

Insight into the Excellent Tribological Performance of Highly Oriented Poly(phenylene sulfide)

Liang-Qing Zhang^{a,†}, Shu-Gui Yang^{b,†}, Jing-Han Zhang^c, Ke-Peng Zhong^c, Zhong-Guo Zhao^d, Yan-Hui Chen^{c*}, Jun Lei^e, Qiu-Yu Zhang^c, and Zhong-Ming Li^e

^a College of Material Science and Engineering, Xi'an University of Science and Technology, Xi'an 710054, China

^b State Key Laboratory for Mechanical Behaviour of Materials, Shaanxi International Research Center for Soft Materials, Xi'an Jiaotong University, Xi'an 710049, China

^c School of Chemistry and Chemical Engineering, Shaanxi Key Laboratory of Macromolecular Science and Technology, Key Laboratory of Special Functional and Smart Polymer Materials of Ministry of Industry and Information Technology, Northwestern Polytechnical University, Xi'an 710072, China

^d National and Local Engineering Laboratory for Slag Comprehensive Utilization and Environment Technology, School of Materials Science and Engineering, Shaanxi University of Technology, Hanzhong 723000, China

^e College of Polymer Science and Engineering, State Key Laboratory of Polymer Materials Engineering, Sichuan University, Chengdu 610065, China

Electronic Supplementary Information

Abstract Achieving low friction and wear of poly(phenylene sulfide) (PPS) without using fillers or blending is a challenging task, but one of considerable practical importance. Here we describe how neat PPS with high tribological performance is achieved by manipulating processing parameters (pressure, flow and temperature). The key to achieving high tribological performance is comparatively high molecular chain orientation, realized in neat PPS, at high shear rates and low pressure. The friction coefficient and wear rate are as low as ~ 0.3 and $\sim 10^{-6} \text{ mm}^3 \cdot \text{N}^{-1} \cdot \text{m}^{-1}$, respectively, which break the record for neat PPS. These values are even better than those for PPS-based blends and comparable to PPS composites. Further studies show, for the first time, that wear rate decreases exponentially with increasing molecular chain orientation, prompting us to revise the classical Archard's law by including the effect of molecular chain orientation. These findings open the possibility of using neat PPS in highly demanding tribological applications.

Keywords Poly(phenylene sulfide); Orientation; Wear rate; Friction coefficient; Processing parameters

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INTRODUCTION

Poly(phenylene sulfide) (PPS), one of the most promising engineering thermoplastics to replace metallic materials, is finding increasing use in aircraft components, automotive industry and military applications because of its low weight, low cost and excellent corrosion resistance.^[1–3] However, due to the high friction coefficient (~ 0.5) and poor wear resistance (10^{-5} – $10^{-4} \text{ mm}^3 \cdot \text{N}^{-1} \cdot \text{m}^{-1}$), PPS has not yet fulfilled its application in highly demanding tribological fields, such as seals, gears, and cams.^[4,5] Many attempts have been made to improve tribological performance of PPS, and the most common solutions are adding anti-wear fillers, blending with self-lubricating polymers, or both. These methods are straightforward and only some of them have been put into practice in the past decades. Tribological performances of PPS composites

and blends, *i.e.* friction coefficient and wear rate, are greatly influenced by the PPS matrix morphology, the type and dispersion of fillers, and by interface properties.^[4–10] Among these factors, the intrinsic tribological performance of neat PPS, as the weak link, has not been fully explored.

Polymer materials with the high molecular chain orientation usually have high tensile strength and modulus, low coefficient of thermal expansion and even excellent toughness, in spite of lacking reinforcing fillers.^[11–14] They can be obtained by optimizing the processing parameters (tensile force, pressure, shear rate or temperature). Additionally, the homogeneous polymer system produced without fillers is more easily recycled and helps achieve sustainability and protects the environment. Furthermore, molecular dynamics simulation studies have predicted that a high degree of molecular chain orientation parallel to the sliding plane would result in a low friction coefficient and wear rate.^[15,16] Thus, theoretically, without the help of anti-wear fillers or self-lubricating polymers, improvement of tribological performance of neat PPS may be achieved by tuning molecular chain orientation. However, until now filler-free technology to improve wear-

* Corresponding author, E-mail: yanhuichen@nwpu.edu.cn

[†] These authors contributed equally to this work.

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resistance of neat PPS has rarely been explored, mainly due to its rigid backbone, which makes it difficult to be stretched by shear flow.^[17–23] In addition, high melting temperature (~ 280 °C) of PPS puts higher demands on processing equipment, *e.g.*, precise temperature control, tight dynamic sealing, the exact implementation of shear flow. Fortunately, pressurizing and shearing devices (PSD) that are capable of controlling processing parameters have been designed and manufactured in our group.^[24–27] In our recent work, tunable molecular chain orientation has been successfully achieved in polyolefins^[24,26] with flexible molecular chains, polylactide^[25] with semi-rigid backbones, and polycarbonate with rigid ones.^[27] Therefore, PSD technology presents its potential to be used to prepare neat PPS with modulated molecular chain orientation, and further to reveal the inner relation between molecular chain orientation and tribological performance.

