Deformation of High-Density Polyethylene Produced by Rolling with Side Constraints. II. Mechanical Properties of Oriented Bars

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ABSTRACT: The mechanical behavior of oriented bars of polyethylene obtained by heavy rolling with side constraints was studied. The bars were rolled to a permanent deformation ratio (DR) of up to 8.3. The elastic modulus and ultimate strength measured along the rolling direction increase with increasing DR. For samples deformed to DR = 8.3, the ultimate strength exceeds 180 MPa, whereas an unoriented material exhibits a strength of merely 15 MPa. Because of the highly ordered lamellar structure of the rolled material, the tensile deformation of the bars along the rolling direction is reversible to a large extent. The oriented bars of polyethylene also demonstrate very high toughness, especially in the direction of side constraints. Izod impact tests revealed that the impact strength of the oriented bars exceeds 200 kJ/m^2. The samples did not fracture on impact and showed only limited delamination in planes parallel to the rolling plane. Most of the delivered energy was consumed during specimen bending rather than fracture. In contrast to the tensile properties, there is an optimum DR around 5, for which the impact strength is the highest. Dynamical mechanical measurements showed that heavy rolling to a high strain, above DR = 6, produces not only a well-developed orientation of the crystalline component but also high orientation and transverse ordering of the amorphous phase, which leads to the anisotropy of the material properties in the loading direction/constraint direction plane, perpendicular to the rolling direction. J Appl Polym Sci 86: 1405–1412, 2002

Key words: mechanical properties; orientation; polyethylene (PE)

INTRODUCTION

In the first part of this study,[1] a new method of orienting polymeric materials is proposed. The method of rolling with side constraints imposed on the material was developed to overcome some limitations of conventional rolling and solid-state extrusion. The method of rolling with side constraints is a combination of channel-die compression and rolling.[2,3] It relies on the rolling of a polymer material inside a channel formed on the circumference of one roll with another roll having a thickness matching the width of the channel in the first roll. The side walls of the channel on the roll constitute lateral constraints as in a channel die. The other roll plays a role similar to that of the plunger. This system of rolls with a channel develops conditions close to plane-strain compression of the rolled material. That deformation mode is known to produce a well-developed single-component texture (quasi-single-crystal) of compressed materials.[4] The advantage of constrained rolling is the possibility of large strain deformation of relatively thick, wide, and infinitely long bars or other profiles in a continuous manner. The resulting profiles may have considerably large cross-section areas and superior mechanical properties comparable to those of fibers.

In the first part of this series,[1] the study of the development of the orientation of high-density polyethylene during rolling with side constraints is reported. Orientation produced by constrained rolling is compared to that obtained by plane-strain compression in a channel die. Plastic deformation proceeds in a very similar fashion in both orientation modes, and the resulting high orientations of the crystalline phases of polyethylene samples deformed by the two methods are very close to each other at comparable strains. In both cases, deformation leads to the formation of a strong single-component texture of the (100)[001] type with chain axes aligned along the rolling (flow) direction. At high rolling or compression rates, {310} twinning occurs when the material is unloaded.

In this part of the study, the mechanical properties of highly oriented bars of polyethylene produced by rolling with side constraints are presented. Because these bars can have relatively large cross sections, approximately a few square centimeters, and can be produced in unlimited lengths, they are potential engineering materials for many new applications.

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EXPERIMENTAL

Two high-density polyethylene resins were used in this study. One was a PE-80 pipe grade (Finathene HP-401, Atofina, Paris; melt-flow index = 0.51 g/10 min at 5 kG and 190°C, density = 0.95 g/cm^3, filled with less than 3 wt % carbon black), and the other was a high molecular mass linear polyethylene (DSM, Geleen, The Netherlands; melt-flow index = 0.46 g/10 min at 5 kG and 190°C, density = 0.945 g/cm^3, delivered in the form of sheets 12.5 mm thick); they are labeled throughout this study PE-1 and PE-2, respectively. The slabs, 100 mm × 12 mm and 1 m long, were machined out from either a thick-wall pipe (PE-1) or a sheet (PE-2).

The rolling of specimens was performed on the apparatus described in the previous article.[1] The rolling speed was set to 200 or 4000 mm/min (the same for both sets of rolls), and the temperature of the rolls was 25, 90, or 110°C. The polyethylene slabs were preheated to the desired temperature before the rolling. Samples were deformed by being rolled to various deformation ratios (DRs), defined as the ratios of the initial and final cross sections of the sample, from DR = 2 to DR = 8.3. For higher DRs, it was necessary to roll a slab in several subsequent passes of smaller DRs.

