Dry Friction and Wear Behavior of Forged Co–29Cr–6Mo Alloy without Ni and C Additions for Implant Applications

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A dry wear behavior of a forged Co–Cr–Mo alloy without Ni and C additions have been investigated using a ball-on-disc type wear testing machine with an alumina ball in ambient air. The wear factor of the Co–Cr–Mo forged alloy without Ni and C additions (hereafter, designated the forged alloy) shows negative contact load dependence. The coefficient of friction decreases with increasing contact load. Worn surfaces are hardened during the wear tests, forming oxide films. This results from significantly high work hardening rate of the forged alloy, caused by the strain-induced martensitic transformation from an fcc-γ phase to an hcp-ε phase, which contributes to the improvement in the dry wear resistance.

Wear mechanisms of the forged alloy are discussed on the basis of Hertzian contact theory and observations of the wear scars formed on the alloy disc and the alumina ball surfaces. It is considered that the dominant wear mechanism of the forged alloy is the mild adhesive wear, though the extrinsic abrasive wear mediated by the wear debris, i.e., third-body abrasive wear, is exerted as an extrinsic wear mechanism. In addition, it is suggested that a delamination wear resulting from the fatigue fracture likely occurs under the present dry wear condition.

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1. Introduction

For the replacement of tissue and organ injured by accidents or disease, artificial materials have frequently used in orthopedic and dental surgery. The artificial materials are called “biomaterials”, which consists of metals, ceramics and/or polymers. Among them, metals and alloys are suitable for structural implants for hard tissues such as artificial joints and denture basal seats, since they possess superior strength and toughness to the other materials. In some promising metallic materials, Co–Cr based alloys have been widely used for the medical devices such as artificial hip and joints.

As-cast Co–Cr–Mo alloys have been applied for the femoral head of artificial joints at present. The compositions of Co–Cr–Mo alloys are standardized in ASTM F75. According to ASTM F75, Ni content in Co–Cr–Mo alloys is allowed to be less than 1 mass%. The Ni addition improves not only castability but also plastic workability of the alloys. However, since the Ni allergy to the human body has been concerned recently, the development of Ni-free Co–Cr–Mo alloys is strongly required. On the other hand, Ni-free Co–Cr–Mo alloys are lack of ductility and workability and it has been quite difficult to control the microstructure by applying thermo-mechanical treatments. Therefore, the development of fine grained Ni-free Co–Cr–Mo alloys becomes a major problem for the improvement of the mechanical properties.

Recently, Chiba et al. have successfully developed forged Co–Cr–Mo alloys without Ni and C additions with a grain size of about 3 μm. The tensile strength of the alloy at ambient temperature shows higher value (1050 MPa) compared to the other Co–Cr–Mo alloys (655 MPa). This alloy is not intended to contain C as well as Ni because of improvement in the forgeability at higher temperatures. When Co–Cr–Mo alloys contain Ni and C, the stacking fault energy is enhanced and thereby the fcc phase (γ phase) is stabilized at ambient temperature. Therefore, in the case of the Co–Cr–Mo alloys without Ni and C additions, γ phase is unstable and in turn hcp phase (ε phase) appears easily by strain induced martensitic transformation.

The fine grained Co–Cr–Mo alloys are expected to display excellent wear resistance. Although the Co–Cr–Mo alloys for biomedical applications should be evaluated in wet conditions, the understanding of the wear behavior in dry conditions would be important for obtaining the basic knowledge on the wear mechanism in wet conditions. The purpose of this study is to examine wear properties of a forged Co–29Cr–6Mo alloy without Ni and C additions under dry condition and to investigate its wear mechanism. Considering the application for medical motivations as artificial joints, it is a worthwhile to investigate the wear resistance of Co–Cr–Mo alloys without Ni and C additions as well as the strength. The friction and wear tests using alumina ball on the alloy will provide the wear mechanism of the alloy and contribute for developing the high wear-resistant Co–Cr–Mo alloys.

2. Experimental Procedure

2.1 Sample preparation

The nominal composition prepared in this study is Co–29 mass%Cr–6 mass%Mo. The alloy ingots were prepared from 99.92 mass%Co, 99.99 mass%Cr, 99.90 mass%Mo by a vacuum induction melting. Prior to forging the ingots, annealing treatment was conducted at 1523 K for 12 h for homogenization and then cooled in the water. The ingots were forged at temperature higher than 1273 K. Total reduction in area of the forged ingot is approximately 62%.

