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The concept of effective hardness in the abrasion of coarse two-phase materials with hard second-phase particles

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During the abrasion of a coarse two-phase material which contains hard second-phase particles, a brittleness mechanism can frequently coexist with plastic grooving. By taking $R(X)$ as the abrasion resistance of given material with X as its second-phase volume fraction, the concept of effective hardness, H_{eff}, gives rise to a modification of the linear rule of mixtures as $R(X) \propto (1 - X)H_m + \alpha X H_s$, where H_m and H_s are the hardness of pure matrix material and pure second phase, respectively. The parameter α decreases with increasing severity of the brittleness effect. By defining $\alpha_c \equiv H_m/H_s$, $dR/dX < 0$ provided that $\alpha < \alpha_c$. The SiC abrasion data of a series of Al–Si alloys (with the volume fractions of pro-eutectic Si particles ranging from 0.023 to 0.219) can be rationalized by the above model since $R(X)$ shows a nice linear fit with X. The α_c value of the test alloys is close to zero. Therefore, as indicated from the wear surface, extensive brittle fracture can occur without deteriorating the abrasion resistance. For those with $dR/dX < 0$, subsurface fracture is also pronounced.

1. Introduction

For a coarse two-phase material which contains hard second-phase particles with X as the second-phase volume fraction, Khruschov's rule of mixtures [1–3]

$$R(X) = (1 - X)R_m + XR_s \tag{1}$$

is frequently adopted as an approximation of its abrasion resistance $R(X)$, where R_m and R_s are the abrasion resistance of pure matrix material and pure second phase, respectively. Strictly, Equation 1 is used to model a plastic grooving process in which the second-phase particles are more groove-resisting than the matrix phase. This linear equation predicts a positive rate of change of $R(X)$ with respect to X, $dR(X)/dX$, equal to $R_s - R_m$.

Interface decohesion as well as brittle cracking and breaking of the second-phase particles usually coexist with plastic grooving during an abrasion process. These brittleness factors effectively reduce the role of second-phase particles in resisting grooving. Therefore, as indicated by the wear data of many systems including high-chromium white cast iron [2] and some hard alloys containing tungsten carbide [1], Khruschov's rule normally overestimates the abrasion resistance.

To modify Khruschov's rule, Zum Gahr [2] proposed that

$$R(X) = (1 - X)R_m + XR_s - X(1 - X)(R_m + R_s) \tag{2}$$

This non-linear equation, which correlates well with the high-chromium white cast-iron data reported by Zum Gahr, predicts a concave upward behaviour of $R(X)$ with respect to X, i.e. $R(X)$ decreases with increasing X until a minimum at $X = R_m/(R_m + R_s)$ and then increases. However, as mentioned by Moore [4], many abrasion data show a concave downward behaviour in which the transition from positive to negative $dR(X)/dX$ with increasing X is due to an increasing contribution of brittleness effect relative to plastic grooving.

Regardless of whether $R(X)$ is linearly proportional to X or not, a linear relation can always be adopted provided that the range of X is properly chosen. As a substitute for Khruschov's rule, the following linear equation was recently proposed by Chen *et al.* [5]:

$$R(X) = (1 - X)R_m + kXR_s \tag{3}$$

where the factor k was imposed to take into account the brittleness effect, i.e. $k \leq 1$ with $k = 1$ for pure plastic grooving. According to Equation 3, $dR(X)/dX$ equals $kR_s - R_m$, which is smaller than that associated with pure plastic grooving, $R_s - R_m$, when a brittleness mechanism is incorporated. The data of some high-chromium white cast irons [5] revealed that $kR_s - R_m$ and thus k are reduced by increasing the severity of the brittleness mechanism involved in the process, and the severity of the mechanism can be varied by changing the load-carrying capacity of the second-phase particles through the control of (for instance) particle size or matrix property.

Garrison [3] stated that Khruschov's rule can be adopted only when the wear rate is proportional to the applied load. As an implication, the severity of the

0022–2461

brittleness mechanism involved in an abrasion system can also be varied by changing the loading conditions. When an enhanced brittleness effect occurs under severe loading conditions, k is sufficiently small, which may lead to $kR_s < R_m$ and thus $dR(X)/dX < 0$ according to Equation 3. In this situation, the existence of second-phase particles is not beneficial in improving the abrasion resistance.

