Deformation of doubly oriented polyethylene, nylon 66 and poly(ethylene terephthalate)

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Doubly oriented specimens of high density polyethylene (HDPE), nylon 66 and poly(ethylene terephthalate) (PET) were re-stretched along a direction perpendicular to the molecular chain axis at temperatures ranging between room temperature and the respective polymer melting points. Brittle failure was observed for PE samples at all the test temperatures with no significant amount of plastic deformation; whereas, for both PET and nylon 66 samples, ductile deformation was observed at elevated temperatures with plastic strain of >400%. The ductile deformation of nylon 66 and PET occurred with an anisotropic change in the cross-sectional dimensions of the specimen, the reduction taking place predominantly in only one lateral direction. The morphological change accompanying the drawing of the doubly oriented PET and nylon 66 material was examined by using X-ray and optical methods. The implications of the difference in deformation behaviour with respect to the morphological differences among oriented PE, PET and nylon 66 materials are discussed.

INTRODUCTION

Doubly oriented specimens of PE, PET, and nylons, etc., have been studied extensively in terms of their morphologies and deformation behaviour[1-9]. Polymeric materials so oriented were chosen for study because the crystallographic directions of the microcrystalline structure are well-aligned throughout the whole sample. To some extent, the entire macroscopic sample can be regarded as an aggregate of three-dimensionally oriented single crystals. This arrangement of crystallites is simpler than that of a spherulitic structure, making measurement of structure parameters less ambiguous, which in turn permits elucidation of various deformation mechanisms with fewer difficulties.

Folded-chain lamellae with adjacent reentry of the crystallizing chain have been proposed as the common structural entity of the crystalline flexible-chain polymer materials. If such an organization prevails one could expect to maximize the possibility of unfolding the chains by suitably aligning the lamellae of doubly oriented samples with respect to the stretch direction.

A folded-chain lamellar structure with random switchboard re-entry has also been considered as the basis crystalline unit for unoriented materials[10]. If such a structure prevails in doubly oriented samples, it would preclude in general the possibility of unfolding the chains.

Yielding and plastic deformation along various directions of the doubly oriented polymers has been studied using PET[1,5], polypropylene (PP)[1], PE[1,7,11] and nylons[7,13]. Although tensile stress has been applied perpendicular to the molecular axis of the lamellae, the test temperature was invariably room temperature; no appreciable amount of plastic flow was observed when the stretch axis was 90° to the orientation axis[5]. It is believed that the intermolecular binding energy within the lamella can be overcome more easily at elevated temperatures, as with most thermally activated processes. Accordingly, elevated temperatures were used in this work on unfolding the chains of the lamellae.

Both wide angle and small angle X-ray scattering (WAXS and SAXS) techniques were used to examine the crystalline structure before and after the doubly oriented samples were deformed further. Birefringence values for some samples of PET were also measured to help explanation of the deformation mechanism because birefringence information is readily available for this particular material[2,6,14,15].

We find that the difference in the deformation behaviour between doubly oriented PE, and each PET and nylon 66 has significant implications as to basic structural differences between these materials.

EXPERIMENTAL

The materials used were Alathon* 7040 high density PE, Vydyne† 21 nylon 66 and VFR‡ 5041 PET. These materials were first injection-moulded into bars having a cross-section of 6.35 mm × 12.7 mm. The moulded bars can essentially be regarded as isotropic because no significant amount of birefringence was observed.

The doubly orienting process used was the same as that used by Young *et al.*[16]. The slot width was 12.7 mm and the samples were squashed to their respective ultimate 'draw' ratios before rupture. For both PET and nylon 66 the ultimate draw ratio was ~5.0; for the PE it was 14.0. The processing temperature for each PET and nylon 66 was 200°C and for PE, 100°C.

No attempt was made in this work to use a doubly oriented PE sample of only five times draw ratio, because the morphology of a fourteen times drawn PE

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* Trademark of DuPont Co., the density of the Alathon 7040 is 0.0960; $\bar{M}_n = 25\,000$ and $\bar{M}_w = 84\,000$.

† Trademark of Monsanto Co., $[\eta] = 1.37$ at 25°C at 90% formic acid.

‡ Trademark of Goodyear Co., $[\eta] = 0.90$ at 25°C in phenol/tetrachloroethene (60/40).

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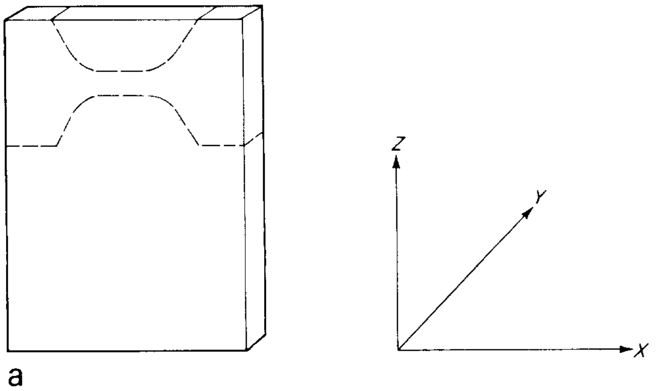

Figure 1a Reference axes of the doubly oriented sample

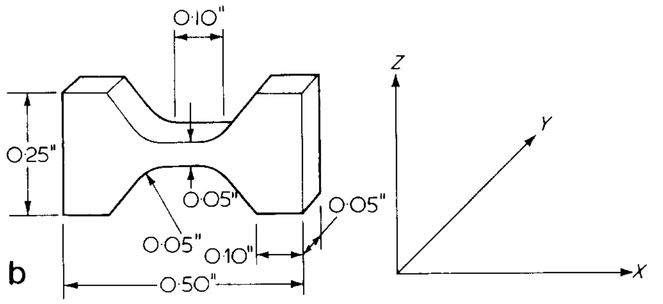

Figure 1b Dimensions and reference axes of the tensile test sample cut from the doubly oriented material

is more comparable to that of a five times drawn nylon or PET. The natural draw ratio after neck formation is ~9 for PE, whereas it is 2 or 3 for PET or nylon[25]. The molecules of a five times drawn PE are much less oriented compared with PET or nylon molecules in samples of the same draw ratio.

The sample geometry of the doubly oriented materials is shown in *Figure* 1*a*. During the double-orientation process, the material was elongated along the Z-axis and compressed along the Y-axis while the dimension in X-axis remained unchanged.

