publ. 634 (1974), p.129.
   H. G. Ave Lallemant, J.-C. C. Mercier, J. V. Ross, N. L. Carter, J. Geophys. Res., in press.
   We gratefully acknowledge discussions with D. Mercier and
- We gratefully acknowledge discussions Anderson, D. Kohlstedt, J.-C. C. Merc C. Mercier, and All derson, D. Konsteut, J.-C. C. Mercler, and A. Singer as well as the critical comments of an anonymous reviewer. This research was sup-ported by NSF grant EAR 7604129. Present address: Department of Geophysics, Texas A&M University, College Station 77843.

3 February 1978; revised 11 July 1978

# Neutron Scattering of Solution-Grown Polymer Crystals: Molecular Dimensions Are Insensitive to Molecular Weight

Abstract. Neutron scattering gives information on molecular conformations in solid solutions of polymers of one isotope in another. Results on crystals of polyethylene grown from solution show a molecular dimension (in the form of a radius of gyration) that is almost invariant with the length of the chain. It is proposed that certain lengths of folded chains fold back onto themselves to form stacks of chain-folded ribbons ("superfolding").

The way in which long polymer chains accommodate themselves within crystal units has recently attracted renewed interest with the application of neutron scattering. This technique, with the use of isotopically labeled molecules, gives information on the trajectories of single molecules in the condensed state. Other techniques have established (1) that the crystal unit is characteristically lamellar, with the chain direction usually at an angle of  $0^{\circ}$  to  $30^{\circ}$  to the lamellar normals. The very existence of lamellae is now quite certain, but there is still lively discussion on precisely how the chains are folded into the crystallites. For crystal growth from dilute solution there has been considerable evidence that, for the most part, each chain tends to fold back on itself along the growth face of the crystal (1). Recent experiments have made it possible to put such models to more definitive tests.

The low-angle neutron scattering tech-

0036-8075/79/0119-0263\$00.50/0 Copyright © 1979 AAAS

Table 1. Values of  $R_g$  and derived molecular dimensions (in angstroms);  $M_w$  is the weightaverage molecular weight of solute and  $M_n$  is the number-average molecular weight of solute. The  $R_z$  values are approximate estimates only, based on differences (see text) between oriented mats and mats with little orientation, the degree of orientation being assessed by wide-angle x-ray diffraction. Predicted values of  $R_z$  are as follows: all in one lamella, 30 Å; equally in two lamellae, 60 Å.

$M_{ m w} \ ( imes 10^{-3})$	$\begin{array}{c} M_{\rm n} \\ (\times \ 10^{-3}) \end{array}$	$R_{\rm g}$ , oriented	$R_{\rm g}$ , unoriented	$\sqrt{R_x^2 + R_y^2}$	$R_z$	$R_x$
93	58	69	74	40	48	56
155	84	76	85	44	58	62
183	64	76	86	44	60	61