The purpose of the current investigation is to analyse the extent of brittleness mechanism while an abrasion process can tolerate while maintaining $dR(X)/dX > 0$. A model such as that given in Equation 2 is insufficient for the analysis since it is limited by considering the parameters X, R_m and R_s only. Bearing in mind the restriction of assuming linearity, the approach used in deriving Equation 3 will be applied in the following by introducing a concept of effective hardness. The data for a series of hypereutectic Al–Si alloys with various pro-eutectic Si contents will be reported for this analysis.

2. Theoretical considerations

A constancy of wear-surface profile is usually retained for a coarse two-phase material under steady-state conditions, as illustrated schematically in Fig. 1. The constancy indicates a global behaviour that the rate of material removal is the same for the two phases regardless of their difference in abrasion characteristic. For a pure plastic grooving process this global behaviour can be rationalized by assuming, as in the case of single-phase material [6, 7], that the abrasion resistance is proportional to the hardness, $H(X)$. According to the rule of mixtures, $H(X)$ can be expressed as

$$H(X) = (1 - X)H_m + XH_s \qquad (4)$$

Where H_m and H_s are the hardness of the matrix material and the second-phase material, respectively. As a result

$$R(X) \propto (1 - X)H_m + XH_s \qquad (5)$$

If the presence of second-phase particles leads to a brittleness mechanism in addition to plastic grooving, $R(X)$ is effectively reduced. As in Equation 3 which takes a brittleness effect into consideration, we propose the concept of effective hardness, $H_{eff}(X)$, to modify Equation 5 as

$$R(X) \propto H_{eff}(X) \equiv (1 - X)H_m + \alpha XH_s \qquad (6)$$

Where $\alpha = 1$ under pure plastic grooving, and $0 < \alpha < 1$ when a brittleness mechanism also occurs. If the second-phase particles are extremely brittle or the interface strength is extremely low, α tends to zero provided that the frictional work associated with particle breaking or interface decohesion is negligible.

To handle the effect of brittleness, the fracture toughness should also be considered besides hardness [8–10]. A wear equation which contains these parameters explicitly is already complex in the case of single-phase material [8]. For coarse two-phase material, Equation 6 simplifies the problem by including them implicitly in the empirical parameter α. This equation can be written alternatively as

$$R(X) \propto H_m + (\alpha H_s - H_m)X \qquad (7)$$

which indicates that the second-phase particles have a positive contribution in improving the abrasion resistance if $\alpha H_s - H_m > 0$, i.e. if $\alpha > H_m/H_s$. By defining $\alpha_c \equiv H_m/H_s$ as a critical ratio, the second-phase particles are beneficial provided that $\alpha > \alpha_c$. Moreover, a smaller α_c, which is associated with higher H_s and lower H_m, implies a larger brittleness effect than the abrasion process can tolerate.

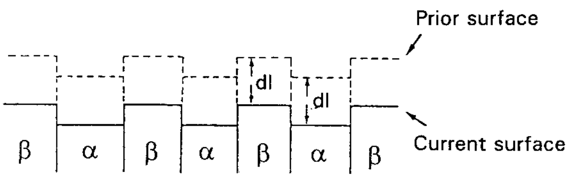

Figure 1 Schematic diagram showing that the two constituents of a composite material are worn at equal rates along the wear-surface normal direction.

3. Experimental procedure

Seven hypereutectic Al–Si alloys were used in this investigation. Table I lists the composition of the alloys determined by emission spectrometry. Also listed in the table are the respective volume fractions of pro-eutectic Si particles, X_{Si}, measured by the line interception method. As indicated, the silicon content varies from 12.6 to 30.5 wt %, whereas X_{Si} varies from 0.023 to 0.219. For illustration, Fig. 2 presents optical micrographs of alloys with $X_{Si} = 0.07$ and $X_{Si} = 0.219$.

A pin-on-disc tribometer was used for the abrasion test. The pin was an Al–Si specimen, with its 3 mm × 3 mm end surface contacting with a commercial SiC polishing paper which was fixed on a rotating disc. With 12 r.p.m. rotation speed and 55 mm contact radius from the disc centre to the pin centre, the relative sliding speed was fixed at $70\ \mathrm{mm\,s^{-1}}$. To achieve various loading conditions, three normal loads of 1.96, 5.88 and 10.8 N were selected while SiC papers of grit numbers 150, 280, 500 and 1000 were used.