The tensile test samples were cut from the doubly oriented specimen as indicated in *Figure* 1*a* and machined to the dimensions shown in *Figure* 1*b*. The gauge portion of the test samples had a square cross-section.

The tensile tests were carried out on a table top Instron machine equipped with miniature sample clamps and a heating chamber. A thermocouple was placed adjacent to the gauge section of the sample for temperature monitoring; a dry nitrogen atmosphere was maintained in the heating chamber throughout the test. A crosshead speed of 0.5 cm min^{-1} was used for all the tensile tests which corresponded to ~200% min^{-1} strain rate.

WAXS and SAXS patterns were obtained along the three axes of the doubly oriented and the re-stretched specimens. An X-ray beam from a copper fine focus tube was used at 40 kV and 26.25 mA; the diffraction pattern was recorded with a Statton-type camera[17], the film to sample distance being 5 cm for WAXS and 32 cm for SAXS.

The birefringence values of the PET samples were measured on thin sections using a polarizing microscope equipped with a Berek compensator. The scanning electron micrographs were obtained using a JEOL SEM-15 microscope operated at 20 kV; the samples were coated with Pd–Au alloy for observation.

RESULTS

General description of the deformation process

The stretch axis (X) for the PE samples was parallel to the [010] direction and all samples failed in a brittle mode. The stress–strain relationships at 90, 110 and 130°C (*Figure* 2) indicate that all samples began fracture in the vicinity of 10% strain. The fracture surface was inclined ~45° to the tensile axis and contained the sample Z-axis along which the molecules were oriented by the original double-orientation process. Microfibrillar structure was prominent on the fracture surface; fracture seemed to occur by shearing the interfibrillar region.

A different deformation behaviour was observed for both nylon 66 and PET samples when they were stretched parallel to X. A tensile strain ~400% was obtained at elevated temperature (*Figures* 3 and 4), however, the ultimate tensile strain did not increase monotonically with temperature. For nylon, 125–220°C is the preferred temperature range for plastic flow, whereas the upper temperature was somewhat lower for PET. For both materials the large drop in stress

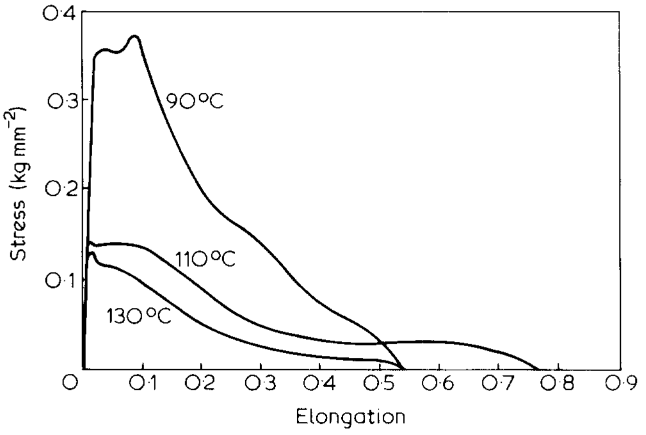

Figure 2 Stress versus elongation curves of the doubly oriented high density PE at various temperatures. The tensile axis is perpendicular to the molecular chain direction and lies in the [100] plane. Deformation rate ≅2.0 min^{-1}

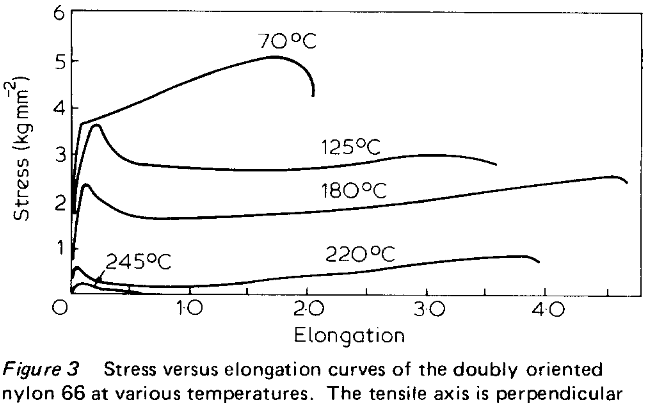

Figure 3 Stress versus elongation curves of the doubly oriented nylon 66 at various temperatures. The tensile axis is perpendicular to the molecular chain direction and lies in the [010] plane. Deformation rate ≅2.0 min^{-1}

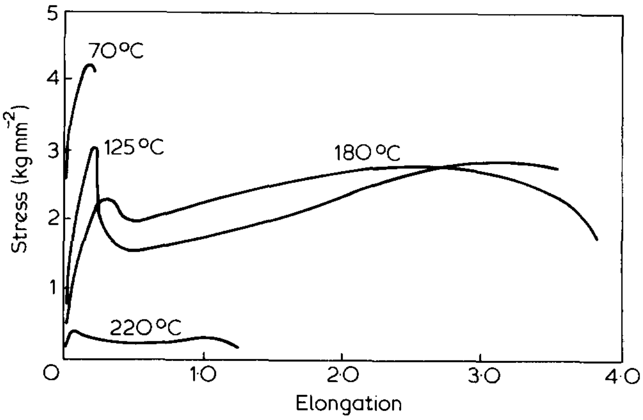

Figure 4 Stress versus elongation curves of the doubly oriented PET at various temperatures. The tensile axis is perpendicular to the molecular chain direction and lies in the [100] plane. Deformation rate ≅2.0 min^{-1}

during the yielding stage was also prominent on the stress–strain curves over a certain temperature range.

Upon deformation there was a dramatic change in the ratio of the cross-sectional dimensions of the samples. The sample dimension along the Y-axis remained almost unchanged for both nylon 66 and PET; all the lateral contraction took place mainly along the Z-axis for both materials. Such deformation converted the original square cross-section of the specimen into a rectangular one with the cross-section retaining sharp corners (*Figure* 5).

The maximum draw ratio for isotropic nylon 66 and PET is about five — further drawing invariably results in sample failure. Although the molecules were originally oriented perpendicular to the stretch direction the maximum draw ratio was still only five times. The limited ultimate draw ratio of five precludes extensive unzipping of the folded-chain lamellae during the stretch process. A draw ratio of the order of 20 would be expected for the complete unzipping process, since that is the ratio between the lamellar thickness and the average lateral distance between molecules. Therefore even though the tensile stress was applied perpendicular to the chains inside the lamellae, it was not found possible to unzip the folded chains.