In this study, a series of processing conditions for producing PPS samples, *i.e.*, pressures (5, 25, 50, and 100 MPa), shear rates ($0\text{--}30\text{ s}^{-1}$) and a fixed undercooling (~ 75 °C), were adopted by utilizing the PSD. The friction coefficient (μ) and dimensionless wear rates (W_s) of selected PPS samples were firstly studied. It demonstrated that both μ and W_s of PPS decreased with the increase of shear rate and decrease of the pressure, and thus, neat PPS with ultra-high tribological performance ($\mu=0.3$, $W_s=1.5\times 10^{-6}\text{ mm}^3\cdot\text{N}^{-1}\cdot\text{m}^{-1}$), was achieved by manipulating processing parameters (5 MPa, 25 s^{-1}). Subsequently, utilizing the synchrotron radiation X-ray scattering studies of PPS, the underlying mechanism of the improved tribological performance was revealed, attributing to comparatively high molecular chain orientation. According to the phenomenon that the wear rate decreased exponentially with the increasing molecular chain orientation, the classical Archard's law was revised by including the effect of molecular chain orientation. In the end, the relationships between tribological performance, processing parameters and molecular chain orientation were fully discussed.

EXPERIMENTAL

Materials

Commercial grade PPS with the weight-average molecular mass (M_w) and polydispersity (M_w/M_n) of $4.8\times 10^4\text{ g/mol}$ and *ca.* 2, respectively, was synthesized by the Sichuan Deyang Special New Material Co., Ltd. (China). The nominal melting point (T_m), onset crystallization (T_{oc}), crystallization (T_c) and glass transition temperature (T_g) measured by differential scanning calorimetry (DSC) are 281.2, 246.7, 236.1 and 92.6 °C, respectively (Fig. 1).

Experimental Procedure

Fig. 2(a) shows schematic of PSD, and the process to load pressure and to apply shear flow. The applied shear rates increased linearly with the radius ($\dot{\gamma} = \omega r/d$, where ω is angular velocity of the PSD rotator, r is radius of observation position, and d is the sample thickness of 0.5 mm). PPS samples crystallized at different processing parameters (pressure, shear rate and temperature) were performed as follows (Fig. 2b): (1) annealing the PPS sample at 320 °C for 5 min to erase thermal memory; (2) cooling down to the experimental temperature; (3) applying the experimental pressure and shear pulse (duration time=45 s); (4) maintaining the pressure and isothermally crystallizing for 30 min; (5) cooling down to room temperature.

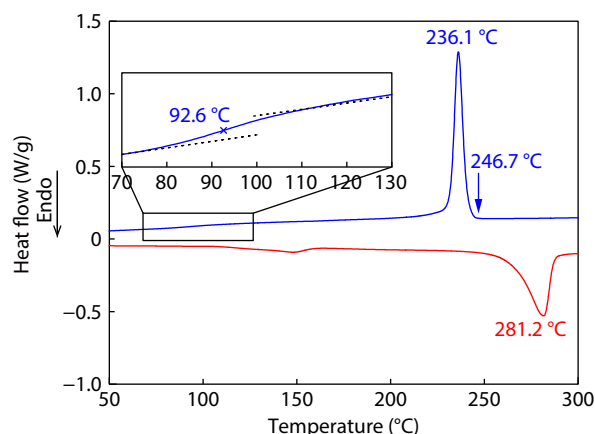


Fig. 1 First heating (red) and cooling (blue) DSC curves of PPS. The heating and cooling rates are 10 °C/min , and the enlarged view shows the T_g .

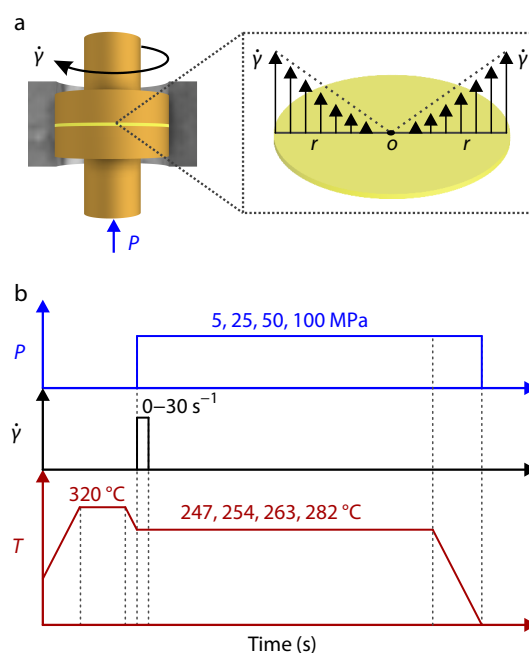


Fig. 2 (a) Schematic illustration of the application of pressure and shear flow, and the radial distribution of shear rate imposed on the yellow disk sample; (b) Experimental protocol.