To avoid any breaking-in period, each specimen was pre-polished under its test condition before testing. In order to accumulate enough weight loss for the

TABLE I Chemical composition and pro-eutectic Si volume fraction, X_{Si}, of the Al–Si alloys

	Composition (wt %)				
X_{Si}	Si	Cu	Fe	Zn	Ni
0.023	12.6	0.02	0.04	0.03	0.01
0.044	14.2	0.06	0.19	–	0.01
0.051	15.6	–	0.23	0.01	0.01
0.070	17.2	–	0.26	0.02	0.01
0.115	21.1	–	0.39	0.02	0.01
0.170	25.3	0.01	0.44	0.05	0.01
0.219	30.5	0.02	0.44	0.05	0.01

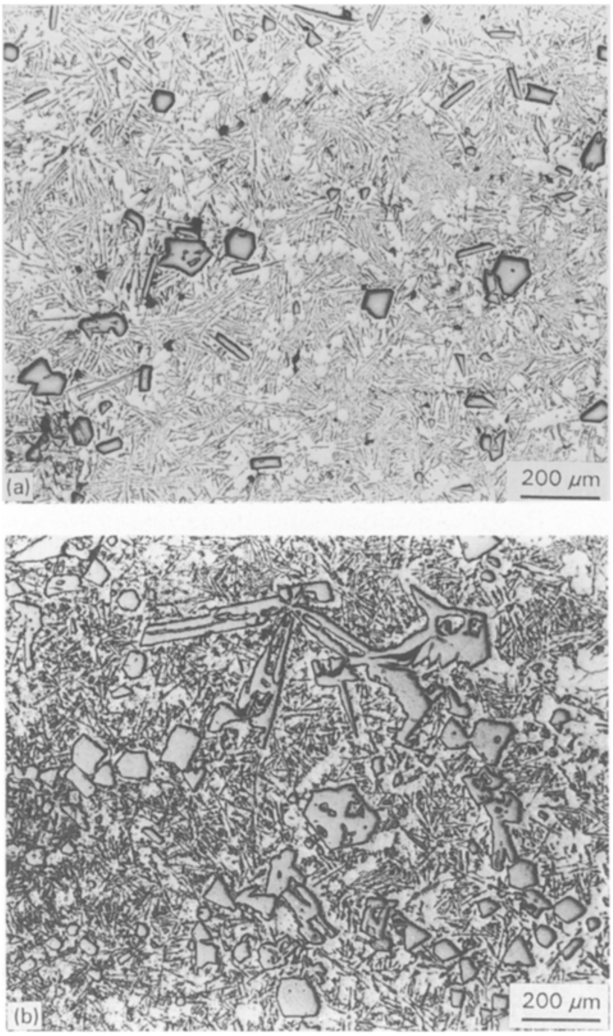

Figure 2 Metallography of test alloys with (a) $X_{Si} = 0.07$ and (b) $X_{Si} = 0.219$.

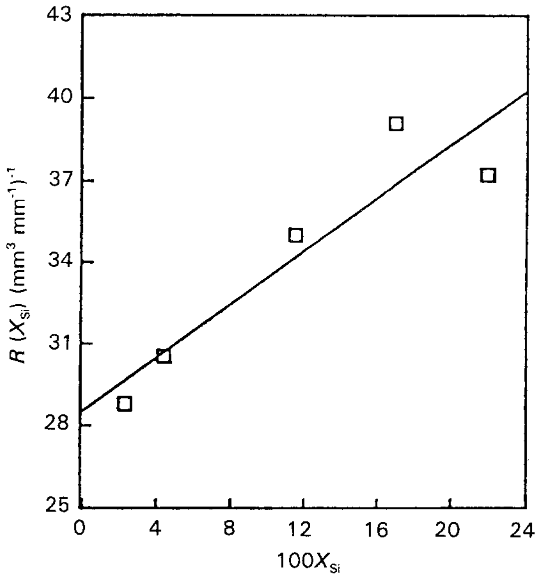

Figure 3 Abrasion resistance $R(X_{Si})$ versus the volume fraction of pro-eutectic Si, X_{Si}, for tests with 1000 grit SiC and 1.96 N normal load.