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The chemical compositions of the forged alloys with ASTM F75 and Stellite 6B (the conventional Co–Cr–W–Mo alloy) are listed in Table 1. Note that the Ni and C contents of forged alloy are less than that of the ASTM alloy.

### 2.2 Characterization

Phase identification of the forged alloy was performed by X-ray diffractometer (XRD) using Cr Kα radiation with a collimator of 0.1 mm in diameter. The XRD profiles were collected at least eight times for a specimen and the typical profile was used for the analysis. Quantitative analyses of the volume fractions of fcc and hcp phases in the alloy were carried out with XRD profiles using the method proposed by Sage and Guillaud. The volume fractions were calculated as follows:

\[
\frac{x}{1-x} = \frac{1}{4} \frac{I_{(200)}\gamma}{I_{(101)}\epsilon}
\]

where \(x\) is the volume fraction of fcc phase, \(I_{(200)}\gamma\) and \(I_{(101)}\epsilon\) were the integrated intensities of \((200)\gamma\) and \((101)\epsilon\) peaks, respectively.

Grain size of the alloy was determined metallographically with an optical microscope (OM). Specimens for the observation were polished up to #1200 emery paper and then electrolycally polished at a voltage of 6 V in a solution of 90 parts methanol plus 5 parts sulfuric acid at room temperature.

Wear scars were also analyzed by electron probe microanalyzer (EPMA) and XRD. Hardness of the scar was evaluated by micro Vickers hardness tester at the load of 9.8 N and compared with that of the non-worn specimen.

### 2.3 Wear tests

The disc specimens for wear tests were cut from the forged ingot by an electro-discharge machine (EDM). The size of disc specimens is 30 mm in diameter and 5 mm in thickness. The surface of disc specimens was polished up to the surface roughness (Ra) of less than 0.05 μm. Wear tests were conducted on a conventional ball-on-disc apparatus (RHSCA FRP-2000) running at a constant speed 0.5 m/s under dry conditions and within the load range from 4.9–49 N. The sliding distance is 10^4 m. Alumina balls with the diameter of 4.8 mm were used in the present wear tests. The wear rate, \(\omega\) and wear factor, \(\omega_s\) of the disc were calculated as follows:

\[
\omega = \frac{M_{\text{loss}}}{L \cdot \rho}
\]

\[
\omega_s = \frac{M_{\text{loss}}}{W \cdot L \cdot \rho}
\]

where \(M_{\text{loss}}\) is the weight loss of the disc, \(L\) the sliding distance, \(W\) the contact load and \(\rho\) the density of the disc.

### 3. Results

#### 3.1 Characterization of a forged Co–Cr–Mo alloy

Figure 1 shows the OM micrograph of the forged alloy. The microstructure consists of equiaxed grains and striations are observed in grains. The mean grain size of the forged
alloy was approximately 47 μm.

Figure 2 shows X-ray diffraction profiles of (a) as-forged Co–29Cr–6Mo alloy and (b) the worn scar after wear test at 9.8 N for 20 ks. XRD analysis reveals that the forged alloy consists of γ and ε phase, as shown in Fig. 2(a). The volume fractions of γ and ε phase are 69 vol% and 31 vol%, respectively. Since this alloy is obtained by water quenching, the ε phase is considered to be athermal martensite. Hence, that striations observed in the micrograph are likely the interface of γ/ε or the valiant boundaries of the athermal martensite. On the other hand, the peaks of the ε phase dominantly appeared after wear test, as shown in Fig. 2(b).

3.2 Friction and wear test

Figures 3(a) and (b) show the wear rate and the wear factor of the forged alloy as a function of the contact load, respectively. Those of a Stellite 6B evaluated by Cooper et al.\textsuperscript{14} are also indicated in the figures for comparison. The wear rate of the forged alloy increases gradually with increasing the contact load. The wear rate of the forged alloy is 2 or 3 orders of magnitude lower than that of the Stellite 6B. In addition, the wear factor decreases gradually with increasing in the contact load. The wear rate of the forged alloy is 2 orders of magnitude lower than that of the Stellite 6B. Figure 4 shows the coefficient of friction for the forged alloy as a function of the contact load. The coefficient of friction decreases gradually from 0.34 to 0.29 with increasing the contact load. From these results as shown in Figs. 3 and 4, it can be said that the forged alloy possesses superior wear properties to the Stellite 6B.