nique has been extensively applied to polymers in melts, amorphous glasses, and concentrated solutions (2). The results are in general agreement with those obtained earlier by x-ray and light scattering for dilute solutions (3) and provide detailed information on the dependence of the radius of gyration  $(R_g)$  on molecular weight (M). A close correspondence has been found with predictions from statistical mechanics, for example, that  $R_{\alpha} \propto M^{\alpha}$ , where  $\alpha$  is, in general, 1/2 or more. The first result reported for the application of this technique to crystalline polymers (4) was that  $R_{\rm g}$  was unaffected by crystallization from the melt, although a smaller  $R_{\rm g}$  was observed for crystallization from solution (5). These results were obtained under favorable conditions of sample preparation; however, under other conditions the blending of the two isotopic species used [namely, hydrogenous polyethylene (HPE) and deuterated polyethylene (DPE)] is not random (4-6). As a consequence, there is often a contribution to the intensity at low  $s (s = 2 \sin \theta / \lambda)$ , where  $2\theta$  is the scattering angle and  $\lambda$  is the neutron wavelength) that can mask the expected signal corresponding to the form factor of a DPE molecule in its matrix of HPE. This phenomenon [sometimes referred to as "clustering" (4)] is in fact (5, 6) caused by fluctuations in isotope concentration on a scale of the order of microns across the sample, which are caused by finite differences in the crystallization rate between the two species (fractionation). For this reason all the experiments carried out on polyethylene thus far correspond to high supercoolings, in which case the unwanted contribution to the intensity is restricted to small enough s (below about  $10^{-3} \text{ Å}^{-1}$ ) that  $R_{\rm g}$  can be measured reliably. The values of  $R_{\rm g}$  will depend on many variables (for example, the structure of the crystal surfaces), but the technique can provide additional information if measurements at s up to 0.06  $Å^{-1}$  are made. Of these experiments we may first distinguish the s range 0.02 to 0.06  $Å^{-1}$  (5, 7), where the scattering is determined almost entirely by the proximity or lack of proximity of "stems" (chain traverses of the lamellae) from contiguous parts of a molecule. These results were interpreted on the basis of analytic expressions which are relatively simple if the (small) amount of material not in the stems is neglected. The mode of folding is not unique; the mutual arrangements of the stems can be effectively random (melt growth) or systematic and compact, very probably with the formation of rows (solution growth). It is likely that these modes represent two extremes. Second, for the range of *s* up to 0.02 Å<sup>-1</sup> (5, 8, 9), numerical calculations (9) have been made for several particular conformations which are specified in some detail and compared with data for melt-grown crystals (8).

We report here more extensive  $R_g$  measurements for solution-grown crystals, with emphasis on their dependence on *M* and also on the aspect ratio of the molecules. The techniques used have been described in detail (5); a mixture of 1 percent DPE in HPE was dissolved in



Fig. 1. Equivalent radii of gyration ( $R_{\rm g}$ ) derived from data of the type shown in figure 3 of (5). Details of sample preparation and background corrections are described there. Triangles refer to molten samples (in this case, the matrix HPE was low *M*;  $M_{\rm w} = 6800$ ,  $M_{\rm n} = 5200$ ). The cross refers to a published value for molten polyethylene (14). The extrapolated "forward intensity" was proportional to the known *M*; this intensity when calibrated corresponded to somewhat less than the known *M* in all cases.

stirred, boiling xylene at 0.2 percent, and crystallization occurred in thin-walled tubes at 70°C. Crystal mats were made either oriented or isotropic. We used the small-angle neutron scattering facility at the Institute Laue Langevin, Grenoble (10); background measurements were carried out on mats of HPE and the sample holder only. By contrast with the  $R_{g}$  values quoted earlier (5) derived with the use of the Guinier formulation, we used here the Zimm method of plotting inverse intensity versus  $s^2$ ;  $2\pi s R_g$  was generally less than about unity. The results are systematically 20 percent larger in the present case since even at small s the two formulations are not equivalent.

Figure 1 shows two important results. The first is the big difference in  $R_g$  between melts and crystals grown from solution, an indication of the fundamental difference between melt and solution growth. Second, there is a rather remarkable insensitivity of  $R_g$  to M; the corresponding value of  $\alpha$  would be approximately 0.1. To our knowledge, it is unprecedented that a molecular dimension of a polymer should vary so little with molecular weight.

Figure 1 refers to oriented mats of crystals. The apparent value of  $R_{g}$  as derived from slopes of plots according to the usual equations for low-angle scattering according to the Zimm or Guinier methods will be dominated by the dimensions in the plane of lamellae. A dimension perpendicular to the lamellae can be derived from a comparison of oriented and unoriented mats (Table 1),  $R_g$  being slightly larger in the latter case. The equations can readily be generalized for application to textured samples. We denote  $R_x$ ,  $R_y$ , and  $R_z$  as measured parallel (x and y) and perpendicular (z) to the lamellae (for example,  $R_z^2 = \langle z^2 \rangle$ , where z is the position of a scattering element; z is not necessarily along the chain).