Structure revealed by WAXS and SAXS

(1) *Polyethylene:* No significant amount of plastic deformation took place in the re-stretching process of the doubly-oriented tensile test specimen, therefore, only the undeformed doubly-oriented material was studied by X-ray analysis. The results (*Figure* 6) indicate that the morphology arising from the squashing process is almost identical to that of the well-documented stretched and rolled PE film[8,18,19]. The crystallite c-axis lies in the Z direction and the [100] plane coincides with the squashed surface denoted as the $X-Z$ plane (*Figure* 1). The occurrence of [110] or [310] twinning[23] is evident in *Figure* 6*c*. Additionally diffraction spots due to the monoclinic phase are also evident in both *Figures* 6*a* and 6*b*.

The only significant difference between the supermolecular structure of this material and that of the drawn/rolled one[19] is revealed by the SAXS pattern taken along the X-axis (*Figure* 6*e*). There was no change in the SAXS pattern showing that there was no change in the lamellar arrangement across the whole thickness, whereas a planar symmetry exists in the lamellar structure on the Y-Z plane of the stretched and rolled PE film. A mosaic structure model such as that proposed by Yamada and Takayanagi[8] can be used to accommodate the X-ray patterns shown in *Figure* 6.

(2) *Nylon* 66*:* The morphology of the doubly oriented nylon 66 in this work is essentially the same as that reported by others[3]. The only noticeable difference is the existence of a strong $[001]_\alpha$ diffraction maximum (*Figures* 7*b* and 7*f*) which was not observed by Bunn and Garner[24]; the presence of intense $[001]_\alpha$ suggests a possible deviation of the chain configuration such as rotation of certain segments of the polymer chain in the crystalline regions compared with Bunn and Garner's model. The presence of crystalline β phase structure is also evident from *Figure* 7*a*. The [010] plane on which the hydrogen bonds lie is parallel to the squashed surface ($X-Z$ plane) as shown by the intensity of the [010] spot on the meridian when the X-ray beam is parallel to the Z-axis (*Figure* 7*c*).

The WAXS patterns along the X, Y, and Z axes were also obtained in the restretched section of the tensile test sample. The WAXS patterns for the sample deformed at 125°C with a stretch ratio of 4.5 times are shown in *Figures* 7*d*, *e* and *f*. The crystallite c-axis has been rotated by 90° and is aligned with the X-axis due to restretching. The stretched section is still doubly oriented with the [010] plane remaining parallel to X-Z surface, but the degree of double orientation is less than in the unstretched doubly oriented material. This is evident from the relative intensities of the [100] and [010] spots on *Figures* 7*d*, *f*.

The SAXS patterns of both the original doubly oriented portion and the restretched portion are shown in *Figures* 8*a*–*c* and 8*d*–*f* respectively. The structural model of the type proposed by Yamada *et al.*[8] was found to be also applicable for representing the supramolecular morphology of both the doubly oriented nylon 66 sample and the restretched sample, but to a smaller extent for the latter.

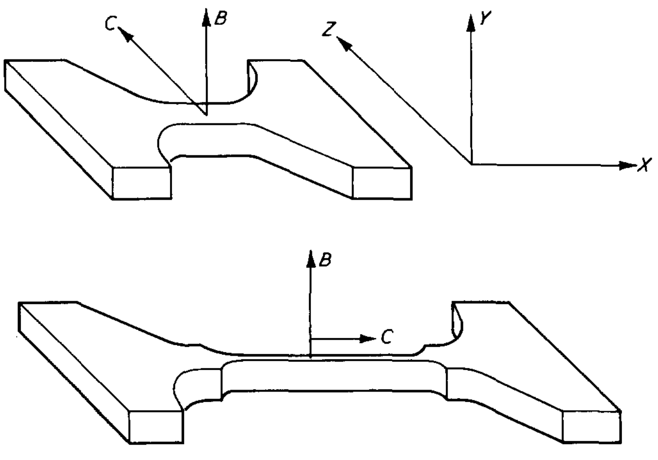

Figure 5 Schematic demonstration of the change in specimen shape and the crystallgraphic directions upon restretching; *c* axis stands for the chain direction and *b* axis stands for [010] in nylon 66, [100] in PET