Ultimately, the disk-shaped sample with a thickness of 0.5 mm and a radius of 10 mm was obtained. Four experimental pressures (5, 25, 50 and 100 MPa) were selected considering the wide pressure range in real processing. To avoid the crystallization of PPS before reaching the processing temperature, the experimental temperature of 247 °C higher than the T_{oc} was selected for 5 MPa. Meanwhile, as the equilibrium melting temperature (T_m^0) of PPS increases with the increase of pressure, different experimental temperatures, 254, 263 and 282 °C were used for 25, 50 and 100 MPa, respectively, to maintain a constant undercooling (~ 75 °C). The applied shear flow ranged from 0 s^{-1} to 30 s^{-1} with increasing the r from 0 mm to 10 mm (detailed discussion about the experimental parameters is released in section 1 of the electronic supplementary information, ESI).

Friction and Wear Tests

A pin-on-disc friction and wear apparatus (HT-1000, Lanzhou Huahui Instrument Technology Co., Ltd., China) was used to study the tribological performance of PPS samples under dry friction conditions (temperature $\sim 25^\circ\text{C}$, relative humidity $\sim 50\%$). Fig. S1 (in ESI) shows the fractional pair. The disk-shaped PPS sample (yellow) was fixed on the stage and rotated at a speed of 1000 r/min. A stainless pin with a diameter of 2 mm was slid against the surface of PPS sample at a normal load (F_N) of ~ 6.7 N. The test duration was set as 30 min. For each PPS sample, three positions, $r \approx 2.5$, 5.0, and 8.3 mm (the corresponding shear rates were 7.5, 15 and 25 s^{-1} , respectively) were tested. Before each test, the pin surface was polished with metallographic abrasive paper ($R_a = 0.09\text{--}0.11\text{ }\mu\text{m}$). Meanwhile, the surfaces of the tested samples were similar to each other as all the samples display the similar water contact angle (Fig. S2 in ESI). The friction and wear tests on each specimen were repeated for at least three times after which the average specific wear rate was determined.

The μ of PPS sample was calculated by the following equation:

$$\mu = F_R / F_N \quad (1)$$

where F_R is the friction force, and the dimensionless W_s of PPS sample was calculated by using the equation:

$$W_s = V / (F_N \times L) \quad (2)$$

where L is the sliding distance (m), V is the wear volume (mm^3). To get V , the cross-section area (A) of the worn track was measured by using a surface profilometer, and then A was multiplied by the perimeter of worn track.

Optical Microscope (OM) Observations

An Olympus BX51 microscopy (Olympus Co., Tokyo, Japan) equipped with an Olympus DP27 camera was employed for the observation of wear surface. OM images were recorded in reflection mode using a Halogen light source.

Wide-angle and Small-angle X-ray Scattering (WAXS and SAXS) Measurements

The crystalline structure of PPS crystallized under different shear rates and pressures was characterized by WAXS and SAXS measurements. The X-ray beam with a wave-length of 0.124 nm was perpendicular to the surface of disk-shaped PPS sample, and scanned from inner to exterior along the radial direction (Fig. S3 in ESI). Various radial positions corresponding to the shear rates of 5, 10, 15, 20, 25, and 30 s^{-1} , respectively, were scanned to obtain both WAXS and SAXS patterns, which were carried out at the beamlines BL15U1 and BL16B1 of Shanghai Synchrotron Radiation Facility (SSRF, Shanghai, China), respectively. The sample-to-detector distances were 150.8 and 1841 mm for WAXS and SAXS measurements, respectively.

WAXS/SAXS patterns were background corrected and circularly averaged at a constant q , resulting in WAXS and SAXS profiles.^[28,29] Meanwhile, because molecular chain axis in the orthorhombic phase of PPS is in parallel to c -axis of the unit cell, the azimuthal intensity distribution functions $I(\varphi)$ along the (200) diffraction ring can be used to estimate the crystalline chain orientational order parameter f_H . By use of the Polanyi equation:^[30]

$$\cos\Theta = \cos\theta_{hkl}\cos\varphi \quad (3)$$

where θ_{hkl} is the Bragg scattering angle, one can correlate the azimuthal angle φ along the Debye circle with the angle Θ between the normal of (200) crystalline planes and the reference direction (flow direction). Then the degree of orientation is estimated by the Hermans' equation:^[31,32]

$$f_H = \frac{3 \langle \cos^2\Theta \rangle - 1}{2} \quad (4)$$

The $\langle \cos^2\Theta \rangle$ is defined as

$$\langle \cos^2\Theta \rangle = \frac{\int_0^{\pi/2} I(\Theta) \cos^2\Theta \sin\Theta d\Theta}{\int_0^{\pi/2} I(\Theta) \sin\Theta d\Theta} \quad (5)$$

where $I(\Theta)$ is the integrated intensity along with the angle Θ . The value of -0.5 indicates that the crystal is perfectly oriented with its c -axis parallel to the flow direction, and 0 means no preferred orientation.