For reference, the samples of the same materials were compressed in a channel die described in previous articles[4] at temperatures and deformation rates comparable to those applied in the rolling experiments.

Specimens for mechanical testing were machined from bars of oriented polyethylene. Oar-shaped specimens designed for tensile testing were, in the narrow section, 40 mm long, 5 mm wide, and 2 mm thick. Wide parts on both ends of the specimens were 50 mm × 10 mm × 6 mm so that a firm hold in the tensile machine grips was ensured. Figure 1(a) illustrates the orientation of a sample with respect to the basic direction of the oriented bar. These specimens were tested in a tensile mode at room temperature with an Instron model 1014 tensile machine (Instron Corp., High Wycombe, UK). The crosshead speed of the machine was set to 2 mm/min, which corresponds to the initial deformation rate of 5%/min. An extensometer (25-mm gauge length) was used to measure sample elongation.

Dynamical mechanical properties of the oriented bars were determined by dynamic mechanical thermal analysis (DMTA) in a double-cantilever bending geometry at frequencies of 1 and 10 Hz, and the temperature was increased from −130 to 120°C at a heating rate of 2°C/min (Mk III DMTA, Rheometric Scientific, Inc., Epsom, UK). The specimens for dynamic measurements were machined from bars in the form of plates 60 mm long, 8 mm wide, and 1 mm thick.

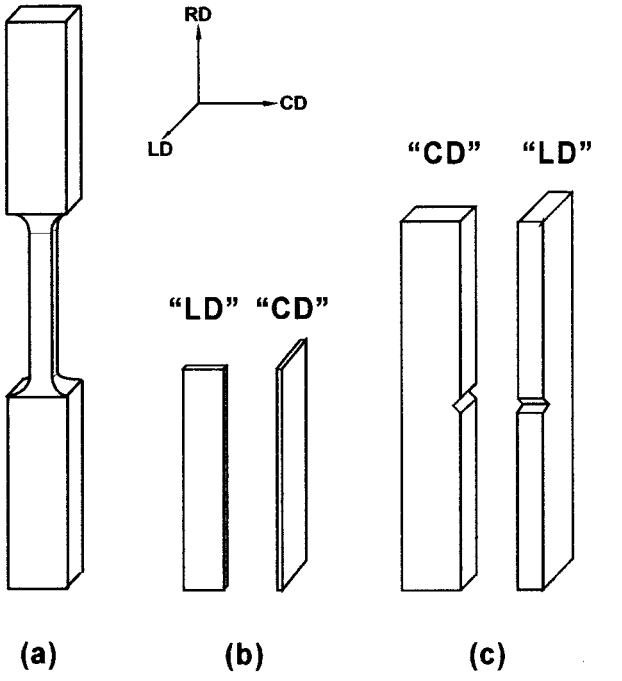

Figure 1 Geometry of the specimens used throughout this study: (a) a tensile specimen to be stretched in the RD, (b) specimens for dynamic mechanical studies to be bent in the LD and CD, and (c) notched Izod impact specimens to be loaded in the LD and CD. The reference axes of the rolled bar are RD, LD, and CD.

Samples of two types of orientation, shown in Figure 1(b), were tested. One sample was designed to bend along the loading direction (LD), whereas the other was designed to bend along the constraint direction (CD).

Impact properties of oriented bars were probed by notched Izod impact testing according ISO Standard 180/1A. The 80 mm × mm 10 × 4 mm specimens were machined from oriented bars and notched with a type A notch (radius of the notch base = 0.25 mm). Two sets of specimens of different orientations with respect to the reference directions of the bar were prepared for the determination of the impact strength of the material along the LD and CD, respectively. The geometry of the specimens is illustrated in Figure 1(c). The samples were tested at room temperature with an instrumented impact tester equipped with a hammer delivering 5.5 J of energy (Resil 5,5, CEAST S.p.A., Pianezza TO, Italy).

The melting behavior of oriented samples was determined by DSC in heating scans at a heating rate of 10°C/min (TA2100, Thermal Analysis, New Castle, DE).

RESULTS AND DISCUSSION

Melting behavior

Figure 2 presents melting data, obtained from DSC heating scans, for specimens of PE-2 rolled at 110°C.

