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A unified theory of wear for ultra-high molecular weight polyethylene in multi-directional sliding

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Abstract

A theoretical wear model based on the effective frictional work concept was developed for ultra-high molecular weight polyethylene in lubricated multi-directional sliding. The model assumes the occurrence of preferential molecular alignment in the principal direction of sliding and rupture or splitting of the oriented molecules in the perpendicular direction associated with secondary sliding. The model unifies the effects of interfacial friction, crosslink density and cross-shear angle in a simple equation and provides a theoretical explanation for the effects of crosslinking and multi-directional motion on the wear of UHMWPE in total hip and total knee replacements. Experimental results correlating each of the three variables to the wear rate of UHMWPE in a hip joint simulator and a multi-directional motion wear test machine verified the general validity of the model. © 2001 Elsevier Science B.V. All rights reserved.

Keywords: UHMWPE; Wear; Cross-linking; Cross-shear; Total joint replacement

1. Introduction

Ultra-high molecular weight polyethylene (UHMWPE) has been used as a major bearing surface in both total hip and total knee joint prostheses since the early 1960s. Owing to its extraordinarily high molecular weight, UHMWPE possesses superior mechanical toughness and wear resistance over most other polymers. As an acetabular component in total hip replacement, UHMWPE wears at an annual rate of approximately 0.1 mm/year or 80 mm^3/year against a 32 mm femoral head [1–4]. This represents a 30-fold reduction in wear rate compared to a PTFE socket that was first introduced by Charnley in the late 1950s [5,6]. Typical acetabular components have a minimum thickness >6 mm. At a wear rate of 0.1 mm/year, it would take at least 60 years to wear through. Indeed, the wear-through of UHMWPE sockets has been extremely rare in clinical practice. However, it was discovered recently that wear debris produced from UHMWPE causes adverse tissue biological responses which result in gradual bone absorption or osteolysis [7–9]. A clinical consequence of debris-induced osteolysis is loosening of the prosthesis and discomfort to the patient which eventually lead to revision surgery. Therefore, to ensure the long-term success of the prosthesis, the quantity and the rate of UHMWPE wear debris generated in vivo must be minimized.

Recent clinical studies of the wear debris retrieved from periprosthetic tissues revealed that the overwhelming majority of the wear particles is micron or submicron in size and fibrous or particulate in shape [10–14]. Although the particulate debris is more numberous in quantity than the fibrous debris, the latter accounts for a majority of the volume of particles. A number of descriptive wear mechanisms have been proposed to explain the morphological nature of the fibrous wear debris [15–17]. These include adhesive, abrasive and fatigue wear mechanisms. However, there has been no quantitative attempt to correlate the wear rate to either extrinsic or intrinsic material variables.

A major break-through in the understanding of the wear mechanisms for UHMWPE is the discovery of the importance of multi-directional motion in the process of wear debris generation [18–20]. Linear-tracking motion either unidirectional or reciprocating produces an extremely small wear rate in UHMWPE which is typically two–three orders of magnitude less than the average wear rate observed clinically [18–22]. Wear rates produced by multi-directional motion hip joint simulators, on the other hand, have consistently matched the average clinical wear rates [18–22]. It was proposed by a number of researchers that the motion-dependent behavior of UHMWPE wear is attributed to the unique molecular structure of UHMWPE whose molecules orient preferentially in the direction of sliding

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Nomenclature

d	average cross-sectional width of UHMWPE fibrils (mm)
k	wear factor (mm^3/Nm)
k'	constant
l	length of the fibril (mm)
L	sliding distance per cycle (m)
M_c	average molecular weight between cross-links (g/mole)
M_0	critical molecular weight between cross-links (g/mole)
$\vec{P}$	normal load (N)
$\vec{q}$	tangential force applied to a fibril (N)
Q	tangential load (N)
S	swell ratio
t	time (s)
$\vec{v}$	linear velocity vector (m/s)
V_l	molar volume of xylene (cm^3/mole)
ΔV	volume loss per cycle (mm^3/cycle)
ΔW	total frictional work per cycle (Nm)
ΔW_x	total frictional work released in the principal sliding direction per cycle (Nm)
ΔW_y	total frictional work released in the secondary sliding direction per cycle (Nm)
X_c	cross-link density (mole/g)
X_0	critical cross-link density (mole/g)
Greek letters	
α	maximum cross-shear angle (radian)
χ	polyethylene–xylene interaction parameter
γ_c	C–C bond energy (Joule)
μ	coefficient of friction
μ_0	critical coefficient of friction
θ	cross-shear angle between an instantaneous velocity vector and the principal sliding direction
ω	angular velocity (radian/s)

[18,19,23]. In linear-tracking motion, molecular orientation leads to strain-hardening of the wear surface which results in an increasing wear resistance as sliding proceeds. In multi-directional motion, the UHMWPE wear surface experiences both shear and tensile stresses in multiple directions. Strengthening in one particular direction results in weakening in a perpendicular direction — a phenomenon that was often observed in oriented linear polymers [19,24]. Recent computer simulation of the human joint kinematics has indeed indicated that stresses experienced by the surface in both the hip and the knee joint are multi-directional [19,20,25].