3.3 Observation of the wear scar

Figure 5 shows SEM micrographs of the wear scars of the forged alloy: (a) the wear scar, (b) the edge of the wear scar and (c) a step in the wear scar. The arrow in Fig. 5(a) indicates a moving direction of an alumina ball. The various narrow grooves parallel to the moving direction of the ball appear in the wear scar. Slip lines were also observed at edge of the wear scar, as shown in Fig. 5(b). A step across the wear
scar also exists as shown in Fig. 5(c). The step formed in the wear scar appears to consist of several layers (see the arrow). Figure 6 shows (a) a backscattered electron micrograph and X-ray maps of (b) Co, (c) Cr, (d) Mo, (e) Al and (f) O. A large number of elongated debris was embedded in the wear scar.

X-ray maps showed that there was no debris or other products containing Al. These results suggest that the debris is not the adhesion from the alumina ball, but the complex containing O, Cr and Mo. Figures 7(a) and (b) show SEM micrographs of the wear scar at the contact load of 9.8 N and 49 N, respectively. The ratio of dark area to the matrix increases with an increase in contact load. Production of layers on the worn surface can affect the wear rate and the coefficient of friction. More detail on the effects is discussed at the next section.

Figure 8 shows SEM micrograph of the worn surface of an alumina ball. Pt coating was applied onto the alumina ball for the observation. The arrow in Fig. 8 shows the moving direction of the alumina ball. The rough wear scar appeared perpendicular to the moving direction. Figure 9 shows (a) SEM micrograph and X-ray maps of (b) Pt, (c) Al, (d) Co and (e) Cr in the wear scar. X-ray maps showed that there was the area containing Co and Cr on the alumina ball. These results indicate the wear products derived from the forged Co–Cr–Mo alloy could be transferred to the alumina ball during the wear test.

3.4 Changes in hardness after wear tests

Table 2 shows the Vickers hardness of as-polished and the worn surface after wear test. It is clearly found that the hardness of the worn surface is higher than that of the as-polished surface. The increase in hardness after wear test could be attributed to the significantly higher work hardening characteristic at the worn surface, which may be due to the formation of strain-induced martensite, as shown in Fig. 2(b).

4. Discussion

4.1 The stress distribution analyzed on the basis of Hertzian contact theory

In order to discuss the wear conditions of the tests, the stress distributions on the wear surface were estimated using the Hertzian contact theory. Figure 10 shows the distributions of vertical stress \( \sigma_z \), radial stress \( \sigma_r \) and hoop stress \( \sigma_\theta \), and shear stress \( \tau \) on the contact surface calculated by eqs. (4)–(6) as follows:

\[
\sigma_z = \frac{a^2}{a^2 + z^2} P_{\text{max}} \tag{4}
\]

\[
\sigma_r = \sigma_\theta = -\left(1 + \nu_2\right)\frac{z}{a} \left(1 + \frac{a}{2} \frac{z}{\sqrt{a^2 + z^2}} \sin^{-1} \frac{z}{\sqrt{a^2 + z^2}}\right) - \frac{a^2}{2(a^2 + z^2)} P_{\text{max}} \tag{5}
\]

\[
\tau = \frac{|\sigma_r - \sigma_\theta|}{2} \tag{6}
\]

where \( z \) is the depth from the surface, \( a \) the radius of the contact circle, \( S_c \) the area of the contact circle, \( P_{\text{max}} \) the maximum contact stress, \( P \) the mean contact stress. The \( a, S_c, P_{\text{max}} \) and \( P \) is calculated using equations of (7)–(10) as follows:

\[
a = \left(\frac{3}{4} W R \left(1 - \nu_1^2 \frac{E_1 + 1 - \nu_2^2}{E_2}\right)^\frac{1}{2}\right) \tag{7}
\]
where \( W \) is the contact load, \( R \) the radius of the alumina ball, \( \nu_1 \) the poisson ratio of the alumina (0.25), \( E_1 \) the Young’s modulus of the alumina (345 GPa), \( \nu_2 \) the poisson ratio of the forged alloy (0.3), \( E_2 \) the Young’s modulus of the forged alloy (230 GPa).