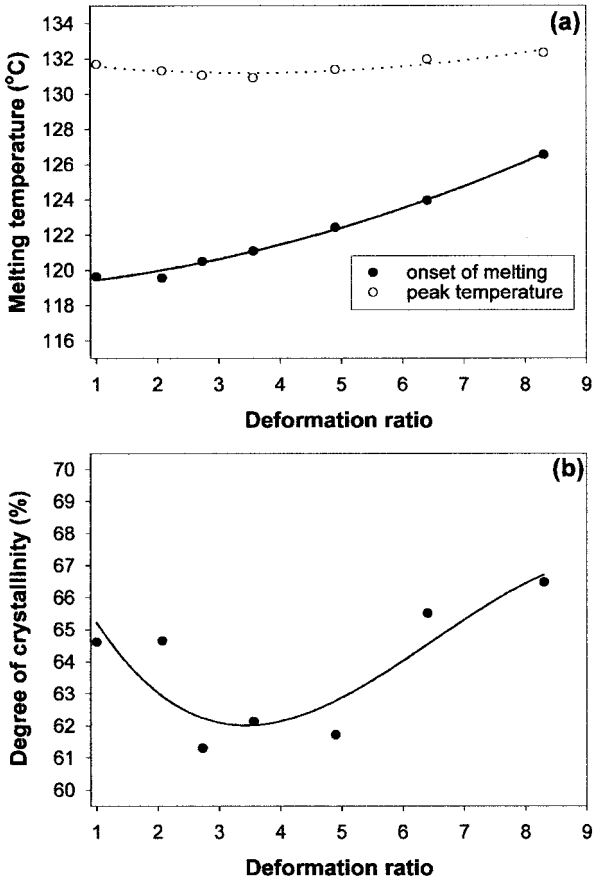

Figure 2 (a) Melting temperature and (b) degree of crystallinity as functions of DRs of rolled samples determined by DSC with heating at a rate of 10°C/min.

The reference, an undeformed specimen (DR = 1), was annealed at a temperature equal to the temperature of deformation (110°C for this series of specimens) before DSC study. The temperature of the peak of melting initially decreases slowly with increasing deformation, up to DR = 3.5, and then increases with a further increase in deformation. Finally, at DR = 8.3, the temperature of the melting peak is about 0.5°C higher than that for the undeformed specimen. However, the temperature of the onset of melting increases continuously with increasing DR. Figure 2(b) shows that the depression of the melting peak temperature at moderate strain is accompanied by a decrease in crystallinity. However, this decrease is followed by an increase in crystallinity with a further increase in deformation; this is similar to the trend observed for the melting temperature. This behavior may suggest that at moderate deformations (DR = 2.7–5), the destruction of a fraction of crystallites takes place because of the deformation process, whereas for higher deformations, this is compensated by strain-induced crystallization, which leads to an increase in overall crystallinity. The increase in the onset temperature suggests that mostly the thinnest crystallites were destroyed during the deformation process, whereas those with larger thickness were preserved, or their amount even increased because of strain-induced crystallization. This feature is similar to the behavior of high-density polyethylene deformed by plane-strain compression in a channel die to similar DRs.[4]

Mechanical properties

Figure 3 shows representative stress–strain curves of the specimens cut from bars of high-density polyethylene oriented to various DRs through rolling with side constraints (PE-2, deformed at 110°C). All the specimens were tested in tension along the direction of molecular orientation, which coincides with the rolling direction (RD). All tensile tests were performed at room temperature with a constant crosshead speed of 2 mm/min, which corresponds to the initial deformation rate of 5%/min. The elastic modulus, maximum stress (σ) and maximum strain (ϵ_{max}), and recovery data determined from the nominal stress/nominal strain curves are presented in Table I. These data show that the elastic modulus increases proportionally to the increasing DR, whereas the ultimate stress of oriented samples is approximately proportional to the second power of this ratio. The strain at break of the oriented material decreases substantially with increasing DR, especially at DR $\geq$ 4.9.

The three last columns of Table I illustrate the strain recovery data obtained for specimens deformed to the strain very close to ϵ_{max}. The elongation was stopped then (before specimen fracture), the load was released, and the sample was allowed to recover. The specimens partially recovered their strain in a few minutes after unloading. The permanent strain (ϵ_{perm}), remaining after 15 min of relaxation was measured. From ϵ_{perm} and ϵ_{max} data, the amount of strain recovery was calculated according to the formula $R = 1 - (\epsilon_{perm}/\epsilon_{max})$. The strain recovery data are presented in the last

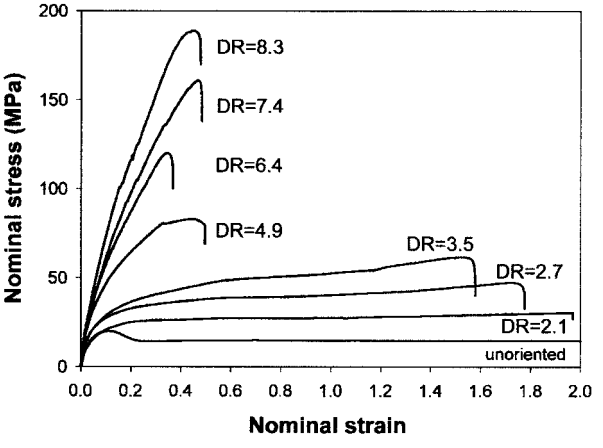

Figure 3 Representative stress–strain curves of specimens cut from bars of PE-2 oriented via rolling to various DRs.