The discovery of the adverse effect of molecular orientation in multi-directional sliding has resulted in the development of using crosslinking as a means to improve the wear resistance of UHMWPE [26–30]. Crosslinking makes the UHMWPE more isotropic. First, crosslinking retards chain mobility and hence reduces the degree of molecular orientation during sliding. Second, crosslinking increases the density of the C–C chemical bonds between adjacent molecular chains and hence makes it more difficult to split one molecule from another. Recent hip joint simulator studies have indeed demonstrated that crosslinked UHMWPE possessed far superior wear resistance than non-crosslinked linear UHMWPE [26–30].

The present study looks at the anisotropic wear behavior of UHMWPE from a theoretical point of view. The physical process of wear debris generation was considered as a intermolecular splitting phenomenon. The UHMWPE wear surface was treated as an composite structure consisting of numberous micro-fibrils oriented in the principal sliding direction. A wear particle is produced as a fibril is separated from its neighbours by tearing rupture in the secondary sliding direction. A quantitative model was then developed. Three separate model experiments were conducted under lubricated multi-directional sliding conditions to verify the validity of the theory.

2. Theory

Fig. 1(a) shows schematically the contact between a smooth deformable UHMWPE surface and a smooth rigid surface under a normal load $\vec{P}$. Relative motion between these two surfaces is considered in such a way that at any given moment the velocity vector $\vec{v}$ rotates at an angular velocity of ω in the x–y plane, Fig. 1(b). Denoting θ as the angle between the velocity vector $\vec{v}$ and the x-axis at time t, α as the maximum angle of the upper extreme position of the velocity vector and $-\alpha$ as the lower extreme position of the velocity vector, then the velocity vector rotates from 0° to α and then reverses to $-\alpha$ and then reverses again back to 0° to complete a motion cycle. As the velocity

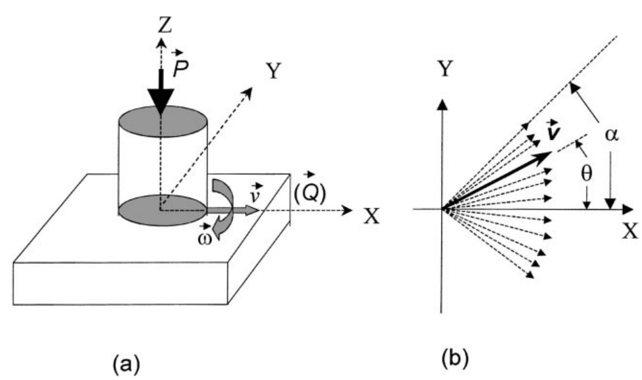

Fig. 1. (a) Contact between two solid surfaces under a normal load $\vec{P}$, and (b) rotation of the velocity vector $\vec{v}$ in the X–Y plane during a motion cycle.

vector changes its direction constantly, the tangential force Q will change its direction also. For simplicity, assuming that the normal load is constant, then the total frictional work released per cycle is given by the following equation:

$$\Delta W = \int_0^t (\vec{Q}\vec{v})\,\mathrm{d}t = 4\int_0^\alpha \left(\mu\vec{P}\vec{v}\frac{1}{\omega}\right)\mathrm{d}\theta = \frac{4\mu P v\alpha}{\omega} \qquad (1)$$

where μ is the coefficient of friction. The total frictional work released in the y direction is given by

$$\begin{aligned}\Delta W_y &= \int_0^t (Q\sin\theta)(v\sin\theta)\,\mathrm{d}t\\ &= 4\int_0^\alpha (\mu P v\sin^2\theta)\left(\frac{1}{\omega}\right)\mathrm{d}\theta\\ &= \frac{2\mu P v}{\omega}\left(\alpha - \frac{\sin 2\alpha}{2}\right)\end{aligned} \qquad (2)$$

The total frictional work released in the x direction is

$$\Delta W_x = \Delta W - \Delta W_y = \frac{2\mu P v}{\omega}\left(\alpha + \frac{\sin 2\alpha}{2}\right) \qquad (3)$$

Let us then consider the interaction between a stress vector $\vec{q}$ and a oriented UHMWPE fibril of length l and diameter d (Fig. 2). In order for the fibril to become a loose wear particle, it must be separated from its neighbours through tensile rupture in the x–z plane and shear rupture in the x–y plane. The energy or work required to separate this fibril from its neighbours is given by

$$\delta w = 2(dlX_\mathrm{c}\gamma_\mathrm{c}) \qquad (4)$$

where X_c is the C–C bond density (number of bonds per area) and γ_c the C–C bond energy.Since the volume of the fibril δV is ld^2, the work required to produce a unit volume of debris is given by

$$\frac{\delta w}{\delta V} = \frac{2dlX_\mathrm{c}\gamma_\mathrm{c}}{ld^2} = \frac{2X_\mathrm{c}\gamma_\mathrm{c}}{d} \qquad (5)$$

Let us assume that ΔW_x contributes to the elongation of the fibrils while ΔW_y contributes to the rupture of the fibrils.