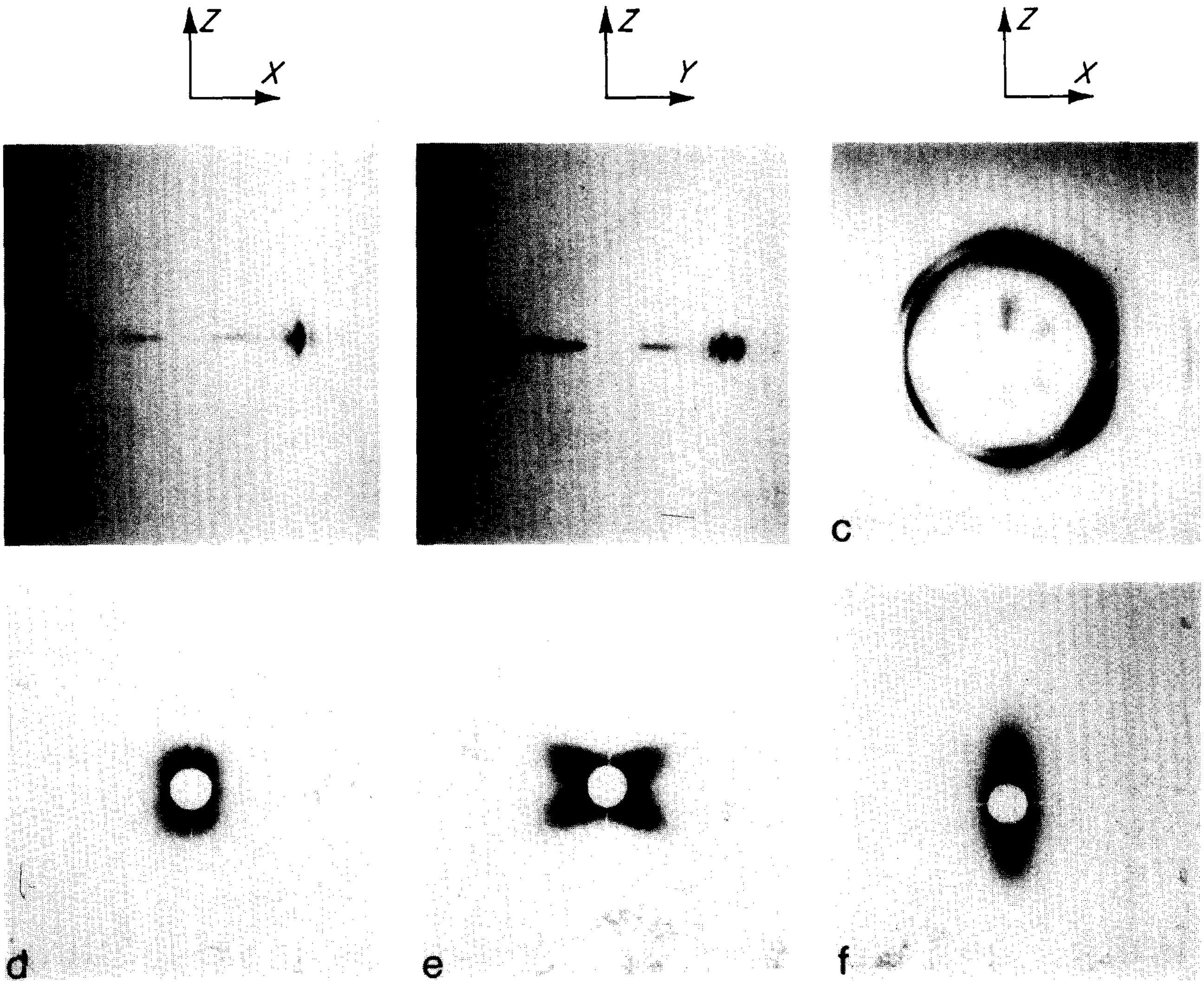

Figure 6 The X-ray scattering patterns of the doubly oriented HDPE sample along the axis designated respectively. The sample was stretched at 100°C. *Figures a–c,* WAXS; *Figures d–f,* SAXS

The long periodicity of the lamellar structure measured from the SAXS pattern along the chain orientation direction (*Figures* 8*a*, *b*) was found to be 97 Å for the doubly oriented material which had been squashed at 200°C. The same value of long periodicity was also obtained from the restretched sample which was drawn at 200°C. In contrast, SAXS-measured periodicity on the sample restretched at 125°C (*Figures* 8*d*, *f*) indicated a long period of only 80 Å. It is evident that the long period of restretched samples is temperature dependent as observed by Peterlin *et al.* in their drawing work on PE[20].

(3) *Poly(ethylene terephthalate)*: As revealed by the WAXS and SAXS patterns shown in *Figures* 9*a*, *b*, *c* and 10*a*, *b*, *c* respectively, the crystalline orientation in the doubly oriented PET material made by squashing is similar to that observed by others[4,9] in one-way stretched PET film. The crystallite [100] plane is parallel to the squashed surface (the X–Z plane). After being restretched 4.5 times at 125°C, the material is marginally doubly oriented as indicated by the relative intensity difference between the [100] and [010] diffraction arcs in *Figures* 9*d*, *f*. It is interesting to see that the degree of perfection of the double orientation of the original sample is substantially different between nylon 66 and PET even though the squashing conditions used for these two polymers were identical. Nevertheless, despite the PET sample being less doubly oriented than the nylon 66, no difference was found between their deformation behaviour. Lateral contraction of the PET tensile sample was greatest along the Z-direction, and very little change in the lateral dimension along the Y-axis was found, as was the case for nylon 66.

The WAXS patterns (*Figures* 9*d* and *f*) indicates that the PET molecules in the crystalline phase are highly reoriented along the restretching direction after a draw ratio of ~4.5 from a state originally perpendicular to the stretch direction. it is of interest to examine the average orientation of the PET samples both before and after the restretching process by measuring the birefringence before and after.

Birefringence of doubly oriented and restretched PET

A similar molecular reorientation process is believed to occur during the restretching process of both nylon 66 and PET. The birefringence study was made only on the PET samples; measurements being made on the actual samples used in the X-ray study. The birefringence values of both the original double-oriented