TABLE I
Mechanical Data of Rolled PE-2 Specimens Determined in Tensile Tests with an Initial Deformation Rate of 5%/min and at Room Temperature

DR	Elastic modulus (GPa)	σ (MPa)	ε_{max}	ε_{perm}	$1 - \varepsilon_{perm}/\varepsilon_{max}$
1.0	0.80	14.9	5.00	4.80	0.04
2.1	1.47	30.5	1.97	1.58	0.20
2.7	1.89	47.4	1.78	1.46	0.18
3.5	2.09	61.6	1.58	1.19	0.25
4.9	2.45	82.9	0.50	0.16	0.68
6.4	2.50	122.0	0.37	0.12	0.67
7.4	2.62	161.0	0.48	0.13	0.74
8.3	3.05	188.7	0.48	0.11	0.79

column of Table I. Although for the unoriented material the strain recovery is relatively small, it increases very rapidly with increasing DR. The recovery is especially high in samples with DRs of 4.9 or higher. In the sample with DR = 8.3, nearly 80% of the tensile strain is recovered on specimen unloading. Such recovery behavior can be explained by the specific supermolecular structure of the rolled specimens. In the companion article,[1] it is demonstrated that high-density polyethylene samples rolled to high DRs develop well-oriented lamellar structures with chevron-type patterns (cf. Fig. 10 in ref. 1). In samples with lower molecular masses than in this study, the lamellar structure is destroyed at a DR greater than 3, and the crystalline blocks resulting from the fragmentation of lamellae reorganize spatially to give rise to the formation of a new long period.[4] For high molecular mass polyethylene, deformation by either rolling or channel-die compression is not able to induce such destruction of lamellae, probably because of the high number of tie molecules interconnecting adjacent lamellae.[1] Instead of widespread fragmentation of lamellae, the fracture occurs only locally within kink bands and leads to a highly oriented lamellar structure; second-order maxima could be observed in small-angle X-ray scattering patterns. The resulting structure, consisting of lamellae well oriented at some angle to the RD and interconnected with numerous tie molecules, when tested in tension along the RD behaves in a semielastic manner, and a large portion of the tensile strain can be recovered on unloading.

The oriented bars obtained from the other polyethylene used in this study (PE-1) demonstrate similar mechanical behavior, although the ultimate stress is slightly lower than that observed for samples of PE-2 at a comparable strain. However, the strain recovery with the unloading of bars of PE-1 is even stronger than that observed for PE-2. This is probably because of the differences in the molecular weight distributions and structures of the chains in both polyethylenes. PE-1 has a broader molecular weight distribution and has more short branches along the chain than PE-2 does. This leads to the formation of a larger number of entanglements in the amorphous phase and tie molecules interconnecting adjacent lamellae. As a result, highly oriented PE-1 behaves in a more quasielastic manner than PE-2.

The annealing of oriented bars at a constant length at a temperature above that of deformation leads to an increase in crystallinity and, consequently, to an increase in the elastic modulus and ultimate strength to a value greater than 200 MPa. The strain to break decreases slightly after annealing, whereas their quasielastic behavior (the recoverable part of the strain) becomes more pronounced; for examples, two specimens of PE-2, both rolled to DR = 6.4, one not annealed and the second annealed at 120°C after rolling, if strained to approximately 35% in a tensile mode, demonstrate ϵ_{perm} after unloadings of 12 and 6–7%, respectively.

The tensile properties observed for the rolled high-density polyethylene were worse than those for the oriented high-density polyethylene obtained by plane-strain compression in a channel die. The samples of the compressed material demonstrated slightly higher modulus and ultimate stress, whereas the strain to break was lower than in rolled samples deformed to a similar strain. As reported in the companion article,[1] the orientations of the materials produced by both methods are very similar, so the large differences in the mechanical properties were not expected. However, the samples of channel-die-compressed material were of a limited size; therefore, the tensile specimens cut from the compressed material had to be substantially smaller than those machined from the rolled bars. The differences in the size and geometry of channel-die and rolled specimens could be one possible source of the discrepancy in their responses in tensile tests.