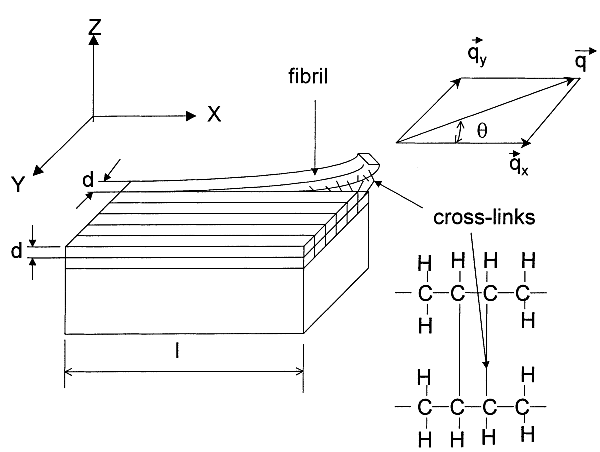

Fig. 2. An inter-fibril rupture model of an oriented UHMWPE surface by an off-axis force vector $\vec{q}$.

Since molecular elongation leads to strengthening, rupture in the elongation direction is unlikely. Hence, ΔW_x may be considered "ineffective" as far as wear is concerned. In the transverse direction, on the other hand, the rupture strength decreases as the molecules elongate. Therefore, ΔW_y may be considered "effective" in producing wear.

Then, the total volume of wear debris produced per cycle is given as the "effective work" divided by the specific work required to produce a unit volume of wear debris, i.e.

$$\begin{aligned}\Delta V &= k'\frac{\Delta W_y}{(\delta w)/(\delta V)} = k'\frac{2\mu P v d}{\omega X_\mathrm{c}\gamma_\mathrm{c}}\left(\alpha - \frac{\sin 2\alpha}{2}\right)\\ &= k'\frac{2\mu P v d}{X_\mathrm{c}\gamma_\mathrm{c}}\times\left(\frac{\alpha}{\omega}\right)\times\left(1 - \frac{\sin 2\alpha}{2\alpha}\right)\end{aligned} \qquad (6)$$

where k' is a proportionality constant. Since $(4\alpha/\omega) = t$, where t is the time required to complete a motion cycle, then Eq. (6) can be rewritten as

$$\Delta V = k'\frac{\mu P d(vt)}{2X_\mathrm{c}\gamma_\mathrm{c}}\times\left(1 - \frac{\sin 2\alpha}{2\alpha}\right) \qquad (7)$$

Since $(vt) = L$, where L is the average linear sliding distance travelled per motion cycle, then Eq. (7) becomes

$$\Delta V = k'\frac{\mu P d L}{2X_\mathrm{c}\gamma_\mathrm{c}}\times\left(1 - \frac{\sin 2\alpha}{2\alpha}\right) \qquad (8)$$

The wear factor k which is defined as the volume loss per unit sliding distance per unit load, i.e. $k = \Delta V/(PL)$, is, therefore, given by

$$k = k'\frac{\mu d}{2X_\mathrm{c}\gamma_\mathrm{c}}\times\left(1 - \frac{\sin 2\alpha}{2\alpha}\right) \qquad (9)$$

Eq. (9) represents the wear rate per unit sliding distance per unit load as a function of the material properties ($X_\mathrm{c}\gamma_\mathrm{c}$), the lubrication conditions μ, the degree of the cross-path motion $(1-\sin 2\alpha/2\alpha)$ and the cross-sectional width of the fibrils d.

At $\alpha = \pi/2 = 90°$, i.e. at the maximum angle of "cross-shear", the wear factor k obtains its maximum value as represented by the following equation

$$k_{\max} = k_{\pi/2} = k'\frac{\mu d}{2X_\mathrm{c}\gamma_\mathrm{c}} \qquad (10)$$

Therefore, Eq. (9) can also be written as

$$k_\alpha = k_{\pi/2}\left(1 - \frac{\sin 2\alpha}{2\alpha}\right) \qquad (11)$$

3. Experiments

3.1. Model Experiment 1: k versus μ

Hemispherical sockets of various diameters were machined from extruded GUR1050 UHMWPE rod (PolyHi Solidur, Fort Wayne, IN) with an average molecular weight

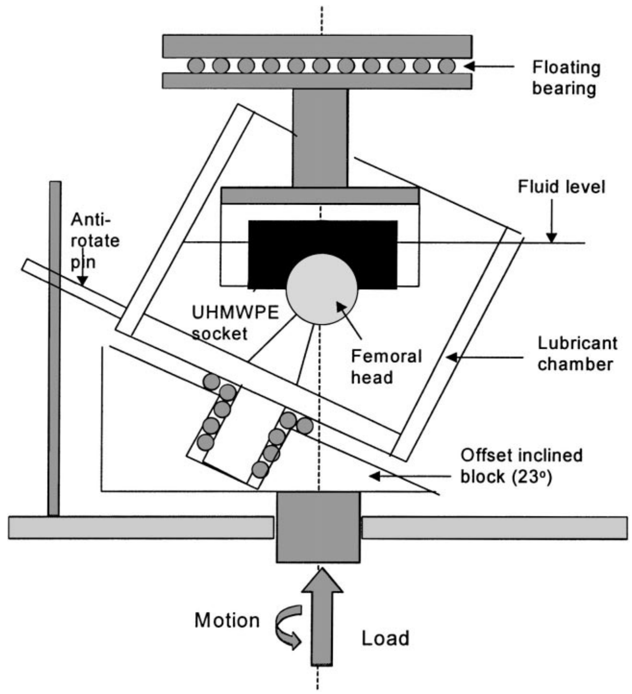

Fig. 3. Schematic illustration of the hip joint simulator.