The apparent  $R_g$  is then given by  $R_g^{2/3} = \langle h^2 \rangle R_x^2 + \langle k^2 \rangle R_y^2 + \langle l^2 \rangle R_z^2$ , where h, k, and l are direction cosines relative to one particular direction perpendicular to the incoming beam. For oriented mats  $\langle h^2 \rangle = \langle k^2 \rangle \approx 1/2$  and  $\langle l^2 \rangle \approx 0$ . Table 1 shows values of both the rootmean-square average of  $R_x$  and  $R_y$  and also an appropriate estimate of  $R_z$  derived in this way. Both the in-plane and out-of-plane dimensions are insensitive to M, although measurements of  $R_z$  are less extensive. The quantity  $R_z$  is somewhat larger than the root-mean-square average of  $R_x$  and  $R_y$ .

On the basis of earlier interpretations (5, 7) that for solution growth the stems are predominantly in rows, a rather spe-

cific model for the complete trajectory for one molecule presents itself. If a molecule continued to deposit as a single row,  $R_{\rm g}$  would have to increase significantly with M (linearly as recorded for oriented mat), which is not observed. Nevertheless, if the rows of stems did not increase in length but did increase in thickness, the observed close invariance of  $R_{g}$  with M would be accounted for. Such a situation would arise if, beyond a certain length, the rows of stems would fold up on themselves. Such a possibility is in fact plausible on a molecular basis. We could refer to the folding back of the ribbons near themselves as "superfolding," although this superfolding is not representatively "adjacent re-entrant." Figure 2 shows a schematic twodimensional representation of the structure. It has always seemed unlikely that one molecule would fold into one uninterrupted ribbon along the growth face; at the very least, there will be competition between different chains for room on the growth face. Such effects are indeed inherent in theories of crystal growth [regime II in (11)]. In particular, when two growing ribbons meet, either one continues to deposit and the remaining parts of the other becomes excluded from the crystal (and may possibly become incorporated elsewhere later) or both double up on themselves. The doubling-up process would represent the least interference with crystallization and is strongly suggested by the present experimental results.

Such a possibility of multiple ribbons does not arise quite unexpectedly. In fact, the idea of the proximity of different rows of stems from the same molecule has already been derived from effects at much higher angles examined in earlier work; the intensities corresponded to multiple rather than single rows of stems for high molecular weight ["interference between ribbons'' (5)]. Use of the mixed crystal infrared technique (12) has shown that, for very-high-molecularweight DPE, stems of the same molecule are in close proximity. Our own infrared measurements, made on the same samples as used here, favor the formation of multiple ribbons (13).

If we suppose that one dimension in the x-y plane is substantially greater than the other, we can derive an estimate for the larger dimension  $[R_x, say;$  in reality this is likely to be along 110 (Table 1)]. From the derived  $R_x$  value an estimate of the longest likely sequence of folded stems is possible, which is 170 Å. With the knowledge of lamellar thickness and stem spacing, we can express this length as corresponding to 40 stems, or M in the

SCIENCE, VOL. 203, 19 JANUARY 1979



Fig. 2. Schematic representation of "stem" positions of two labeled molecules (large dots) as viewed in the chain direction (that is, 001). The 100 direction is shown horizontal, and the fold plane is (110)(1). Not to scale: the length of the rows is likely to be approximately 40 stems.

neighborhood of 50,000. On this basis, the apparent  $R_{g}$  should decrease for Mbelow a value in this range. Unfortunately, at low M the excess low-angle intensity is large compared with the signal from the individual solute molecules, and an  $R_{\rm g}$  is not obtainable. The fact that  $R_x$  is close to  $R_z$  (Table 1) implies sheets that are about equidimensional.

The estimates of  $R_z$  (even though imprecise since these are given by differences in  $R_{\rm g}$  values) are larger than the 30 Å predicted from the known lamellar thickness of 105 Å. We can attribute this to the fact that some chains crystallize partly in neighboring lamellae; for example, a chain may divide itself approximately equally between two lamellae, or between three lamellae with the central lamella containing most of the chain. This is quite reasonable in view of the multilayered morphology of these crystals.