It should be noted that the shear stress has a maximum at the region of about 0.5\( a \) in depth. The maximum shear stress \( \tau_{\text{max}} \) is expressed by eq. (10) using the \( \tilde{P} \).

\[
\begin{align*}
\frac{1}{2}P_{\text{max}} &= \frac{3W}{2\pi a^2} \\
\tilde{P} &= \frac{W}{\pi a^2} = \frac{2}{3}P_{\text{max}}
\end{align*}
\]

(8) (9) (10)

When the alumina ball contacts the disc specimen, the shear stress at the depth of about 0.5\( a \) is obtained as 0.47\( \tilde{P} \). The shear stress decreases with decreasing \( z \) value from 0.5\( a \) to 0. The shear stress just below the frictional surface (\( z = 0 \)) is expressed as follows:

\[
\tau_{\text{surface}} = \frac{|\sigma_Z - \sigma_I|}{2} \left. \right|_{z=0.0a} \approx 0.087P_{\text{max}} = 0.13\tilde{P}
\]

(11) (12)

Figure 11 shows the shear stress as a function of the contact load at \( z = 0.5, 0.4, 0.1, 0.05 \), and 0, respectively. The forged alloy shows the yield strength of 590 MPa. The maximum shear stress is applied to the yielding condition of
the forged alloy (Tresca’s condition) in the following calculation. When $\tau_{\text{max}}(0.47P)$ reaches 295 MPa, which is a half of 590 MPa, the plastic deformation occurs at the vicinity of $0.5a$ in depth. In the range of the contact load which generates the shear stress less than 295 MPa, elastic deformation occurs. Namely, the complete elastic contact condition is provided. The critical contact load for elastic contact is defined as the upper limit load for complete elastic contact ($W_{\text{UE}}$). When the contact load increases, the plastic deformation also occurs at shallower area than $z = 0.5a$. Moreover, when the shear stress on the frictional surface $\tau_{\text{surface}}(0.13P)$ reaches the 295 MPa, the plastic deformation occurs even at the frictional surface (i.e. complete plastic contact). This load is defined as the lower limit load for complete plastic contact, $W_{\text{LP}}$. From eqs. (11) and (12), $W_{\text{UE}}$ and $W_{\text{LP}}$ are estimated as 1.1 N and 51.5 N, respectively. In this study, the minimum and maximum contact loads were 4.9 N and 49 N, respectively. From the above considerations, elastic and plastic mixed state below the contact surface is provided in the experimental conditions. Hence elastic deformation region exists below the contact surface and the plastic deformation region locally exists at the deeper region.

4.2 Wear mechanism of the forged alloy against the alumina ball

The friction and wear process of the forged alloy against the alumina ball is discussed by considering the elastic contact between them. It was observed that the layer consisting of Co and Cr is transferred to the alumina ball, as shown in Fig. 8. This suggests that adhesive wear occurs between the forged alloy disc and the alumina ball. The real contact surfaces interact and bond with the counterpart in atomic level during the friction and wear process. When the shear adhesive strength is higher than the strength of counterparts, fracture occurs at the area around the contact.

Fig. 7  SEM micrographs of wear scars at the contact load of (a) 9.8 N and (b) 49 N.

Fig. 8  SEM micrograph of worn surface of the alumina ball.
surface. This behavior leads to adhesive wear. Since the shear fracture stress of the forged alloy is lower than that of the alumina, the adhesive wear occurs in the forged alloy. Therefore, wear debris formed by the adhesive fracture of the forged alloy could be transferred, as shown in Fig. 8. In the case of adhesive wear, the wear factor in dry air is generally in the range of $10^{-7}$ to $10^{-8}$ mm$^2$/N. It was revealed from the present wear tests that the wear factor of the forged alloy was $10^{-8}$ mm$^2$/N. Thus it is found that adhesive wear dominantly occurs at the real contact area between the forged alloy disc and the alumina ball. Adhesive wear is divided into two mode; one is mild wear and the other is severe wear, depending on experimental condition such as contact load, sliding speed, testing atmosphere and etc. The wear factor of adhesive severe wear in dry air usually ranges from $10^{-7}$ to $10^{-8}$ mm$^2$/N. Therefore, the wear mechanism of the forged alloy corresponds to severe wear according to the definition based on the magnitude of the wear factor.

The relationship between the contact load and the adhesive wear mode is discussed. Normally, real contact area on the worn surface increases with an increase in contact load. When the contact load increases and reaches the load beginning to start to plastically deform the contact surface of the disc, the plastic deformation region at the real contact part overlaps each other. Accordingly, it can be thought that

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**Table 2** Micro Vickers hardness of the forged Co–Cr–Mo alloy before and after wear test. (Wear condition: Contact Load 9.8 or 49 N, Distance $10^4$ m, Velocity 0.5 m/s.)