of approximately 6 million. These sockets have the following diameters: 32.05, 32.55, 33.05, 33.55, 34.05, 35.05 and 36.05 mm. Polished CoCr femoral heads with a common diameter of 32.00 mm and a center-line average roughness of 0.015 mm (R_a) were used as an articulating counterface against the UHMWPE sockets in both a friction simulator [31] and a wear simulator [19]. In the friction simulator, the femoral head oscillates inside the socket about the load axis at 30°/s. A constant load of 2000 N was applied. A diluted alpha-fraction serum (50% serum + 50% deionized water) was used as a lubricant. Frictional force was measured via a torque cell. The frictional test was run for 3000 cycles. During the initial 500 cycles, the frictional force showed large fluctations. After 500 cycles, it reached a steady state. Therefore, only the steady-state values were reported here. Two samples of each size were measured. In the wear simulator, the femoral head moves inside the socket at a biaxial rocking motion of ±23°, which gave an average linear sliding distance of 33 mm according to an earlier computer simulation [19].Meanwhile, a physiological load with 2450 N maximum and 50 N minimum (average load: ∼1250 N) was applied. Motion and loading were synchronized at 1 Hz. The combination of the biaxial rocking motion and physiological loading produced a multi-directional surface shear stress pattern at the articulating interface. Fig. 3 shows a schematic of the wear simulator while Fig. 4 shows a distribution of the normalized surface shear stress vectors at the head/socket interface during a loading cycle [19]. The shear stress vector was seen to change its magnitude and direction constantly during the loading cycle. The maximum angle of “cross-shear” was 90°. The wear simulator test was run under the same lubrication condition as the friction test. The test was run for 2 million cycles. Two samples of each size were tested. Weight loss of each socket was measured at every 0.5 million cycle intervals, which was then converted to volume loss (mm^3) by dividing by the material density (0.933 g/cm^3). Volumetric wear rate was defined as volume loss per motion cycle which was obtained by linear regression. Since the average load and sliding distance per

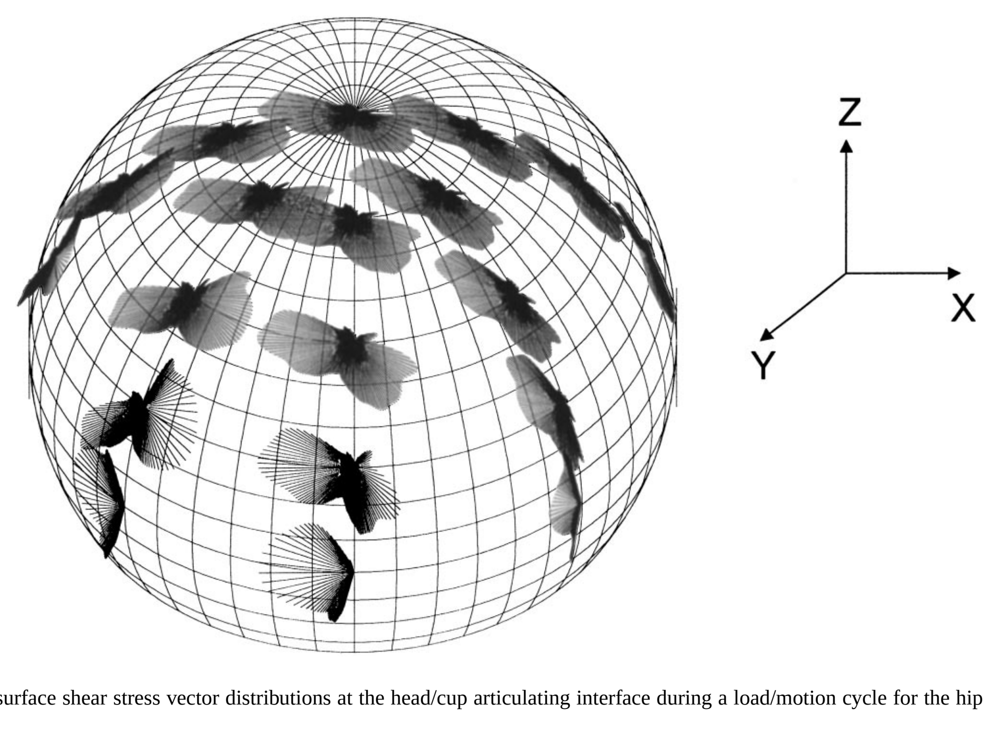

Fig. 4. Normalized surface shear stress vector distributions at the head/cup articulating interface during a load/motion cycle for the hip joint wear simulator.