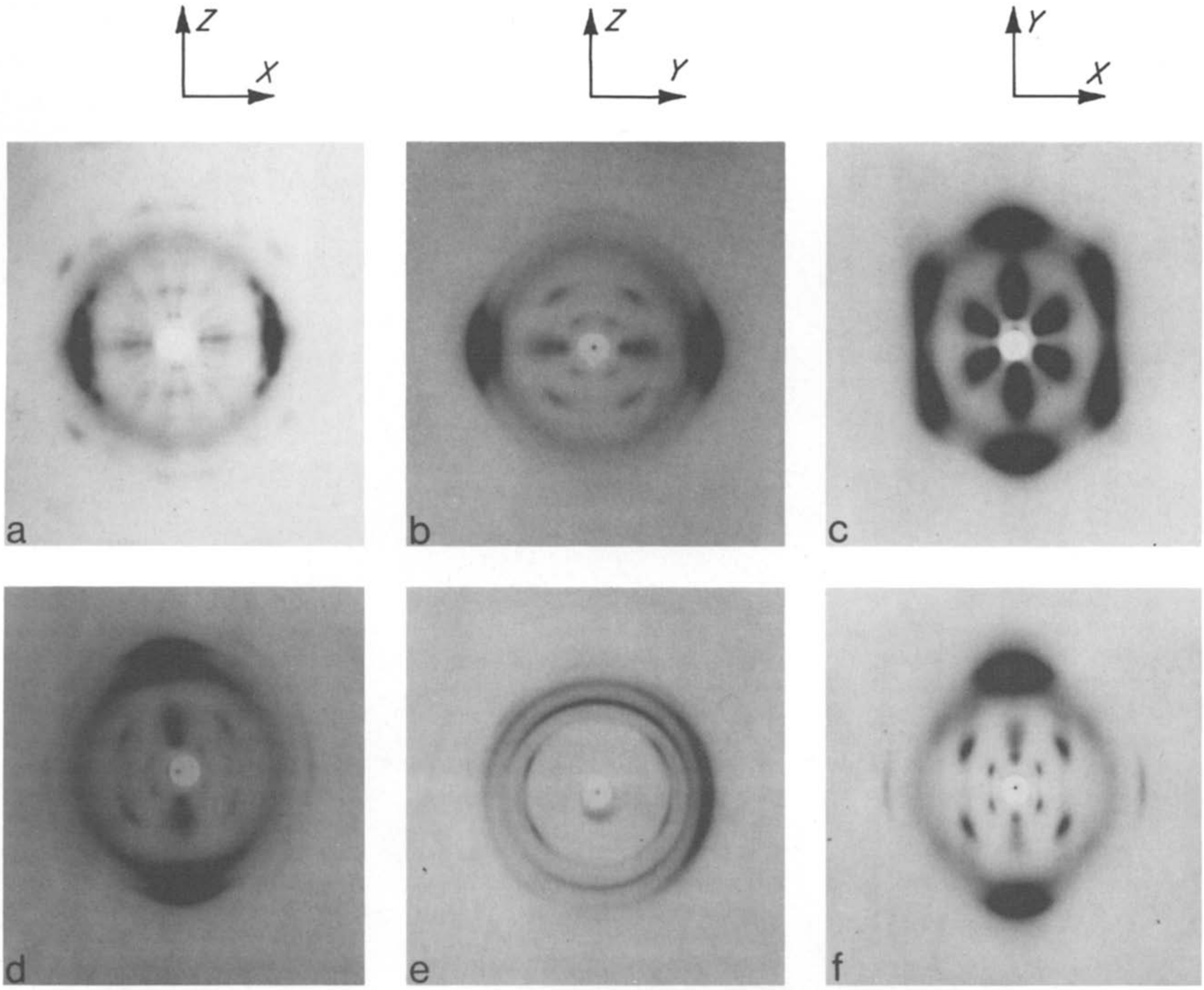

Figure 7 The WAXS patterns of both the original double oriented and the restretched nylon 66 samples along the axes designated respectively. The samples were doubly oriented at 200°C and restretched at 125°C. *Figures a–c,* original doubly oriented specimen; *Figures d–f,* restretched specimen

material and the restretched portion are listed in *Table* 1 averaged over five measurements. The birefringence, n_z–n_y, of the restretched sample is hard to measure because the sample dimension in the *Z*-axis makes it difficult to obtain useful specimens by microtoming. The magnitude of the change found in the values of n_z–n_x was unexpected in light of the fact that the birefringence value of an oriented fibre drawn 4.5 times from the isotropic state is in the range of 0.180[14,15]. The implication of the change of birefringence will be discussed later.

DISCUSSION

Structure models for oriented PE, PET and nylon 66

One of the well-accepted structural models for oriented polymers is the microfibrillar one[21,22]. The microfibrillar model has not only been found to accommodate the structural information obtained upon plastic deformation using techniques such as X-ray scattering and electron microscopy, but this model has also been successful for relating the mechanical behaviour, i.e. strength, modulus and fracture, to microfibrillar structure parameters. In addition to its application to polyethylene and polypropylene, the microfibrillar model has been proposed for other uniaxially oriented polymers such as nylons, polyoxymethylene (POM).

One of the essences of the microfibrillar model is the existence of interfibrillar regions which are believed to be the weak parts in oriented polymers[21]. The major deformation mechanism proposed for oriented polymers with a microfibrillar structure is the shear of the interfibrillar regions. Rupture of the sample is considered to occur by failure at these weaker regions. Otherwise, if the strength of the interfibrillar regions were the same as that of the interior of the microfibrils or fibrils then a question would arise as to the necessity of invoking the microfibrillar model to account for the mechanical properties and fracture behaviour even though the data from X-ray scattering and electron microscopy, etc. are consistent with such a model.

A difference in the ultimate draw ratio in the lateral restretch experiments was observed among the doubly-oriented nylon 66, PET and PE samples. The PE samples failed in a brittle fashion with fibrillar struc-