Figure 4 shows DMTA spectra obtained for PE-1 samples rolled to DR = 5.6 and 7.1 at 90°C. The samples were tested in the double-cantilever bending mode at a frequency of 1 Hz. Two sets of specimens to be loaded dynamically along the LD and CD, respectively, were prepared and then studied. The curves of the storage modulus (E') of samples of DR = 5.6 and

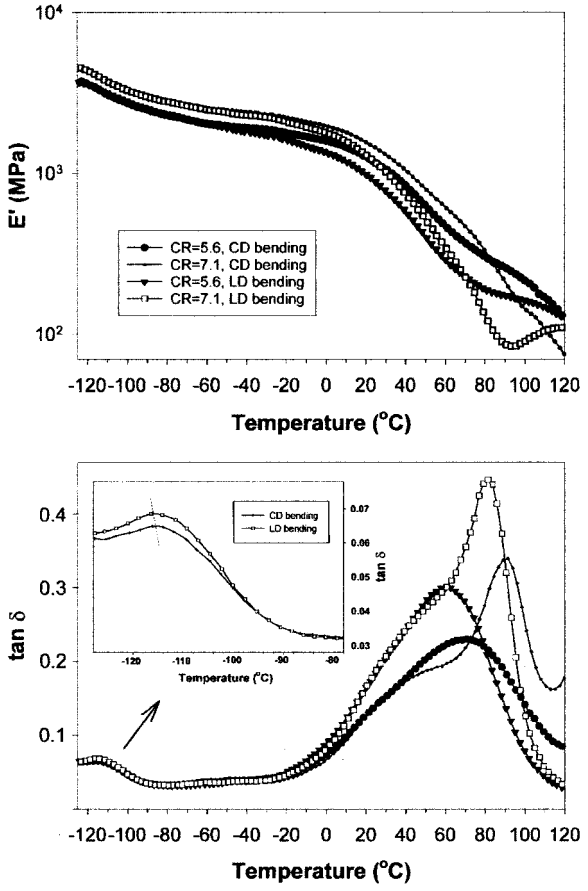

Figure 4 DMTA curves determined in a double-cantilever bending mode at a frequency of 1 Hz and at a heating rate of 2°C/min for samples of PE-1 rolled to DRs of 5.6 and 7.1: (a) E' and (b) tan δ. The inset shows an enlargement of the tan δ curve determined for a sample with DR = 7.1 probed in the CD and LD.

7.1 do not differ in shape in the low-temperature range, although the values determined for a sample with DR = 7.1 are higher than those for the sample with DR = 5.6, as expected. However, in the high-temperature range, above 70°C, the modulus of the sample with higher orientation decreases faster with temperature than E' of the sample with DR = 5.6. A minor influence of sample orientation on E' can be also noticed, especially in the high-temperature range.

Curves of the loss tangent (tan δ) demonstrate the differences in the relaxation behaviors in the specimens studied. These differences are more pronounced than for E'. In the range of γ and β relaxations (centered around −115 and −45°C, respectively), there is little difference between samples of various DRs and sample geometries. In the range of α relaxation, the behavior of the specimens tested is more diversified. For a sample with DR = 5.6, one can observe a broad maximum resulting from the convolution of two relaxation maxima. The lower one is the α relaxation itself, whereas the higher one can be assigned to the shrinkage of the oriented specimen in the RD. The shrinkage intensifies as the temperature approaches the temperature of rolling. Both maxima in the specimen tested along the LD are located at lower temperatures than the respective maxima observed in the specimen tested along the CD. Such behavior can be understand if one takes into account that both the α relaxation and the shrinkage are related to the presence of the crystalline phase in the sample. The rolled material demonstrates a quasi-single-crystal texture; therefore, the amorphous phase layers located between crystallites and intimately connected to them can also show some anisotropy in the LD–CD plane.[5] Moreover, small-angle X-ray scattering data[1] demonstrated specific spatial orientation of lamellae and interlamellar layers as well as their considerably different lateral sizes when they were probed in the FD–CD (larger) and FD–LD planes (smaller), respectively (where FD is the flow direction). These features lead to different material responses along the LD and CD, respectively. In the sample with DR = 7.1, the maxima associated with the α relaxation and the shrinkage are considerably separated from each other, and the latter is more developed than the respective maximum in the sample with DR = 5.6. Although the maxima in the sample with DR = 7.1 are shifted toward higher temperatures compared to those in the sample with DR = 5.6, the different material response in tests along the LD and CD can be still observed, as for the sample with DR = 5.6. The aforementioned behavior can be again attributed to the crystalline phase, which is even better oriented in a sample with a higher DR.