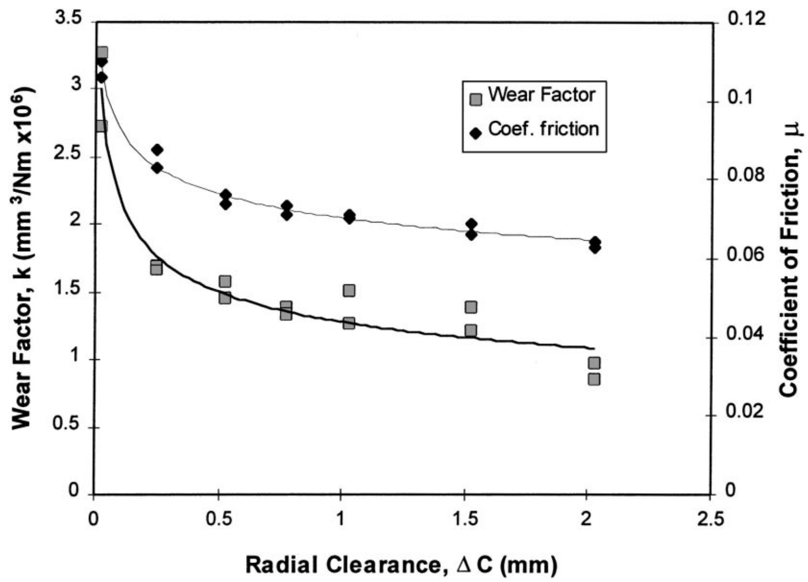

Fig. 5. Coefficient of friction and wear factor as a function of the head/cup radial clearance.

motion cycle were 1250 N and 0.033 m, respectively, the wear factor k (unit: mm^3/Nm) can be calculated from the volumetric wear rate by the following equation

$$k = \frac{\Delta V}{\bar{P}L} = \frac{\Delta V}{41.25}\ (\mathrm{mm}^3/\mathrm{Nm}) \tag{12}$$

Fig. 5 shows the coefficient of friction and the wear factor as a function of the ball/socket radial clearance. Both μ and k decreased as clearance increased. Fig. 6 shows a plot of k as a function of μ, which was seen to increase linearly with increasing μ. However, the best-fit straight line of the data ($R^2 = 0.90$) does not go through the origin as indicated by Eqs. (9) and (10). Rather it intersects with the horizontal axis at $\mu = \mu_0 = 0.04$. This suggests that there exists a critical value of coefficient of friction below which rupture between fibrils would not occur. To better reflect the experimental results, Eq. (9) may be modified as follows

$$k = k' \frac{d(\mu - \mu_0)}{2X_c\gamma_c} \times \left(1 - \frac{\sin 2\alpha}{2\alpha}\right) \tag{13}$$

3.2. Model Experiment 2: k versus M_c

UHMWPE (GUR4050) bars were irradiated with γ-rays at 3.0, 4.0, 5.0, 7.5, 10.0 and 15.0 Mrads. Following irradiation, the bars were annealed at 130°C to eliminate all remaining free radicals. Hemispherical sockets with a common diameter of 32.05 mm were then machined from these bars. The

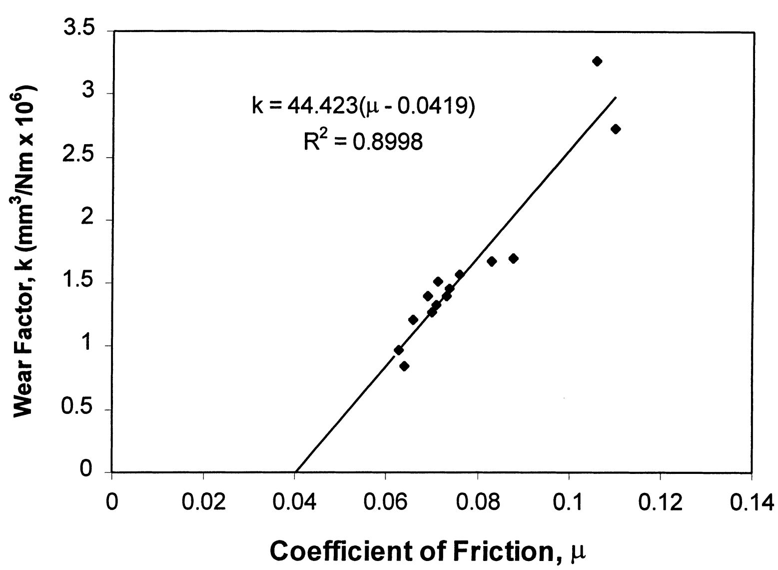

Fig. 6. Wear factor vs. coefficient of friction.

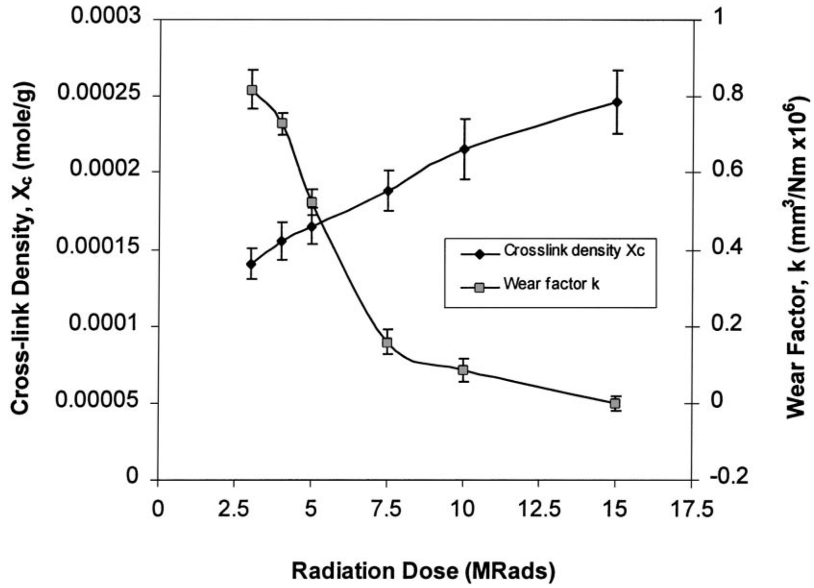

Fig. 7. Crosslink density and wear factor vs. radiation dose.