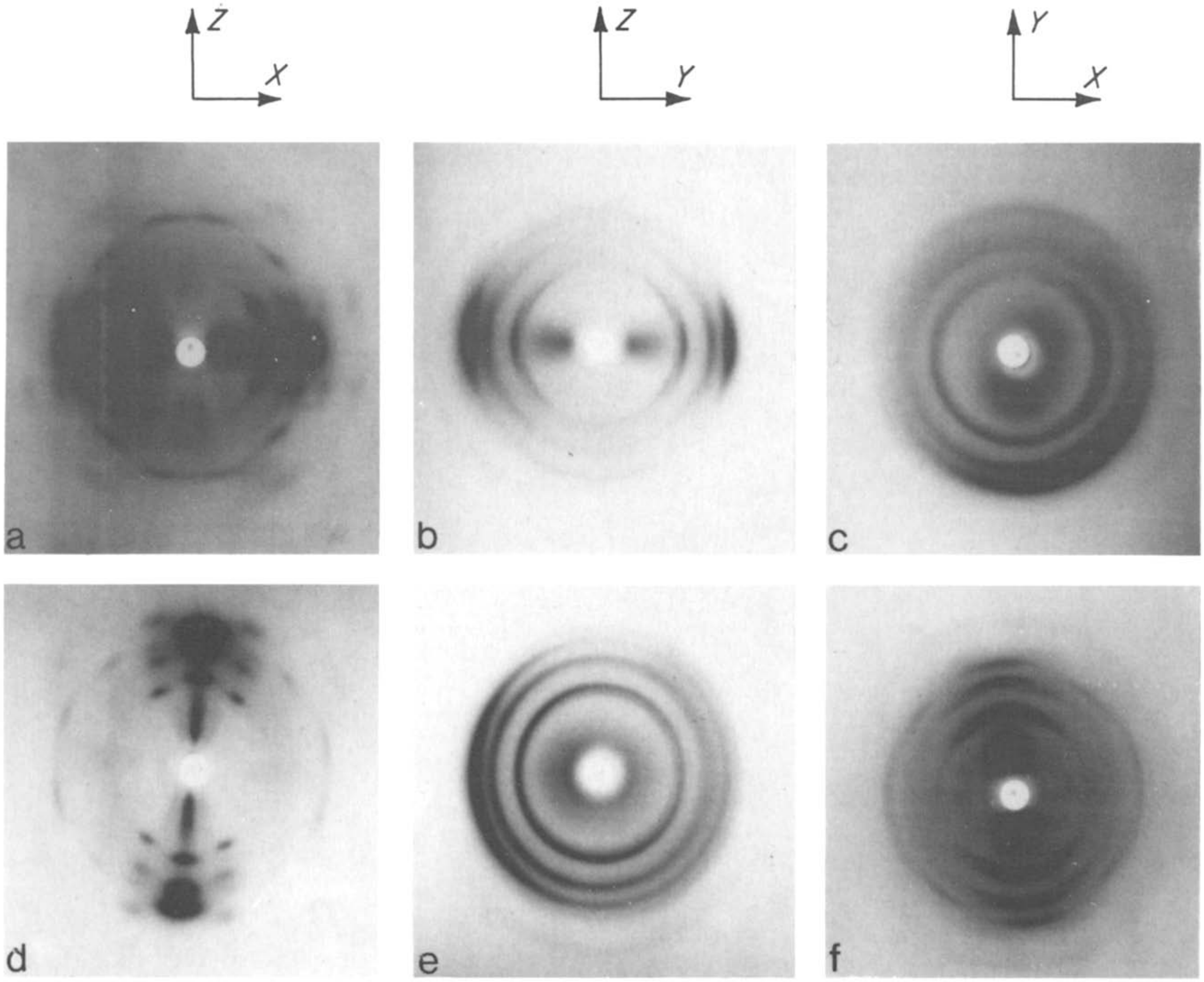

Figure 8 The SAXS patterns of both the original doubly oriented and the restretched nylon 66 samples along the axes designated respectively. The samples were doubly oriented at 200°C and restretched at 125°C. *Figures a–c*, original doubly oriented specimen; *Figures d–f*, restretched specimen

ture left on the fracture surface, while both PET and nylon 66 samples were stretched 4–5 times before failure. This result strongly suggests that the microfibrillar model is not as applicable to doubly oriented PET or nylon 66 as to doubly oriented PE in their respective ultimate doubly oriented states. The failure of the fully oriented (X14) PE samples to exhibit a highly cohesive nature in the lateral direction like the doubly oriented PET and nylon 66 does not preclude the possibility that a PE sample doubly stretched to only 5 times its original length could be restretched out. On the other hand, a five times stretched PE does not have the same degree of double orientation as PET or nylon 66 having the same stretch ratio[25]. One of the purposes of this work was to demonstrate the structure difference between fully doubly-oriented PE and other polymers.

Deformation mechanisms of doubly oriented PET and nylon 66

Various slip mechanisms, such as interfibrillar, interlamellar and intralamellar slips, have been proposed for plastic deformation of crystalline polymers in general[7,23]. If indeed slip mechanisms are the bases for plastic deformation in such polymers, then the fact that the strain along the Y-axis approximates to zero during the restretching of doubly oriented nylon 66 and PET indicates that the slip systems must be perpendicular to the Y-axis, [010] for nylon 66 and [100] for PET. A slip system is defined by the combination of a given slip plane and a given slip direction; that a slip system is perpendicular to a certain axis means either the normal of the slip plane or the slip direction is perpendicular to that axis. The present finding of zero strain along the Y-axis neither confirms nor rules out the importance of slip mechanisms in the deformation of nylon 66 and PET however, this result does preclude certain slip systems as active ones during plastic deformation. The precluded slip systems are these which have components in both the applied stress direction and the sample Y axis.

Apparent slip traces on the X–Z plane of the restretched tensile bars have been observed near the shoulder region of a PET sample restretched at 125°C (*Figure* 11). These traces are ill-defined due to the lack of the out-of-plane displacement component. No attempt has been made to explain the details of the microdeformation mechanisms involved in the formation of these slip traces. It is important not to

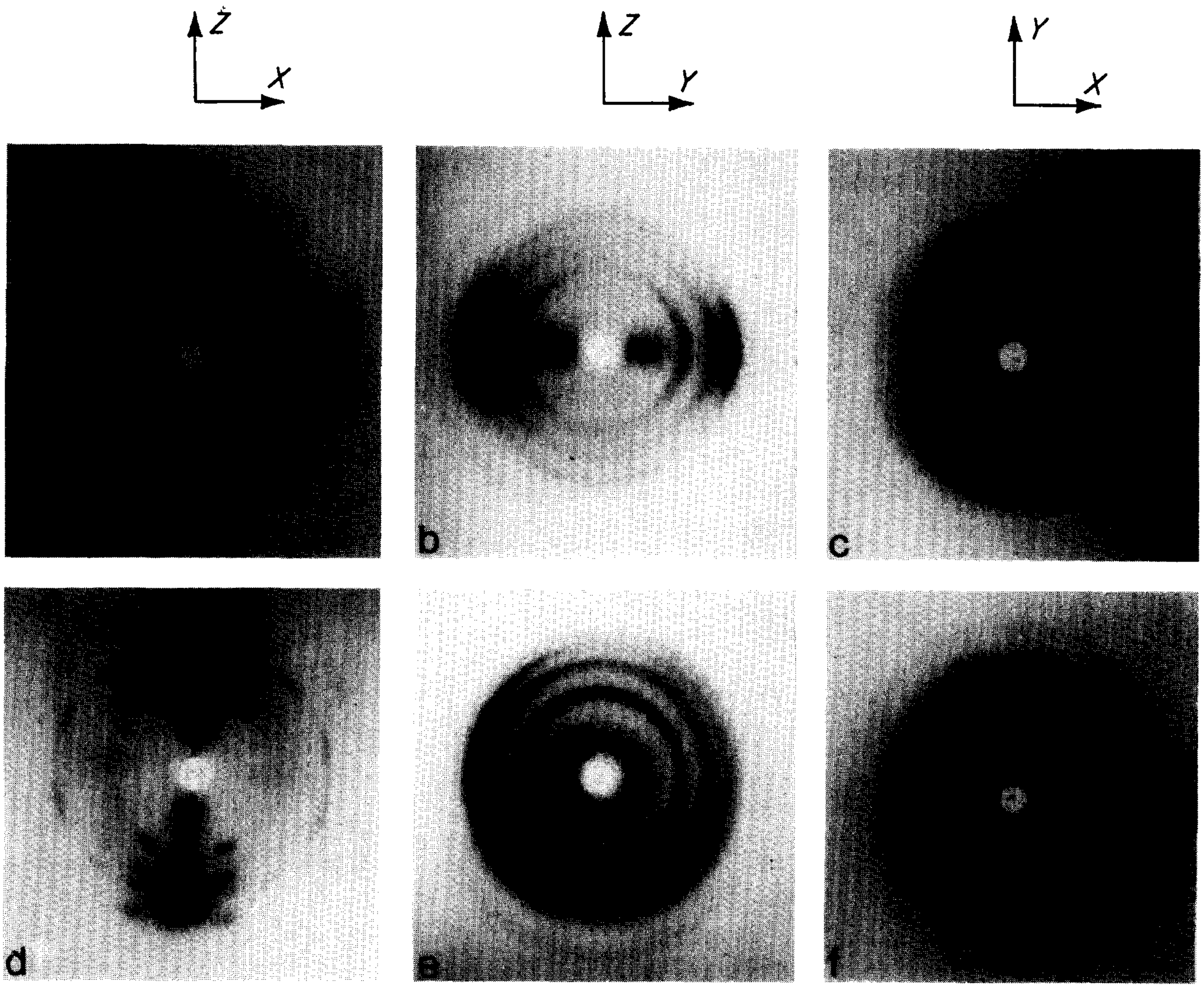

Figure 9 The WAXS patterns of both the original doubly oriented and the restretched PET samples along the axes designated respectively. The samples were doubly oriented at 200°C and restretched at 125°C. *Figures a–c,* original doubly oriented specimen; *Figures d–f,* restretched specimen