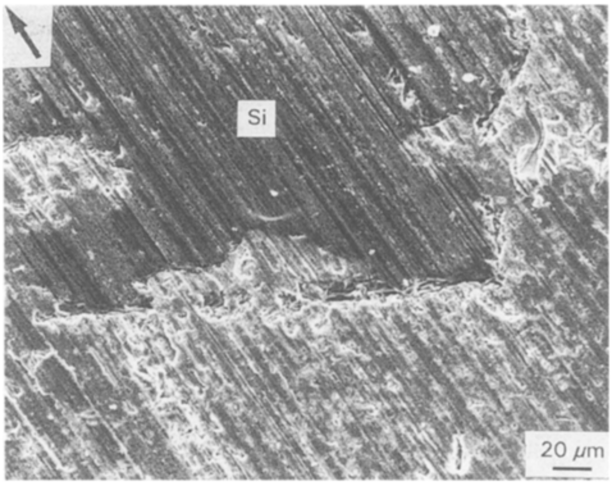

Figure 4 Typical wear-surface morphology of tests with 1000 grit SiC and 1.96 N normal load: a close-up indicating plastic grooving as the predominant mechanism on Si particles. The arrow in this figure and hereafter indicate the sliding direction of the SiC grits.

measurement, which was done by using a microbalance of precision 0.1 mg, the tests with 100 and 280 grit SiC papers were terminated after five rotation cycles. Those with 600 grit used two fresh SiC papers, each for ten cycles. Those with 1000 grit also used two fresh SiC papers, each for 15 cycles. All weight measurements were performed after cleaning the specimens ultrasonically with acetone. The weight loss datum of each test condition was averaged from at least eight runs. Its associated abrasion resistance was obtained by converting this datum to the inverse of volume loss per unit sliding distance. For the conversion, a density of 2.69 g cm^{-3} was adopted for the eutectic matrix, and 2.33 g cm^{-3} for pro-eutectic Si [11].

4. Results

For the tests with 1000 grit SiC and 1.96 N normal load, which correspond to the lightest loading condition of this study, the abrasion resistance $R(X_{Si})$ is plotted against the volume fraction of pro-eutectic Si, X_{Si} in Fig. 3. The figure shows a tendency of increasing $R(X_{Si})$ with X_{Si}. As illustrated in Fig. 4, the wear surface reveals mainly plastic grooving in which the pro-eutectic Si particles are more grooving-resistant than the matrix phase. Owing to the presence of fine eutectic particles, cracks coexist with surface grooves on the matrix phase. Cracks also exist at the pro-eutectic Si particle–matrix interface.

When the abrasion condition becomes more severe, i.e. 1.96 N normal load with 600, 280 and 150 grit SiC, Fig. 5 reveals a transition of increasing $R(X_{Si})$ with X_{Si} to essentially constant $R(X_{Si})$. Fig. 6 depicts extensive brittle fracture of the pro-eutectic Si particles. The matrix phase is subjected to a larger deformation than that shown in Fig. 4.

Under the two most severe test conditions of this study, namely 150 grit SiC with 5.88 N and 10.8 N normal loads, as indicated in Fig. 7, $R(X_{Si})$ decreases with increasing X_{Si}. Like those illustrated in Fig. 6, these tests also lead to severe brittle cracking of the pro-eutectic Si particles (e.g. Fig. 8). Unlike other test conditions (e.g. 280 grit SiC and 1.96 N normal load,