same hip wear simulator was used to test these sockets. The same test conditions and protocols as in Model Experiment 1 were used in the present test. In addition, the crosslink density of the materials was measured via the swell-ratio method according to ASTM D2765. The following equation was used to calculate the crosslink density [30,32]

$$X_c = \frac{1}{M_c} = -\frac{\ln(1 - S^{-1}) + S^{-1} + \chi S^{-2}}{V_l S^{-1/3}} \tag{14}$$

where X_c is the crosslink density (mole/g), M_c the average molecular weight between crosslinks (g/mole), V_l is the molar volume of xylene (136 cm^3/mole) [33], χ the xylene–polyethylene interaction parameter ($\chi = 0.33 + 0.55/S$) [34] and S is the swell ratio. For both the wear test and swell-ratio measurement, three specimens from each material were used.

Fig. 7 shows the wear factor k and the crosslink density X_c as a function of the radiation dose. While k decreased significantly with increasing the radiation dose, X_c increased as the dose increased. Fig. 8 shows k as a function of $1/X_c$ (or M_c). A linear relationship between k and M_c was obtained. However, the best-fit line of the data ($R^2 = 0.94$) does not go through the origin as M_c approaches zero. The line intersects with the horizontal axis at $M = 1/X_c = 1/X_0 =$

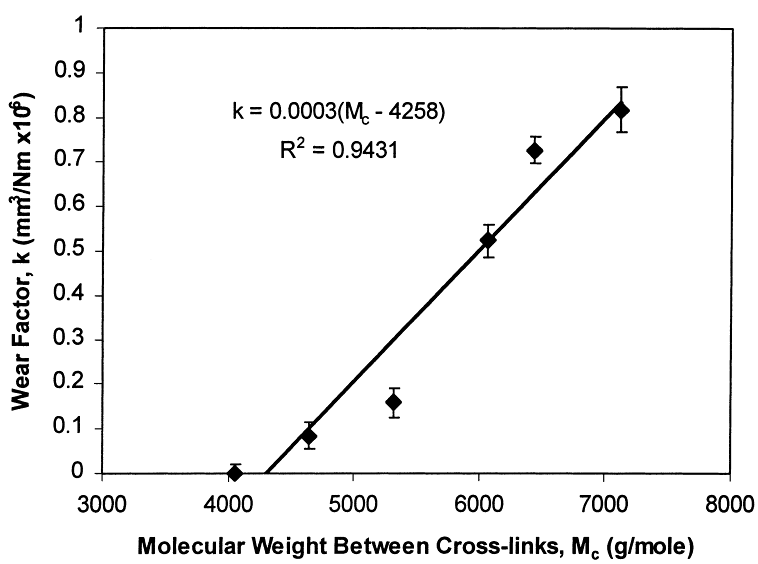

Fig. 8. Wear factor vs. molecular weight between crosslinks.

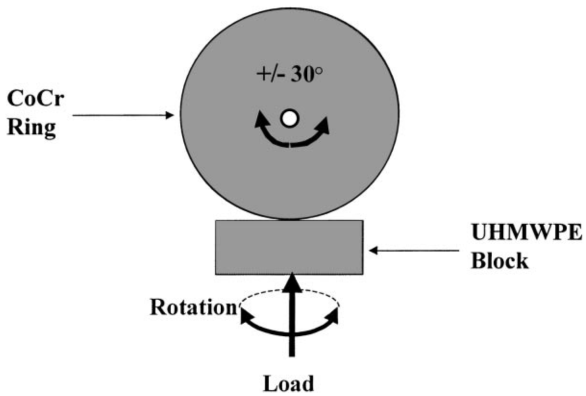

Fig. 9. Schematic of the loading/motion configuration of the biaxial line-contact wear tester.

4250 g/mole. This implies that there exists a critical value of the crosslink density above which rupture between fibrils cannot occur.

3.3. *Model Experiment 3: k versus α*

In this experiment, a ring-on-block wear tester was used to verify the relationship between the wear factor k and the maximum cross-shear angle α. Rectangular blocks with dimensions of 12.5 mm × 12.5 mm×10 mm were machined from a GUR1050 rod. These specimens were irradiated at 3.0 Mrads and then annealed at 130°C. The ring was made of CoCr alloy with the dimensions of 75 mm diameter and 25.4 mm wide. The circular surface of the ring was highly polished (R_a: 0.025 μm). The UHMWPE block was loaded against the CoCr ring at a constant load of 1120 N. The ring oscillated against the block at ±30° while the block oscillated about the load axis at 0, ±7.5° and ±15°, respectively. Therefore, the maximum angles of cross-shear (α) simulated here were 0, 7.5° and 15°, respectively. The average linear sliding distance was 79 mm. Fig. 9 shows a schematic of the biaxial ring-on-block wear tester. Fig. 10 shows the surface shear stress vector distribution during a motion cycle. The test was also run for 2 million cycles under serum lubrication. Three specimens were used for each test conditions.