<table>
<thead>
<tr>
<th>Sample</th>
<th>Hardness (Hv)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-polished surface</td>
<td>383</td>
</tr>
<tr>
<td>After wear test (Contact load 9.8 N)</td>
<td>543</td>
</tr>
<tr>
<td>(Contact load 49 N)</td>
<td>590</td>
</tr>
</tbody>
</table>

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Fig. 9 SEI micrograph (a) and X-ray maps of (b) Pt, (c) Al, (d) Co and (e) Cr on the worn surface of the alumina ball.
the mean contact stress at the contact real part reaches the
yield stress and the real contact area rapidly approaches the
apparent contact area, resulting in a rapid increase in a wear
rate. Thus it can be concluded that increasing the contact
load, i.e., good enough contact load to plastically deform the
contact surface can make the transition from mild adhesive
wear to severe adhesive wear.

According to eq. (12), the mean contact stress ($P$) of
2270 MPa is calculated for the yielding of the worn surface
by using the eqs. (4), (6) and (7). Thus, when the contact load
is over 51.5 N (i.e. $P = 2270$ MPa), the contact surface shows
plastic yielding and the severe adhesive wear can occur
theoretically. Under the load conditions in this study, $P$ is
2232 MPa at the load of 49 N. Since the value is lower than
2270 MPa, the mild adhesive wear may occur in the forged
alloy. Although the wear factor obtained in the present study
was in the range of severe abrasive wear, the wear factor is
not agreed with that estimated from the stress condition of the
Hertzian contact theory as described above.

It should be discussed why the obtained wear factor shows
large value in spite of the mild wear condition. Wear
mechanisms depend on not only the load but also the sliding
rate and the atmosphere. Therefore, it is necessary to consider
the different wear mechanisms other than the adhesive wear.

Many scratches along the wear direction in the width of a
few microns were observed in the wear scar, as shown in
Fig. 5. In addition, wear debris having the diameter of a few
microns were also observed in the wear scar. These debris
were examined by X-ray maps, thereby the results was
similar to that of elongated debris, as shown in Fig. 6. These
results suggest that the scratches could be formed in the third-
body abrasive wear in association with the alumina ball, wear
debri and the forged alloy disc. The wear factor of abrasive
wear ranges from $10^{-5}$ to $10^{-7}$ mm$^2$/N.

The area ratio of the abrasive scar to the total worn scar is
approximately calculated at 10%. Assuming that the contribu-
tion ratio of the abrasive wear to the total wear is 10%,
the total wear ratio could be estimated by the summation of
10% of the abrasive wear ($0.1 \times 10^{-7}$ mm$^2$/N) and 90% of
the mild adhesive wear ($0.9 \times 10^{-7}$ mm$^2$/N). The estimated
wear ratio is $0.1 \times 10^{-7}$ mm$^2$/N, which is roughly agreed
with the experimental value. Accordingly, the mild adhesive
wear could be the dominant for the forged alloy against the
alumina ball and the abrasive wear caused by the third-body
abrasive wear mediated by the wear debris could be the
secondary mechanism.

As discussed above, the shear stress below the contact
surface exceeds the yield strength of the forged alloy on the
basis of Hertzian contact theory. The cyclic shear stress was
applied to the alloy during the wear tests at the frequency of
9.95 Hz. Thus, fatigue fracture is a potential mechanism at
the plastic deformation region below the surface. As shown
in Fig. 5(c), wear debris and steps were observed, indicating
that the delamination fracture$^{17,18}$ appears to form the steps.
The wear debris could be resulted from the delamination,
however, it needs further investigation to clarify the
formation.

4.3 The characteristic friction and wear process

A wear volume, V is expressed by eq. (13) as follows:

$$V = K \frac{WL}{H}$$

Where, K is the wear coefficient, W is the load, L is the
sliding distance and H is the hardness. The hardness is an
important factor for the wear resistance of materials. The
tribological behavior of the forged alloy is discussed from the
point of the hardness.

4.3.1 Effects of the contact load on the hardness and the
wear factor

The hardness on the worn surface of the forged alloy shows
higher value than that of as-forged alloy, as listed in Table 2.
As shown in Fig. 2(b), the strain-induced martensitic trans-
f ormation occurs during the friction and wear process. The
martensitic transformation is attributed to the low stacking fault energy of the metastable γ phase.\textsuperscript{10} The formation of strain-induced martensite accompanied by the plastic deformation is the major factor for the high work hardening of the alloy. It has been reported that the amount of the strain-induced martensite increases with an increase in contact load and the strain.\textsuperscript{12,20} The amount of the strain-induced martensite increases with an increase in contact load, resulting in the increase of the hardness of worn surface. Hence, the increase of the hardness with the increase of the contact load suppresses the wear rate.