A careful examination of the γ-relaxation maxima determined for the specimens with LD and CD geometry revealed that this relaxation is also affected by the orientation process. In the sample with DR = 5.6, there is almost no difference in the shape and size (intensity) of the γ-relaxation maxima in LD and CD bending, respectively. However, some differences can be found in the sample of higher orientation, with DR = 7.1. The γ-relaxation maximum observed in LD geometry (i.e., bending along the LD) is slightly higher and shifted toward lower temperatures compared with that observed in CD geometry [see the inset in Fig. 4(b)]. The temperature shift is small, approximately 1°C, but it was observed clearly in several independent measurements, so an artifact produced by experimental error can be excluded. The γ relaxation is a relaxation of chain segments in the amorphous state and is not influenced by the texture of the crystalline component. It implies that the differences observed result exclusively from the anisotropy of the amorphous phase; it seems to be stiffer when probed along the CD instead of the LD. In our earlier studies,[5] we demonstrated that plane-strain compression of polyethylene in a channel die leads at high strain to ordering of the amorphous phase with the formation of a pseudohexagonal packing, in which the chain axes are oriented

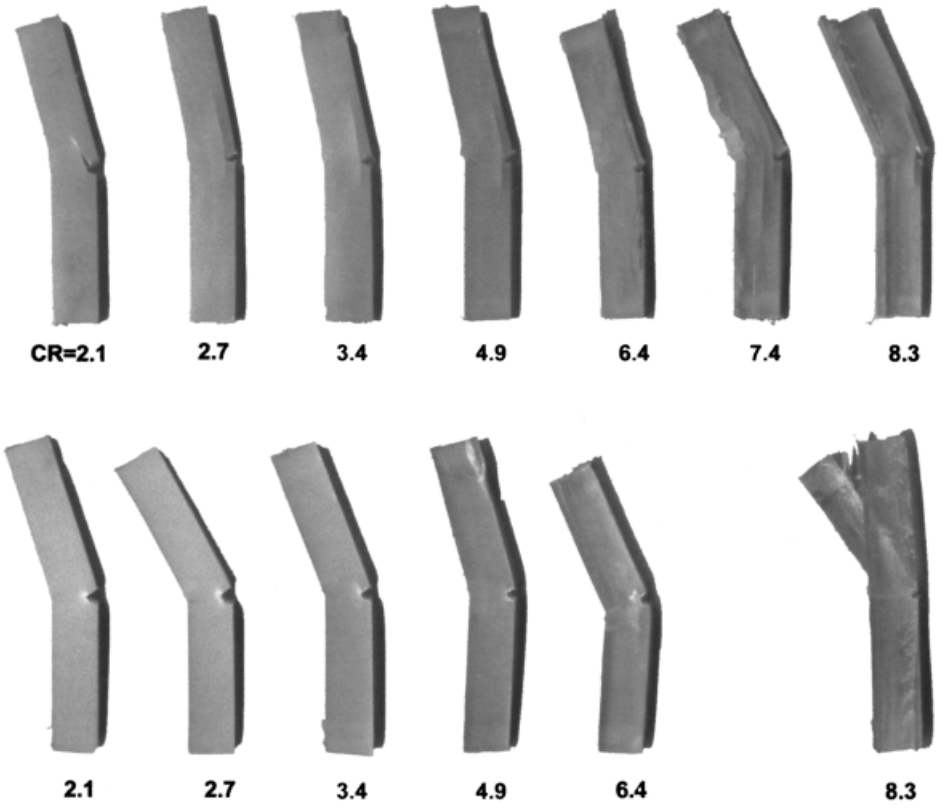

Figure 5 Photographs of Izod specimens after impact testing. The upper row shows specimens tested along the LD, and the lower row shows those tested along the CD. The DRs of the samples are indicated.

along the FD, whereas the (100) pseudoplane is oriented parallel to the LD. Most likely, such an anisotropic structure generates an anisotropy of the mechanical response of the sample in the LD–CD plane. In the companion article,[1] it is shown that both channel-die compression and rolling with side constraints realize the same deformation mode, a plane-strain compression. The resulting structure of the deformed material is the same,[1] so that the amorphous phase is oriented on rolling in a way similar to that in channel-die compression. It must also be noted that the DMTA studies of the specimens of polyethylene deformed in a channel die demonstrated a nearly identical anisotropic mechanical response of an amorphous phase in the range of γ relaxation.