RESULTS

Friction and Wear Properties

The tribological properties of PPS crystallized under various pressures (5, 25, 50, and 100 MPa), shear rates ($0\text{--}30\text{ s}^{-1}$) and a fixed undercooling ($\sim 75^\circ\text{C}$) were studied using the pin-on-disc friction and wear apparatus. Figs. 3(a)–3(d) show the evolution of μ with time. Basically, two stages are observed: (a) μ increases with time in the early stage of sliding as the surfaces try to accommodate the applied wear; (b) After the running-in period, μ tends to remain stable due to the formation of PPS transfer film on the pin.^[33–35] We found that all PPS samples crystallized at 25 s^{-1} and at a given pressure exhibit a lower equilibrium μ than those at 7.5 s^{-1} . Meanwhile, the decrease in pressure also decreases μ . The lowest μ of neat PPS measured in this work is ~ 0.31 (5 MPa, 25 s^{-1}), which is significantly lower than the common value of ~ 0.5 for neat PPS. One may note that μ fluctuates significantly even after the running-in period. This might be related to an unstable transfer film behavior (transfer film buildups and breakdowns).^[36,37]

Figs. 3(e₁)–3(e₈) show OM images of the wear surfaces. In general, groove marks, along with surface lumps and detached areas on the wear surface are clearly visible, suggesting that the dominant wear mechanism are abrasion by surface asperities on the pin, and adhesion in the contact area.^[7,10,33] It is evident that PPS samples crystallized at lower pressures and higher shear rates show much smaller areas with removed surface, indicative of a highly improved wear resistance. It is also exciting to note that the wear surface of PPS crystallized at 5 MPa, 25 s^{-1} is rather smooth, and only thin groove marks can be noticed, indicating its low W_s .

In the next step, W_s of the tested samples have been calculated (Table S1 in ESI) and compared with those of other PPS systems. As shown in Fig. 4, W_s of neat PPS in this work ranges from $1.5 \times 10^{-6}\text{ mm}^3\text{N}^{-1}\text{m}^{-1}$ to $2.5 \times 10^{-5}\text{ mm}^3\text{N}^{-1}\text{m}^{-1}$. The higher value of W_s is close to that of neat PPS reported in the literature,^[33,36] corresponding to PPS crystallized at 7.5 s^{-1} at different pressures (upper four spheres). As the shear rates increase to 15 s^{-1} (middle four spheres) and 25 s^{-1} (bottom four spheres), the W_s of PPS is reduced substantially. An ultra-low value of W_s ($1.5 \times 10^{-6}\text{ mm}^3\text{N}^{-1}\text{m}^{-1}$) is achieved for neat PPS crystallized at 25 s^{-1} and 5 MPa, which breaks the record of W_s