### D. M. SADLER

A. Keller

H. H. Wills Physics Laboratory, University of Bristol, Royal Fort, Tyndall Avenue, Bristol BS8 1TL, England

#### **References and Notes**

- A. Keller, Rep. Prog. Phys. 31, 623 (1968).
   R. G. Kirste, W. A. Kruse, J. Schelten, Makromol. Chem. 162, 299 (1973); D. G. H. Ballard, G. D. Wignall, J. Schelten, Eur. Polym. J. 9, 965 (1973); J. P. Cotton, D. Decker, H. Benoit, B. Farnoux, J. Higgins, G. Janninck, R. Ober, C. Picot, J. des Cloizeaux, Macromolecules 7, 863 (1974); M. Daoud, J. P. Cotton, B. Farnoux, G. Jannick, G. Sarma H. Benoit, B. Duplescix Janninck, G. Sarma, H. Benoit, R. Duplessix, C. Picot, P. G. de Gennes, *ibid.* 8, 804 (1975).
- C. Frich, F. G. de Gennes, *Ibid.* 8, 804 (1975).
   O. Kratky, *Pure Appl. Chem.* 12, 483 (1966).
   J. Schelten, G. D. Wignall, D. G. H. Ballard, *Polymer* 15, 682 (1974).
   D. M. Sadler and A. Keller, *Macromolecules* 10, 1128 (1977).
- G. C. Summerfield, J. S. King, R. Ullman, *ibid*. 11, 218 (1978). 6. C
- 7. D. M. Sadler and A. Keller, Polymer 17, 37
- 8. J. Schelten, D. G. H. Ballard, G. D. Wignall, G.
- J. Schelten, D. G. H. Ballard, G. D. Wignall, G. Longman, W. Schmatz, *ibid.*, p. 751.
   D. Y. Yoon and P. J. Flory, *ibid.* 18, 509 (1977).
   W. Schmatz, T. Springer, J. Schelten, K. Ibel, J. Appl. Crystallogr. 7, 751 (1974).
   J. D. Hoffman, L. J. Frolen, G. S. Ross, J. I. Lauritzen, J. Res. Natl. Bur. Stand. 79A, 671 (1975)
- (1975)12.
- . Krimm and J. H. C. Ching, *Macromolecules* 209 (1972); J. H. C. Ching and S. Krimm, *ibid*. 894 (1975)
- S. Spells, D. M. Sadler, A. Keller, in prepara-13. tion
- G. Lieser, E. W. Fischer, K. Ibel, J. Polym. Sci. Polym. Lett. Ed. 13, 39 (1975).
   We thank the staff of the Institute Laue Lange-
- vin, in particular R. Duplessix, for valuable assistance with the measurements.
- 30 May 1978; revised 11 September 1978

# In vitro Model for Stretch-Induced Hypertrophy of **Skeletal Muscle**

Abstract. Mechanical stretch of embryonic chicken skeletal myotubes developed in vitro leads to many of the biochemical changes seen in skeletal muscle hypertrophy. These include increased amino acid accumulation, increased incorporation of amino acids into general cellular proteins and myosin heavy chains, and increased accumulation of total protein and myosin heavy chains. This model system should aid in understanding how the growth rate of skeletal muscle is regulated by its activity.

Skeletal muscle is a highly adaptive tissue that must continuously respond to changes in the external environment. Thus, muscle activity leads to many biochemical changes, the type of activity determining the changes seen. For example, increased muscle work against high resistance leads to increased muscle size (hypertrophy) (1), whereas repetitive work against little resistance leads to increased oxidative capacity of the muscle (2). Although it has been known for more than 80 years that mechanical stretching of skeletal muscle increases its metabolic rate  $(\beta)$ , the mechanism in-

volved is still unknown. Because muscle is an elastic tissue that stretches and relaxes during activity, knowing the relationship of the mechanical movement of skeletal muscle to its metabolism is essential to understanding how muscle regulates its growth rate and the role of the nervous system in this regulation. Stretch has been implicated in such processes as early muscle growth and development (4), denervation hypertrophy (5), and compensatory hypertrophy (6). In this report we describe an in vitro model system in which it has been shown that mechanical stretch leads to skeletal

0036-8075/79/0119-0265\$00.75/0 Copyright © 1979