The toughness of oriented bars of polyethylene was probed with notched Izod impact testing on specimens machined from bars with various DRs. Izod tests were performed at room temperature for two types of specimens with two different orientations with respect to the principal directions of the deformation process, as shown schematically in Figure 1(c). Samples denoted LD and CD were designed to be struck by the hammer along the LD (the direction of loading on rolling) and CD (the direction of side constraints during rolling), respectively. The impact tests revealed that the samples did not fracture if struck along neither the LD nor the CD. For a strike along the LD, the specimens bent, and some delamination occurred along a single plane perpendicular to the LD [see Fig. 5(a)]. The delamination started from the tip of the notch. The plane of fracture and delamination coincide with the plane of preferred orientation of the (100) plane of the crystallites. The (100) plane of orthorhombic crystals of polyethylene is the most densely packed plane, and so the force to cleave the crystal along it should be the lowest, just as for the lowest plastic resistance (i.e., the lowest critical resolved shear stress) of the crystallographic slip operating in that plane, the (100)[001] slip system. When the specimens were struck along the CD, almost no fracture occurred, and the impact resulted in specimen bending [see Fig. 5(b)], allowing the striking hammer to swing by. In the vicinity of the notch tip, some very localized delamination in several parallel planes was observed. The delamination planes were again perpendicular to the LD, as for the LD strike. The energy delivered by the impact hammer was dissipated mainly for the bending of the specimen. The examination of the angle of bending demonstrated that a considerable fraction of the bending was reversible; this is similar to the recovery of strain in the tensile experiments.

Impact tests revealed the extremely high toughness of the rolled bars and the anisotropy of their impact behavior, which was the result of their strong quasi-single-crystal texture.[1] Figure 6 presents exemplary time–force curves obtained from the instrumented Izod impact testing of the sample of PE-2 rolled to DR = 4.9. For reference, the curve obtained for unoriented sample is also shown. The maximum of the force of the oriented sample is higher and shifted toward a longer time (higher strain) than that of the unoriented polyethylene. This reflects the fact that the impact strike generates in the oriented sample only limited fracture and delamination (cf. Fig. 5), and the energy is dissipated mostly on the deformation of the specimen by bending. The force at peak, the energy dissipated to that point, and the total dissipated energy, determined from the time–force curves of impacted oriented specimens of various DRs, are presented in Figure 7. The dissipated total energy of greater than 5 J observed for highly oriented samples corresponds to an Izod impact strength approaching a value of 200 kJ/m^2, which is nearly 15 times higher than the strength of 14 kJ/m^2 determined for an unoriented material.

The plots presented in Figure 7 show that the force, as well as the energy, is always greater for specimens tested along the CD than for specimens tested along the LD. Both LD and CD specimens tend to delaminate along planes perpendicular to the LD. In the LD specimen, the geometry is usually a single fracture plane extending from the notch toward the specimen end. However, in the CD specimens, the fracture usually develops in a few parallel planes, each perpendicular to LD, but it is limited spatially to the vicinity of the notch. Such a difference in the fracture behavior results in higher force generated on impact and higher energy dissipated at the peak by CD specimens. In the LD specimens, the effective cross section of the specimen after delamination becomes smaller than that in CD specimens. Consequently, the energy dissipated for the deformation (bending) of CD specimens is higher than that for LD specimens, and the total dissipated energy follows this trend.

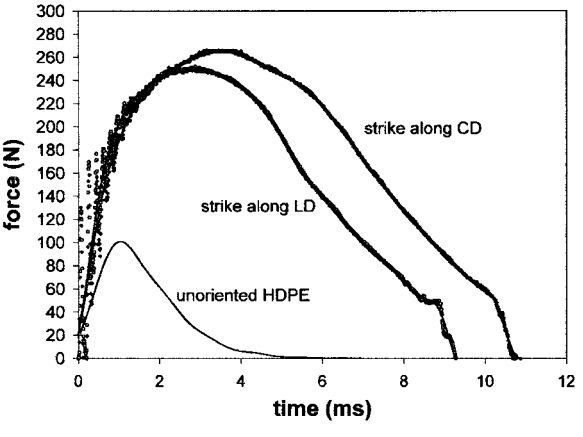

Figure 6 Time–force curves determined for PE-2 oriented samples with DR = 4.9 tested along the LD and CD in Izod impact testing. For reference, the curve obtained for isotropic, unoriented PE-2 is also shown.