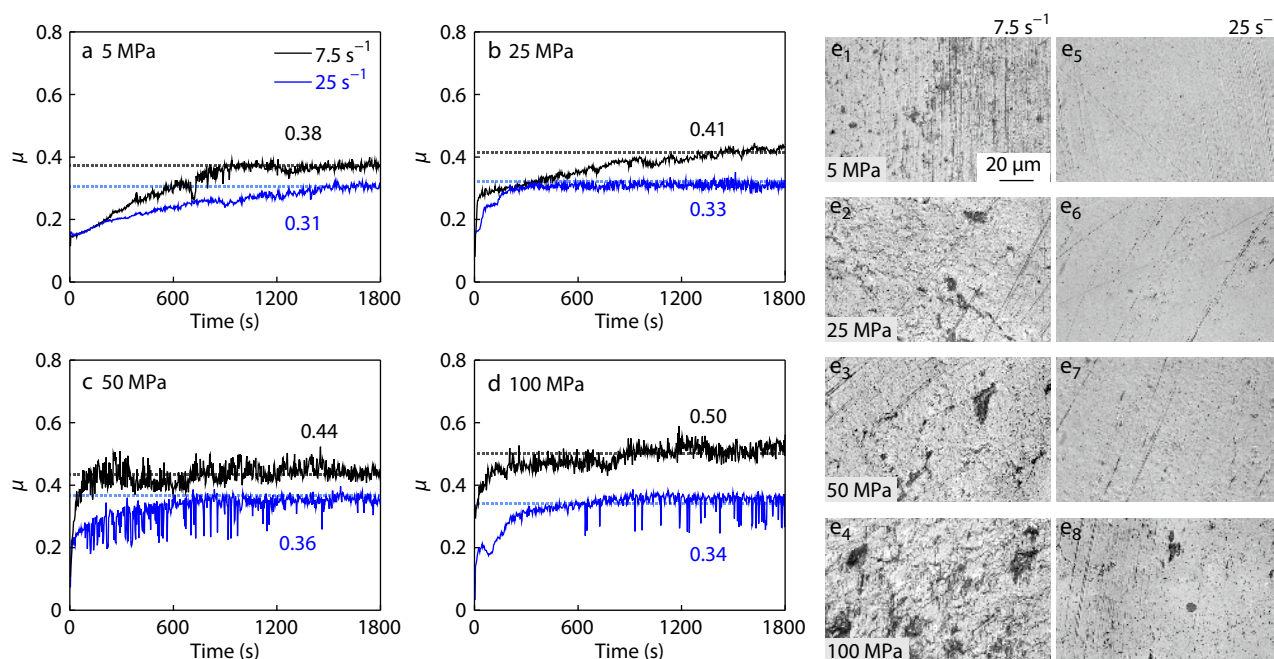


Fig. 3 Time-dependence of μ of neat PPS samples crystallized under different pressures and shear rates (a–d), and the corresponding OM images of worn surfaces after the test (e₁–e₈).

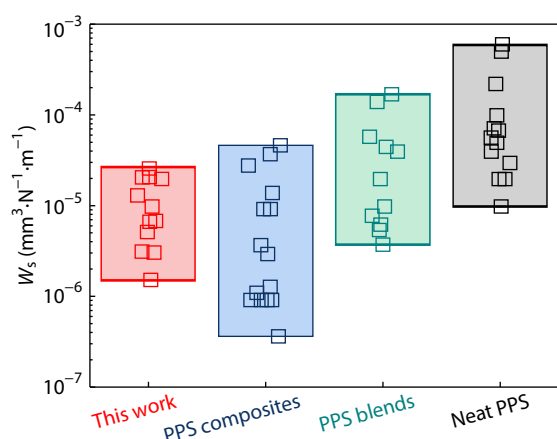


Fig. 4 Comparison of the dimensionless wear rate of neat PPS in this work with other PPS systems, i.e., PPS composites,^[5,36,37,41] PPS-based blends,^[7–9,37,39,40] and neat PPS^[5,7,36–38] reported in the literature. All the data cited from literature are listed in ESI (Table S2).

for neat PPS in the literature.^[5,7,36–38] Overall, the wear resistance of neat PPS in our work is over 10 times better than that of common PPS, even better than those of PPS-based blends (3.8×10^{-6} – 1.7×10^{-4} mm³·N⁻¹·m⁻¹)^[7–9,37,39,40] and is comparable to those of PPS composites (4.0×10^{-7} – 5.0×10^{-5} mm³·N⁻¹·m⁻¹).^[5,36,37,41] Therefore, the unprecedented tribological properties, i.e., ultra-low μ and W_s , have been realized in neat PPS system, by controlling the processing parameters in our current work.

Crystalline Structure and Molecular Chain Orientation

In order to reveal the influence of the processing parameters, we examined the crystalline structure of neat PPS crystallized under different shear rates and pressures in detail by WAXS and SAXS measurements. As shown in Fig. 5, all the WAXS patterns

display the characteristic diffractions of the orthorhombic phase of PPS (see the linear WAXS profiles in Fig. S4 in ESI), regardless of pressure and shear rate.^[42] At 5 s⁻¹, a sharp diffraction ring corresponding to the (102)/(200)/(111) crystallographic planes can be observed in WAXS patterns for different pressures, showing no preferred orientation. Similarly, SAXS patterns are also isotropic. This is understandable since the molecular relaxation time (i.e., the reptation time τ_d) for PPS crystallized at 5, 25, 50 and 100 MPa are 0.093, 0.105, 0.122 and 0.166 s (Detailed explanation is presented in section 1 in ESI). Thus, a shear rate of 5 s⁻¹ is not high enough to align the rigid molecular chains. When the shear rate is increased (≥ 10 s⁻¹), the WAXS ring corresponding to $d=4.34$ Å, dominated by the (200) component, coalesces near the equator under all pressures, and most strongly at the lowest pressure. This implies that the crystallographic *c*-axis, i.e., the chain axis, is parallel to the flow direction.^[42] Consistent with the WAXS results, with increasing shear rate, SAXS long period intensity, which comes from stacked lamellae, also coalesces near the meridian^[43–45] and remains at almost the same scattering vector (q) (see the linear SAXS profiles in Fig. S5 in ESI). The crystallinity and lamellar thickness of PPS samples crystallized at different pressures and shear rates were thus calculated and released in the Supporting Information (Fig. S6 in ESI). It shows that all of these PPS samples have similar crystallinity and lamellar thickness.