Table 1 Birefringence values of the originally doubly oriented and the pet sample restretched 4.5 times at 125°C

	$n_z - n_x$	$n_z - n_y$
Doubly oriented	0.139	0.157
Restretched	–0.170	–

mistake these apparent slip traces as evidence of crystalline or lamellar slip as mentioned before because fine slip traces are commonly observed in deformed glassy polymers[26].

For both nylon 66 and PET materials, the WAXS results indicate that the *c*-axis in the crystallites of the restretched portion is at 90°C to that in the original doubly-oriented material. Furthermore, the double orientation characteristics persisted to some extent during the restretch process. The [100] and [010] planes for PET and nylon 66 respectively, remained in the X–Z plane after being restretched. Based on the facts above, a simplified deformation scheme of crystallite rotation about the Y-axis might suggest itself. However, the long spacing along the orientation direction measured with SAXS changed with the deformation temperature in a trend similar to that observed in the necking draw of polymers[18]. Therefore, the deformation mechanism involved in the restretching of doubly-oriented PET or nylon 66 is obviously more complicated than simply the rotation of the crystallites by 90°. Such crystallite rotation is believed to be inhibited by the molecular interconnection of the crystallites and the adjacent amorphous regions.

A mechanism involving partial melting and subsequent recrystallization cannot accommodate all the experimental results either. The fact that double orientation persisted after restretching, and the finding of zero strain along Y-axis upon restretching, tends to preclude the partially melting mechanism.

The birefringence of the PET sample, restretched at 125°C to 4.5 times, was compared with the birefringence value of drawn fibre stretched from an isotropic material. The deformation ratio λ is defined with respect to the original isotropic material along the three principal axes X, Y, and Z. For the restretched sample, λ has the values of $\lambda_x = 4.5$, $\lambda_y = 0.18$ and $\lambda_z = 1.23$; whereas for the doubly-oriented samples, λ has the values of $\lambda_x = 1.0$, $\lambda_y = 0.2$ and $\lambda_z = 5.0$.

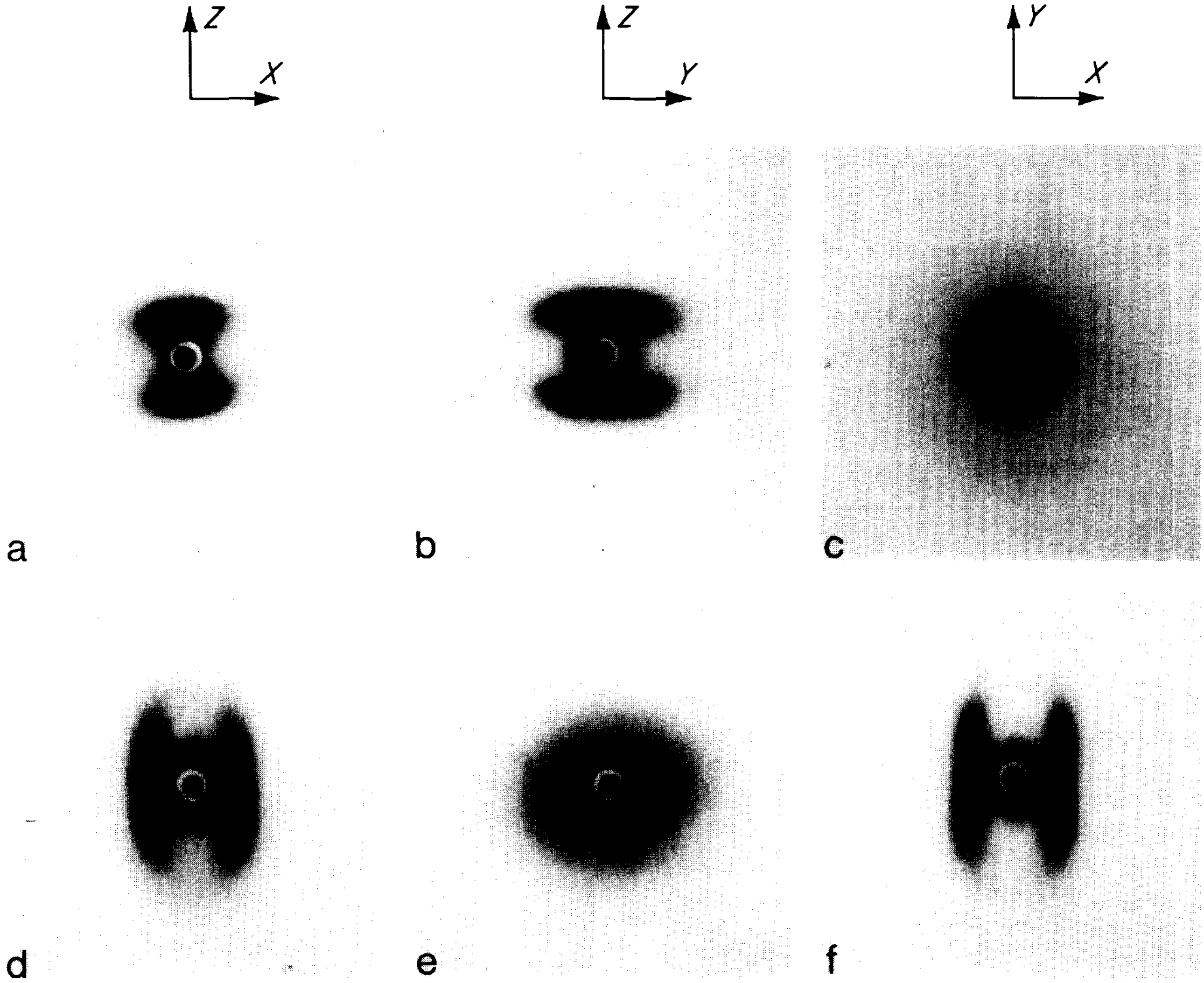

Figure 10 The SAXS patterns of both the original doubly oriented and the restretched PET samples along the axes designated respectively. The samples were doubly oriented at 200°C and restretched at 125°C. *Figures a–c,* original doubly oriented specimen; *Figures d–f,* restretched specimen

The deformation ratio of the restretched sample with respect to the squashed (doubly-oriented) one is denoted as λ'; the λ' values are as follows: $\lambda'_x = 4.5$, $\lambda'_y = 0.9$ and $\lambda'_z = 0.247$.