From eq. (13), the wear factor \( \omega_s \) is expressed by eq. (14) as follows:

\[
\omega_s = \frac{V}{WL} = \frac{K}{H}
\]  
(14)

eq. (14) indicates that the wear factor decreases with an increase in the contact load and the hardness. The results as shown in Fig. 3(b) agree with the tendency of the wear factor estimated from eq. (14). Therefore, the wear factor also decreases with the increase of the hardness owing to the formation of the strain-induced martensite.

The wear factor of the forged alloy is lower than that measured by Cooper. The difference between the materials is the content of Ni and C. Since both Ni and C increase the stacking fault energy of γ phase, the martensitic transformation is difficult to occur. Therefore, the hardness of the forged alloy could increase compared to that of the alloys measured by Cooper et al. during the wear test.

4.3.2 The effect of the clearness on the coefficient of friction

The coefficient of friction, \( \mu \), is expressed by eq. (15) as follows:\textsuperscript{21}

\[
\mu = \frac{1}{\sqrt{\alpha(k^2 - 1)}}
\]  
(15)

where \( k \) is the clearness factor (0 < \( k < 1 \)), showing the clearness on the frictional surface and \( \alpha \) is a coefficient. When the \( k \) value increases up to 1, the complete adhesion easily occurs and then the \( \mu \) value increases. On the other hand, the \( \mu \) value decreases with decreasing \( k \) value, since the wear surface is contaminated and then the adhesive strength decreases at the real contact area.

An oxide film containing Cr and Mo is generally formed on the surface of Co–Cr–Mo alloys. The oxide transfer repeatedly occurs with the adhesion occurring between ball and disc during wear process. The large oxide transfer is gradually generated between the ball and disc, resulting in being the complex containing Cr and Mo oxides.\textsuperscript{22} Furthermore the large oxide transfer results in forming the wear debris compressed and elongated with frictional motion in the wear scar.

As shown in Fig. 4, the \( \mu \) value decreases with increasing contact load. The increase of contact load contributes to forming the complex which contains Cr and Mo oxides (Fig. 7). The complex formed on the wear surface as shown in Fig. 6 possibly works as the contamination and is likely to decrease the \( k \) value.

4.3.3 The effect of the hardness of the wear surface

It may be mentioned that the increase of the hardness of the alloy affects the stability of the complex, which contributes lowering the coefficient of friction. When the complex cannot follow the plastic deformation of the forged alloy, the complex is fractured and spalled from the surface and then changed into the wear debris. The alumina may contact the forged metal directly after the fracture. Therefore, the coefficient of wear will not decrease but possibly increase.

The forged alloy showed the increase of the hardness accompanied with the strain-induced martensite transformation. Accordingly, alumina ball could run on the complex due to the work hardening of the forged alloy. Anyhow, the decrease of the coefficient of wear with the increase of the contact load suggests that the work hardened alloy due to strain-induced martensite may work as the stable substrate for the complex formed during the wear tests.

5. Conclusions

A dry wear mechanism and behavior of a forged Co–29Cr–6Mo alloy without Ni and C additions with the low stacking fault energy (hereafter, designated the forged alloy) was investigated using a ball-on-disc type wear testing machine with an alumina ball in dry air. The obtained results are summarized as follow.

1. The wear factor and coefficient of friction of the forged alloy decrease with an increase in a contact load.
2. Since the wear rate and factor of the forged alloy is lower than that of the Stellite 6B, the prominent wear resistance of the forged alloy is demonstrated.
3. The strain-induced martensitic transformation, which is attributed to the low stacking fault energy, occurs below the frictional surface during wear process. The hardness on the worn surface increase concomitantly by the martensitic transformation.
4. From a consideration based on the Hertzian contact theory, it is found that the forged alloy elastically contacts an alumina ball over the entire contact loads tested, resulting in the stress distribution causing mild adhesive wear on the frictional surface. Estimating the contribution of the abrasive wear (the area fraction of the abrasive scars) on the basis of the observation of the wear scar, the calculation of the whole wear factor results in that of the \( 10^{-8} \) mm\(^2\)/N order. The calculate value is correspond to the order of the measured wear factor. Hence, the dominant wear mechanism is mild adhesive wear, though the extrinsic abrasive wear mediated by the wear debris likely coexists.

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