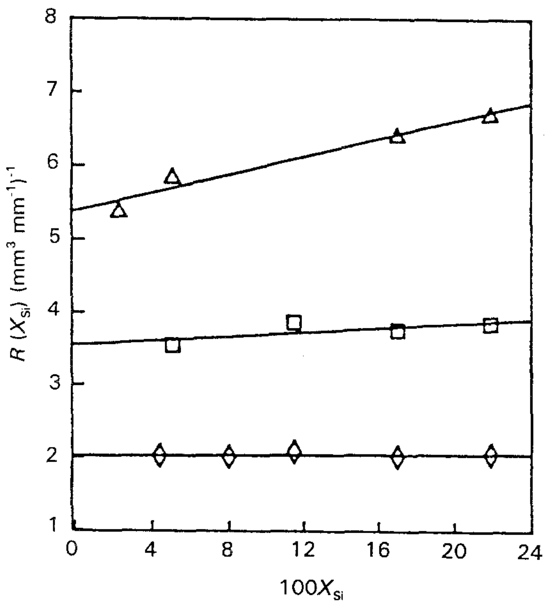

Figure 5 $R(X_{Si})$ versus X_{Si} for the tests with 1.96 N normal load and with SiC grit numbers (△) 600, (□) 280 and (◇) 150.

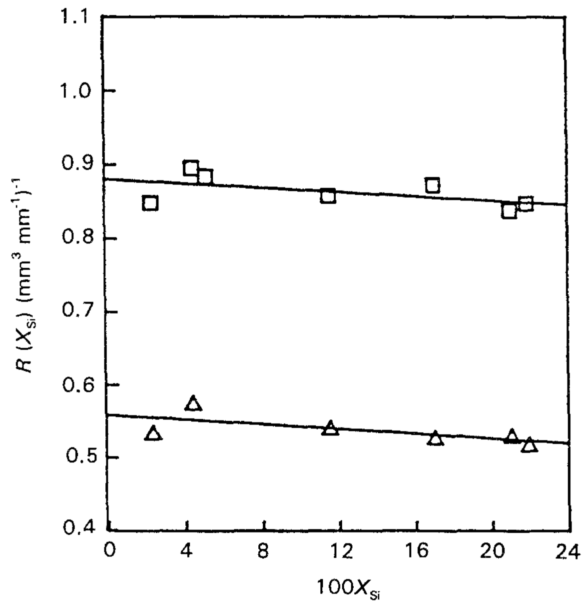

Figure 7 $R(X_{Si})$ versus X_{Si} for tests with 150 grit SiC, (□) 5.88 N and (△) 10.8 N normal loads.

Fig. 9a, those of Fig. 9b clearly lead to breaking and bending of Si particles along the sliding direction in the subsurface region.

The Vickers hardness number of the pro-eutectic Si particles is 1300 ± 60 VHN, averaged from ten in-

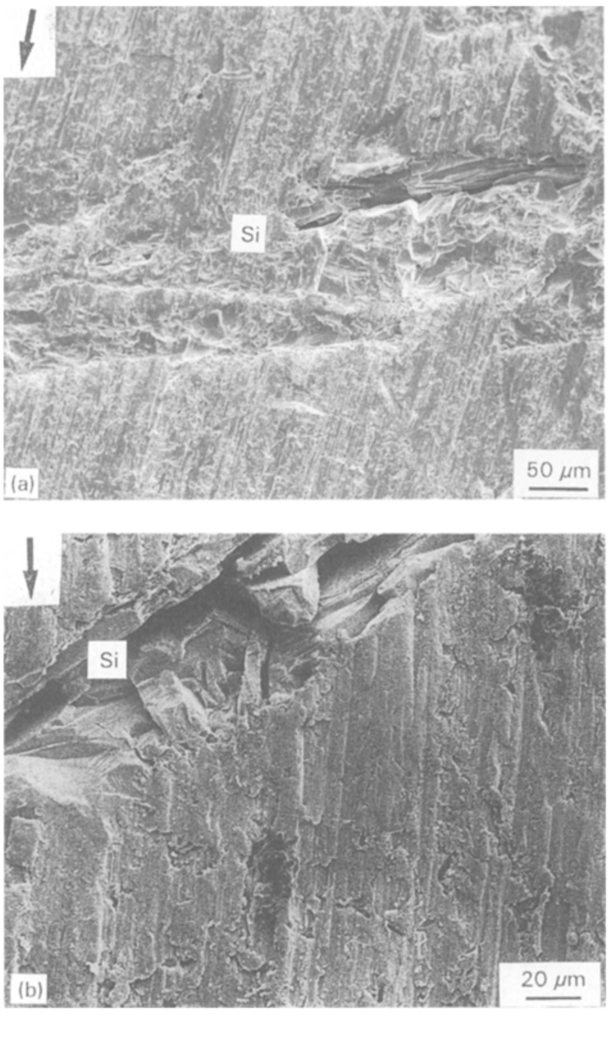

Figure 6 Wear-surface observation: (a) 0.219 X_{Si}, 1.96 N normal load, 600 grit SiC; (b) 0.17 X_{Si}, 1.96 N normal load, 150 grit SiC.