Apparently, the WAXS and SAXS patterns suggest that the most obvious change caused by the variations of pressure and shear rate is the crystal orientation (i.e., molecular chain orientation for PPS). As shown in Figs. 6(a)–6(d), the azimuthal intensities of (102)/(200)/(111) diffractions are evenly distributed for different pressures when the shear rate is 5 s⁻¹. While the shear rate reaches 10 s⁻¹, broad peaks at 90° and 270° are observed. Further increasing the shear rate, the peaks at 90° and 270° become sharper, and additional peaks

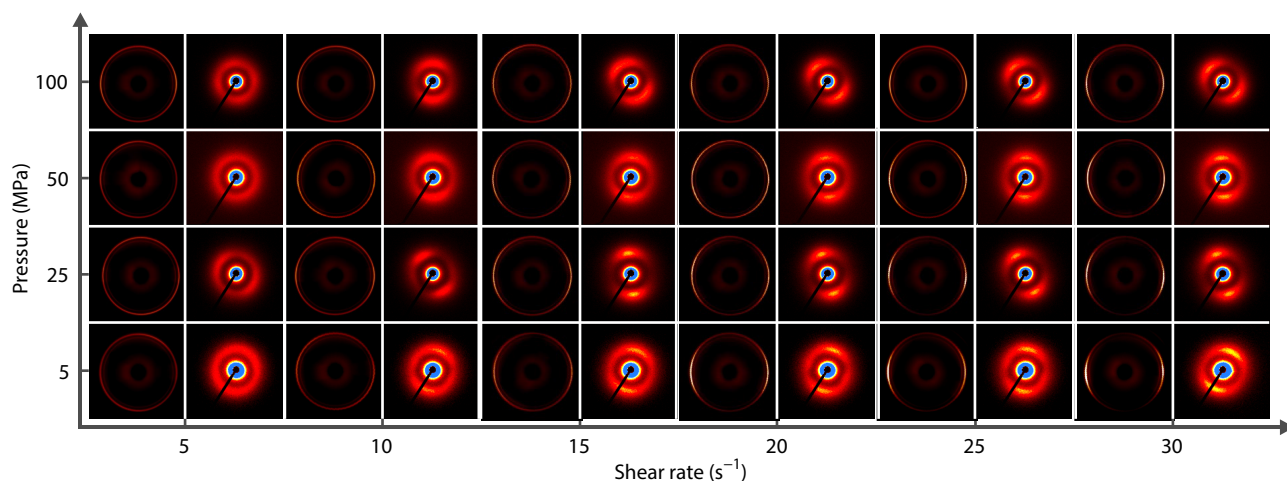


Fig. 5 WAXS and SAXS patterns of PPS processed under different shear rates and pressures. The flow direction is vertical.

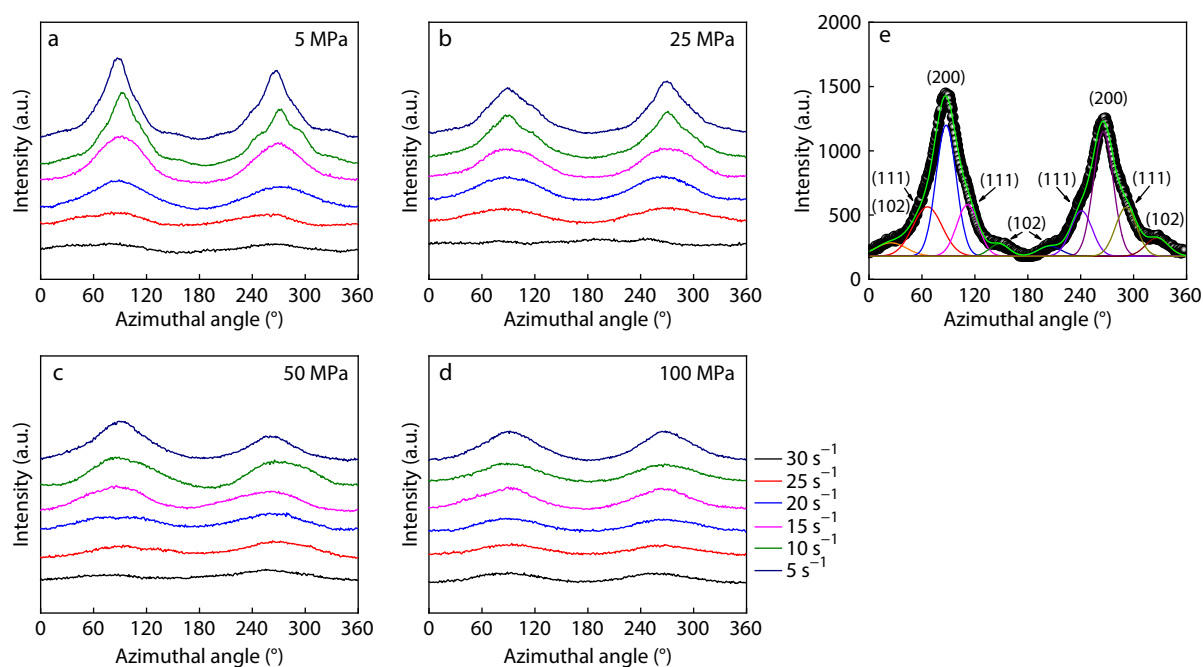


Fig. 6 (a–d) Azimuthal distributions of (102)/(200)/(111) diffractions of PPS crystallized under different pressures and shear rates. The profiles at different shear rates have been shifted by the identical value along y-axis, and the y-axes for different pressures are in the same range. (e) The azimuthal profiles of (102)/(200)/(111) diffractions of oriented crystalline structure can be fitted by ten Gaussian peaks that are related to (102), (200) and (111) reflections, respectively.