The measured birefringence value $n_{x,z} = 0.170$ of the restretched sample (*Table* 1) is much higher than that reported in the literature for doubly-oriented PET film of an equivalent set of deformation ratios[2]. A value of $n_{x,z} = 0.128$ was reported for doubly-oriented PET film with 5.0 stretch ratio. This $n_{x,z}$ value of 0.170 for the restretched sample demonstrates that the PET molecules in both the crystalline and amorphous phase are reasonably well aligned with the restretch direction. The high degree of crystalline orientation is quite apparent from the WAXS patterns (*Figure 9d*).

The birefringence value of 0.139 for the original doubly oriented PET sample generated by squashing at 200°C to the deformation ratios of $\lambda_x = 1.0$, $\lambda_y = 0.2$ and $\lambda_z = 5.0$, indicates that the squashed material prepared in this work had been oriented to the same extent as a PET film stretched 4.0 times[2]. The value of the birefringence of stretched PET film observed by other authors[2] was found to be consistent with a pseudo-affine deformation model. However, the birefringence result obtained in this work on the restretched sample seems to indicate that a simple pseudo-affine deformation model is inadequate to predict the birefringence values of the current deformation process accompanying restretching doubly-oriented PET if the final resultant deformation ratio of the restretch samples is used to calculate the birefringence value.

The degree of perfection of the alignment of [100] planes on the X–Z plane in the restretched PET samples seems to be less than that of the material of equivalent deformation ratio prepared directly from isotropic starting material. It is believed that this lower perfection of the alignment of the [100] planes is the main cause of the high birefringence value, $n_{x,z}$. For more detailed information regarding the birefringence values of doubly oriented PET materials and the intrinsic polarizability of PET molecules see Reference 2.

The decrease in the degree of double orientation during the restretching process can be rationalized by the fact that it is the sample dimension along Z-axis that decreases during restretching, as shown by the fact that the [100] plane of PET and [010] planes of nylon

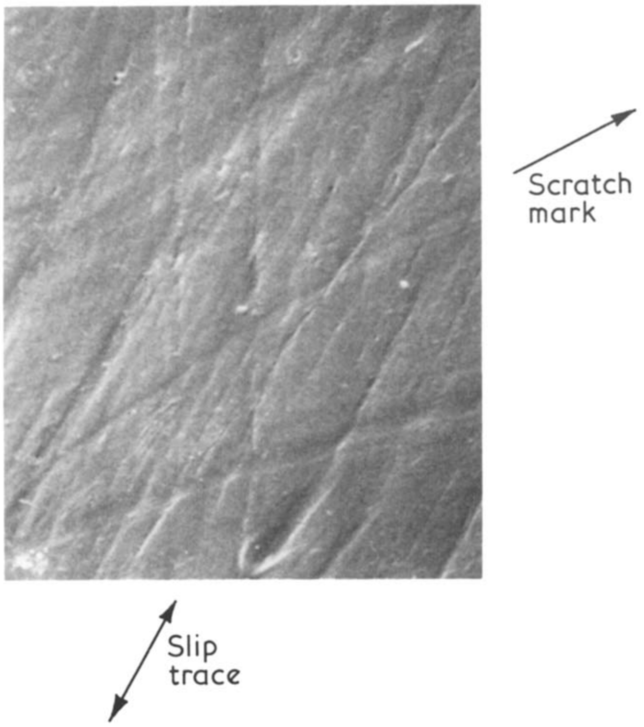

Figure 11 Scanning electron micrograph of the slip traces observed in the shoulder region of a restretched PET sample. The picture was taken on the X–Z plane; the restretching temperature was 125°C

66 are no longer well aligned in the X–Z plane. It is conceivable that the [100] and [010] planes for PET and nylon 66 respectively tends to align with the X–Y plane instead of the X–Z plane as in the case of stretching an isotropic PET film. The thickness direction of the originally isotropic film, instead of the other lateral dimension, will decrease with stretch, and concurrently the [100] and [010] planes within PET and nylon 66 samples respectively tend to align themselves with the film surface.

CONCLUSION

Doubly oriented PET and nylon 66 can only be stretched 4 or 5 times even when the tensile stress is applied in a direction perpendicular to the original molecular axis on samples maintained at elevated temperatures. This result suggests that a massive unfolding of the molecules in the crystallites within the doubly oriented samples does not occur. The presence of intense SAXS spots in both restretched PET and nylon 66 samples also indicated the lamellar nature of the crystallites within these restretched samples. The absence of the massive chain unfolding during restretching PET and nylon 66 samples may be a simple consequence of the strong lateral bonding between the molecules in the crystalline phase. The lamellar structure within these doubly oriented materials might be of a random switch board type, and such a crystalline structure would certainly preclude massive chain unfolding.

The anisotropic lateral contraction accompanying the restretching of both doubly oriented PET and nylon 66 specimens can be attributed to the absence of certain slip systems. Any slip systems having a component in [100] direction for PET and [010] direction for nylon 66 must be precluded as a result of this work.

The double orientation was preserved during the restretching process for both PET and nylon 66, whereas the SAXS long periodicity was found to change with test temperature. The above experimental facts suggest neither melting then recrystallization, nor crystallite slip and rotation to represent a complete picture of plastic deformation process in polymers, at least in PET and nylon 66.

In contrast to the lateral restretch of doubly oriented PET and nylon 66 samples, the lateral stretching of doubly oriented PE resulted in fracture of the specimens at quite low strain. A weak lateral bonding among the fibrils or the microfibrils is believed to be the cause of the early failure.

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