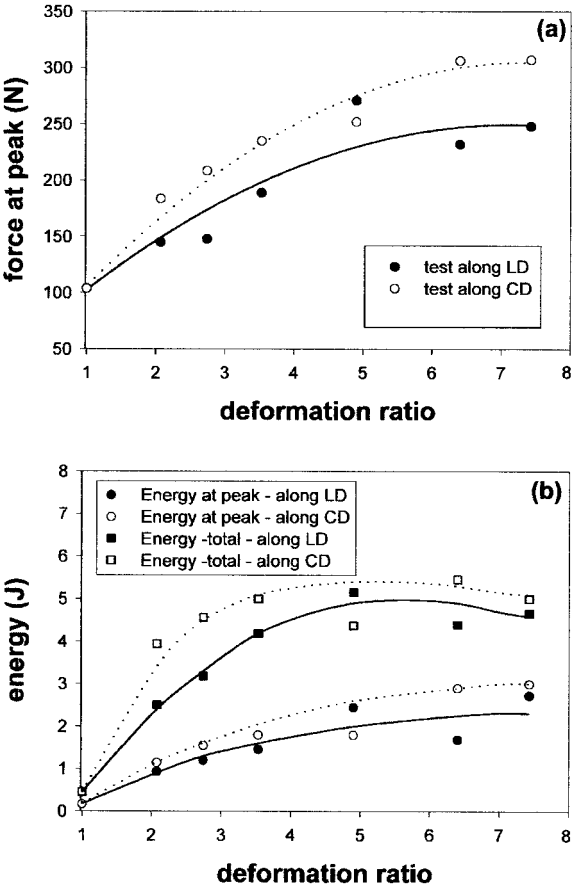

Figure 7 (a) Peak force and (b) peak energy and total energy dissipated, as determined from time–force curves obtained in Izod impact testing of samples of rolled PE-2, plotted as functions of their DRs.

The dependence of the total dissipated energy on the DR passes through a broad maximum near DR = 5. Samples with higher DRs delaminate over a larger length because of the higher perfection of the (100)[001] texture, and so the energy necessary for the plastic deformation and bending of such delaminated material is lower than that in a sample with a lower DR and smaller delamination. As a result, the total dissipated energy tends to decrease with a DR greater than 5.

The oriented bars obtained by the rolling of PE-1 demonstrate similar impact behaviors. The energy dissipated during impact is also very high, approaching the capacity limit of the instrument used in the study. However, they delaminate a bit less when impacted, and so they consume more energy during the plastic deformation stage, and their toughness is even higher than that of samples of rolled PE-2. This can be attributed to the more branched structure of PE-1 material

in comparison with that of PE-2 (the impact strength of unoriented PE-1 is also higher than that of PE-2) and to the presence of a small amount of the carbon black filler disturbing locally the perfection of the alignment of (100) planes of orthorhombic polyethylene crystallites during the rolling process.

CONCLUSIONS

We studied the mechanical behavior of oriented bars of polyethylene obtained by heavy rolling with side constraints. Polyethylene deformed according to this method had tensile properties similar to those of polyethylene deformed by other methods, such as conventional rolling or plane-strain compression. The elastic modulus and ultimate strength measured along the RD increased with increasing DR. For samples deformed to DR = 8.3, the ultimate strength reached 180 MPa and could be further increased by postdeformation annealing. Because of the highly ordered lamellar structure of the rolled material, the tensile deformation of the bars along the RD was reversible to a large extent.

Along with good tensile properties, oriented bars of polyethylene demonstrated very high toughness, especially in the CD. The Izod impact tests demonstrated that the samples did not fracture on impact and showed only limited delamination in planes parallel to the rolling plane. Most of the delivered energy was consumed by specimen bending instead of fracture. The impact strength of the oriented bars exceeded 200 kJ/m^2. In contrast to the tensile properties, there was an optimum DR around 5, for which the impact strength was the highest. For higher orientations, more delamination developed during impact, bending was then easier, and the total energy absorbed by the material consequently was slightly lower than for a sample with DR = 5.

Dynamical mechanical measurements showed that the heavy rolling to high strain (DR $>$ 6–7) produced not only a well-developed orientation of the crystalline component but also high orientation and transverse ordering of the amorphous phase, which led to the anisotropy of their properties in the LD–CD plane, perpendicular to the RD.

Orientation by rolling with side constraints allows for the continuous production of oriented materials of unlimited length, similarly to conventional rolling yet with relatively large cross sections (>1 cm^2). Because of their size and very good mechanical properties, the rolled bars or profiles produced from commodity polymers such as polyethylene may become very attractive engineering materials.

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