at approximately 30° , 60° , 120° , 150° , 210° , 240° , 300° , 330° can be noticed. Based on the orthorhombic unit cell with $a=8.67$ Å, $b=5.61$ Å and $c=10.26$ Å,^[42] the diffraction peaks at 90° and 270° should be assigned to (200) reflections, and the diffraction peaks at 30° , 150° , 210° and 330° are related to the (102) reflections, and the remaining peaks are corresponding to (111) reflection. Using a deconvolution procedure, the full spectrum is well fitted by a number of reflections originating from (200), (102) and (111) (Fig. 6e). Intriguingly, except for the increase of shear rate at the same pressure, the diffraction peaks also get sharper when decreasing the pressure at the same shear rate. For instance, at a shear rate of 30 s^{-1} , the FWHM of (200) diffraction peaks are approximately 52° , 41° , 33° and 23° for 100, 50, 25 and 5 MPa, respectively. The vari-

ation of FWHM of (200) diffraction peaks implies that high degree of crystalline chain orientation is favorable under stronger shear flow and at low pressure.

The molecular chain orientation of PPS is further quantitatively analyzed based on the azimuthal distributions of (200) diffraction intensity. As shown in Fig. 7, f_H decreases from about 0 to -0.2 with increasing shear rate, indicating the higher order of crystal orientation with its c -axis parallel to the flow direction. The fact that crystal orientation increases with increasing shear rate is the most common phenomenon in the field of flow-induced polymer crystallization, as more chains are strongly stretched and crystallize into highly oriented structures at higher shear rates.^[18–21] Meanwhile, f_H shows a tendency to decrease with lowering the pressure,

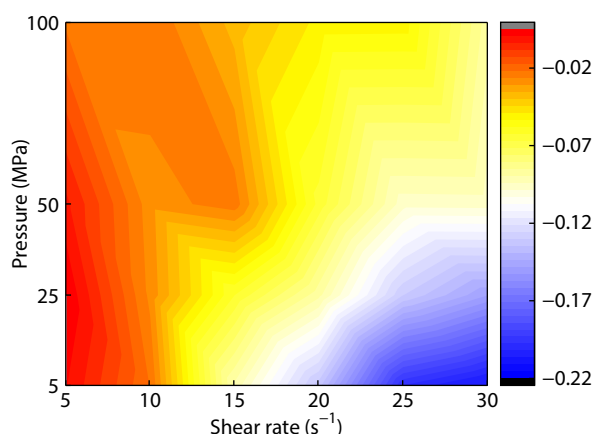


Fig. 7 f_H of PPS crystallized at different shear rates (5, 10, 15, 20, 25 and 30 s^{-1}) and pressures (5, 25, 50 and 100 MPa). The color sequence from red to blue represents the decreasing f_H .

e.g., at 30 s^{-1} f_H is approximately -0.21 , -0.15 , -0.10 and -0.09 for 5, 25, 50 and 100 MPa, respectively. The weaker crystal orientation at high pressures is likely to be caused by the sluggish crystallization under these conditions. Increasing pressure causes an exponential increase in melt viscosity, leading to a significant reduction in crystallization rate.^[46,47] Thus, at higher pressures, most of oriented chains may relax back to the coil state, which later crystallize into unoriented crystals. Notably, the effect of pressure on flow-induced crystallization of rigid-chain polymers is rarely explored. The present results imply that oriented crystals of rigid-chain polymers tend to form under stronger shear flow and at lower pressures. This empirical conclusion can serve as a guide on the manipulation of chain orientation in rigid-chain polymer products to enhance their performance.

DISCUSSION

The present results indicate that processing parameters, i.e., pressure and shear rate, affect tribological properties of neat PPS by altering the molecular chain orientation. Nonetheless, the reason why molecular chain orientation can improve the

tribological performance remains unclear. In our work, the entangled network of PPS chains could be deformed when the shear rate was faster than the relaxation time. Parts of chains were stretched out from the cluster of unoriented molecules, and were likely to develop into oriented chain bundles. The oriented chain bundles evolved into row-nuclei, and provided a set of sites for the assembly of PPS chain segments into edge-on oriented crystals through the multistep crystallization process (i.e., conformation transition, density fluctuation, forming precursors, and oriented crystallites),^[17–20,48] the sliding direction during tribological tests is thus along the chain backbone (right part of Fig. 8), giving rise to a low energy barrier to interfacial slip.^[15,49] This explains why the molecular chain orientation results in a low friction force. Meanwhile, the extended chain segments possess a rigid link between molecular units and a regular spacing that meets the equilibrium van der Waals' interactions.^[42] Hence, the majority of friction energy can be dissipated with minor structural changes in terms of the disruption and reconstruction of van der Waals interactions within the adjacent PPS chain segments, finally leading to a low W_s . But for PPS with unoriented crystal, the sliding direction can be either parallel, or titled and perpendicular to the crystals. The lack of covalent bonds in neighboring chain segments provides some freedom in their spacing, resulting in an intense structural reorganization for energy dissipation during sliding.^[15,49,50] Besides, when the sliding direction is perpendicular to the chain segments, chain scission (breakage carbon-sulfur bonds) is more likely to occur near the contact zone of the randomly arranged lamellae due to the accumulation of friction heat.^[41]