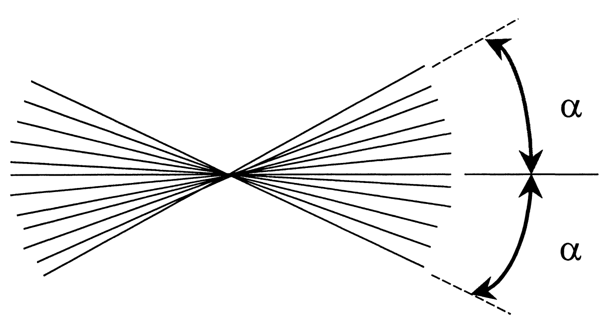

Fig. 10. Distribution of surface shear stress vectors during a motion cycle for the biaxial wear tester.

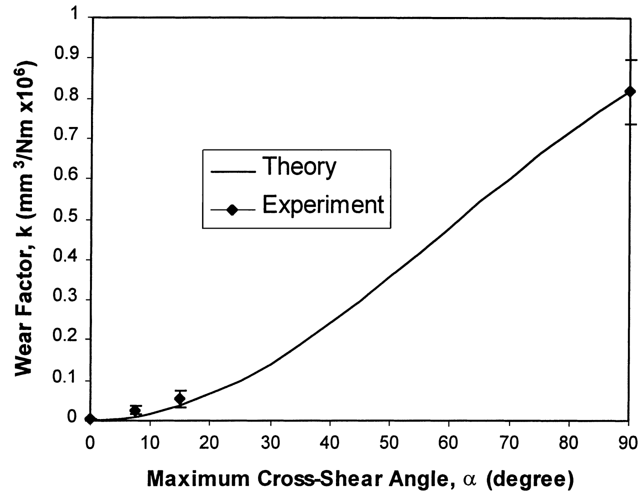

Fig. 11. Wear factor vs. maximum cross-shear angle.

Fig. 11 shows the wear factor as a function of maximum cross-path angle. For comparison, the wear factor $k_{\pi/2}$ as obtained from the hip simulator test in which $\alpha = 90^\circ$ was also included in Fig. 11. With the $k_{\pi/2}$ known, the theoretical relationship between k_α and α as shown in Eq. (11) could be calculated and was plotted in Fig. 11. The experimental wear factor k increased dramatically with increasing α as predicted by the theory. From $\alpha = 0$ to 7.5°, k increased by an order of magnitude; from 7.5 to 15°, k increased by a factor of 3; from 15 to 90°, k increased by another order of magnitude. Therefore, the theoretical relationship between k and α was verified by Model Experiment 3.

4. Discussion

The present study represents a first attempt to quantify the significance of multi-directional motion on the wear rate of UHMWPE. By considering the anisotropy of UHMWPE molecules in response to a rotating shear stress vector, a wear theory was developed and successfully explained a number of recent discoveries in the wear of UHMWPE prosthetic bearing surfaces. First of all, the theory revealed the fundamental importance of the secondary motion of the human joints in the wear of UHMWPE. For the hip joint, the secondary motion is the abduction/adduction; for the knee joint, it is internal/external rotation. This secondary motion must be simulated in a wear test in addition to the principle motion. Second, the theory explained the critical role of crosslinking in the wear of UHMWPE. Third, the theory indicated that contact stress per se is not the dominant factor in UHMWPE wear. In fact, increasing the contact stress at the expense of contact area under a given load reduced the coefficient of friction and the wear rate.

The experimental results indicate that there may exist a threshold experimental condition (extrinsic variables) for a given UHMWPE and a threshold crosslink density (intrinsic variable) under a given test condition to initiate or surpress wear debris formation. For a non-crosslinked UHMWPE, the coefficient of friction must exceed 0.04 in order to produce wear. It should be noted here that the coefficient of friction recorded here was under the conditions of linear sliding. In multi-directional motion, the coefficient of friction may be different as reported by Briscoe and Stolarski [35,36] under dry sliding. Therefore, the assumption made in Eq. (2) that the coefficient of friction is the same along and across the direction of fibril orientation may not be accurate. Under the experimental conditions used in the hip simulator test, the crosslink density must exceed 0.00025 mole/g ($M_c \leq 4250$ g/mole) in order to reduce the wear rate to a non-measurable value. Taking these threshold conditions into consideration, the theoretical model may be better written as follows

$$k = k' \frac{d(\mu - \mu_0)}{2\gamma_c} \times \left(\frac{1}{X_c} - \frac{1}{X_0}\right) \times \left(1 - \frac{\sin 2\alpha}{2\alpha}\right) \quad (15)$$

or

$$k = k' \frac{d(\mu - \mu_0)}{2\gamma_c} \times (M_c - M_0) \times \left(1 - \frac{\sin 2\alpha}{2\alpha}\right) \quad (16)$$

where M_c is the average molecular weight between crosslinks and M_0 the critical average molecular weight between crosslinks. In a biaxial pin-on-disk study, Muratoglu et al. found that the wear rate of crosslinked UHMWPE increased linearly with M_c [30]. These authors also obtained a threshold value of approximately 4500 g/mole for M_c for “zero” wear rate. These findings provided further experimental support for Eq. (16).

According to the theory outlined in Eqs. (9), (15) and (16), the wear factor k should increase with increasing the average cross-sectional width of the fibril d. Since d is a consequence of the final wear event and cannot be determined theoretically without conducting a quantitative analysis of the wear debris itself, the theory remains semi-empirical. It may be argued that d should be proportional to the depth or thickness of the highly deformed surface layer during steady state sliding. A recent study using transmission electron microscopy by Edidin et al. revealed that the depth of the highly oriented surface layer was much less for a highly crosslinked UHMWPE than for a non-crosslinked UHMWPE [37]. These authors then attributed the increased wear resistance of highly crosslinked UHMWPE to the reduced thickness of the oriented surface layer or the so-called “plasticity-induced damage layer”. The present theory suggests that the reduced thickness of the “damage layer” is only part of the explanation for the observed benefits of crosslinking. Another benefit of crosslinking, perhaps a major one, comes from the increased resistance to inter-fibril rupture due to increased crosslink density. In other words, crosslinking contributes to increased wear resistance in two major ways: (a) increased resistance to plastic strain accumulation or molecular orientation, and (b) increased resistance to inter-molecular rupture.

A major limitation of the theory is the uncertainty of the constant k' which is defined as the fraction of the total frictional work input in the direction perpendicular to the molecular orientation direction that goes directly to produce a wear particle. Therefore, k' is always less than unity. A majority of the frictional work is dissipated as heat. Experimental variables that affect the mechanisms of heat transfer from the articulating interface to the surrounding environment may affect k'. It is, therefore, expected that the wear factor k may be affected by the frequency of motion, the thermal conductivity of the test fixtures and the components, the equilibrium temperature of the lubricant, and so on.

The theory developed in this study may not apply to unlubricated sliding contact where polymer transfer film formation prevails. In fact, Briscoe and Stolarski observed an opposite effect of multi-directional motion on the wear rate of linear polymers under dry sliding contact conditions in a pin-on-disk setup [35,36]. They found that an addition of load-axis spin to linear sliding actually decreased the wear rates while increasing the coefficient of friction for PTFE, HDPE and UHMWPE. In the absence of transfer film formation, the wear rate of the UHMWPE is determined by the cohesive energy density of the molecular bonds. The mechanism of wear in the lubricated multi-directional sliding of UHMWPE is clearly belong to the cohesive wear category although the wear takes place within an extremely thin surface layer. Interestingly, the counterface used in the present experiments was extremely smooth (R_a: 0.01–0.025 μm). With such a smooth counterface, abrasive wear was unlikely to occur. Micro-fatigue wear which results from elastic contact of microasperities and results in the generation of particulate wear debris was also unlikely since the debris produced from the present experiments was mostly fibrous in shape indicating extensive plastic deformation of the wear surface before failure. The remaining likely wear mechanism was micro-adhesive wear. However, unlike the classic definition of adhesive wear where wear debris is produced by shear rupture of adhesive junctions [38,39], the special micro-adhesive wear mechanism that may apply to the current situation requires surface rupture to occur prior to subsurface shear rupture. Otherwise, the motion-dependent wear phenomenon could not be explained.

A unique feature of the wear theory proposed here is the concept of the effective frictional work. The theory assumed that only the frictional work released in the direction perpendicular to the principal sliding direction participates in the generation of wear debris. In other words, the frictional work released in the principal sliding direction is wasted and hence ineffective. This concept is not so easy to comprehend since both components of the total work released have to coexist in order to produce a fibrous wear particle. Without either one, the nature of sliding becomes linear-track motion and the wear rate of the UHMWPE becomes negligibly small. Perhaps a mechanistic model based on a critical failure criterion other than the energy criterion used in this study would give a better answer to this paradox.

5. Conclusions

A theoretical wear model based on the effective frictional work concept was developed for UHMWPE under lubricated multi-directional sliding conditions. According to the theory, the wear factor k is related to the coefficient of friction μ, the cross-link density X_c, and the maximum cross-shear angle α by the following equation

$$k = k' \frac{d(\mu - \mu_0)}{2\gamma_c} \left(\frac{1}{X_c} - \frac{1}{X_0}\right) \times \left(1 - \frac{\sin 2\alpha}{2\alpha}\right)$$

where k' is a constant, d the diameter of the fibrils, m_0 the critical coefficient of friction for initiating surface failure and X_0 the critical cross-link density for surpressing surface rupture.

In a hip joint simulator test, the wear rate and the coefficient of friction of UHMWPE acetabular components decreased with increasing the head/cup clearance. The wear rate was shown to increase linearly with increasing the coefficient of friction.

Radiation crosslinking significantly decreased the wear rate of UHMWPE as shown by the hip simulator result. The wear rate was seen to increase linearly with increasing the average molecular weight between crosslinks.

The wear rate of UHMWPE increased dramatically with increasing the maximum cross-shear angle. The wear factor k increased by three orders of magnitude as the maximum cross-shear angle increased from 0 to 90°.

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