To reveal the quantitative relation between W_s and molecular chain orientation, the existing wear model, that is, the widely-accepted Archard law^[51,52] is revised on the basis of our findings. In 1953, Archard formulated the law of adhesive wear, $W_s = V/F_N L = K_0/H_0$, where V , K_0 , F_N , L , and H_0 are wear volume, wear coefficient, normal load, sliding distance, and hardness for polymers, respectively. In this equation, the parameter related to the molecular feature of the material is only H_0 , without concerning the effect of molecular chain orientation. However, as a matter of fact, some reports have previously revealed that the densely packed and rigidly linked

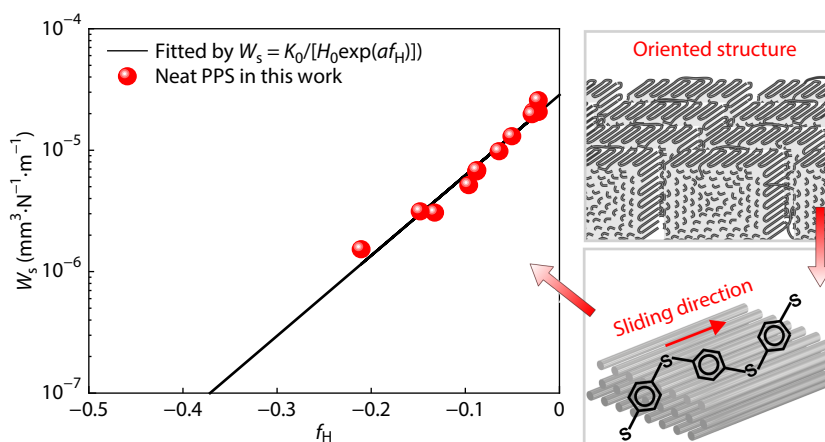


Fig. 8 The relation between W_s and f_H , fitted by the revised Archard law (left). Schematic representation of oriented crystalline superstructure in PPS sample (top right) and the parallel sliding friction (bottom right).

molecules demonstrated high lateral stiffness when sliding the tip along the backbone direction.^[50,53,54] Currently, our work also found that molecular chain orientation of PPS significantly reduces the W_s of PPS. Therefore, we revised the Archard law by introducing the effects of f_H on the W_s . The derivation of modified Archard law is shown as follows: according to the microhardness studies of polymer materials,^[55–57] the H_0 of folded-chain crystal is proportional to the elastic modulus (E), $H_0 \approx E/10$. Meanwhile, the exponential relationship between chain orientation and E has been widely reported in the literature,^[58–60] $E(f_H) \approx E(f_H=0) \exp(af_H)$, where a is the fitting parameter. Combining the above empirical correlations, one can obtain a similar exponential relationship between the hardness of oriented polymer samples, $H(f_H)$ and the f_H (the detailed deduction process is displayed in section 10 of ESI). As shown in Fig. 8, the relation between W_s and f_H is well fitted by the revised Archard equation, $W_s = K_0 / [H_0 \exp(af_H)]$, where a is the constant related with temperature. When H_0 is set as 400 MPa,^[61] a is equal to -15.2 , and K_0 is approximately 1.1×10^{-5} . Noted that the calculated value of K_0 is accordant with the reported value of PPS.^[62,63] At this moment, W_s is a function of f_H . The revised Archard law further complements the wear mechanism for oriented polymer systems, which may help the plastic engineers to predict the tribological performance of polymer products without doing destructive tribological tests.

CONCLUSIONS

In summary, neat PPS with ultra-high tribological performance has been successfully fabricated by optimizing the processing parameters (pressure and shear rate). The tribological properties of PPS can be strongly enhanced by manipulating the molecular chain orientation in polymer processing. An exponential relation between the wear rate and molecular chain orientation of PPS is presented, which is well fitted by the revised Archard law. Our current work unfolds the possibilities of manufacturing PPS products with high tribological performance without using additives, and thus, extends their application in the high-demanding tribological field.

NOTES

The authors declare no competing financial interest.

Electronic Supplementary Information

Electronic supplementary information (ESI) is available free of charge in the online version of this article at <http://doi.org/10.1007/s10118-022-2672-x>.

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