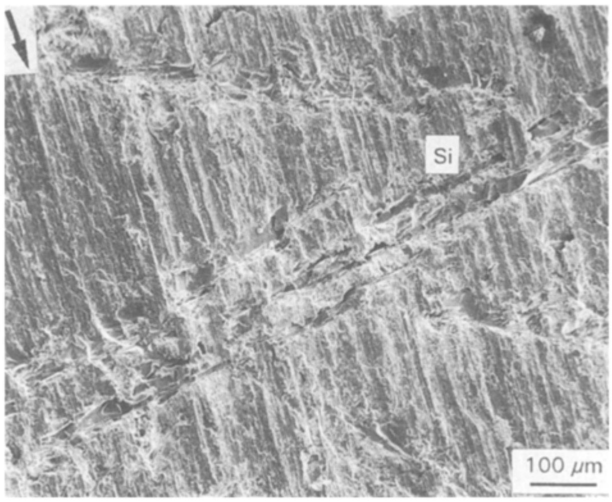

Figure 8 Wear-surface observation on the specimen with 0.219 X_{Si} (test conditions: 150 grit SiC 10.8 N normal load).

dentation tests. The work-hardened matrix phase possesses an average hardness of 60 ± 5 VHN, measured from a specimen with $X_{Si} = 0.023$ after abrasion under 150 grit SiC and 10.8 N normal load. To obtain this average, the specimen was tapered and 30 indentation tests were done on the near-surface region within about 20 μm depth. All the above indentations were made under 25 g load.

5. Discussion

Since pro-eutectic Si particles are more grooving-resistant, Fig. 4 reveals a protrusion of the particles above the eutectic matrix. This phenomenon is the same as that sketched in Fig. 1 in which a constant wear-surface profile was implicated. As a consequence, the modification of Khruschov's rule given in Equation 3 can be transformed to Equation 6, i.e. $R(X_{Si}) \propto H_{eff} \equiv (1 - X_{Si})H_m + \alpha X_{Si} H_s$, where $\alpha < 1$

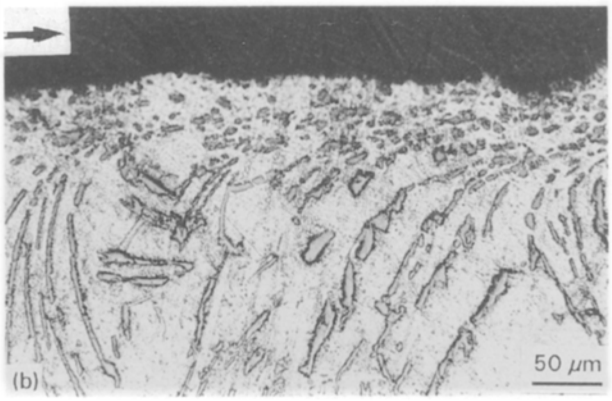

Figure 9 Longitudinal cross-sections of specimens with 0.023 X_{Si}, tested under (a) 280 grit SiC and 1.96 N normal load, (b) 150 grit SiC and 10.8 N normal load.

owing to brittle effect. The two equations can both be adopted for analysing the abrasion resistance data presented in Fig. 3, 5 and 7 since these data show a nice near fit. However, Equation 6 allows a further step to predict the transition stage of positive to negative $dR(X_{Si})/dX_{Si}$ by setting $\alpha = \alpha_c \equiv H_m/H_s$ at this stage.

With $H_m = 60$ VHN and $H_s = 1300$ VHN as reported in the previous section, α_c is about 0.05, which is rather small compared to $\alpha = 1$ for pure plastic grooving. Therefore, according to the model interpreted in Equation 6 (or Equation 7), the Al–Si test alloys can endure a substantial brittleness effect yet $R(X_{Si})$ still increases with X_{Si}.

As can be seen from Fig. 5, the abrasion tests with 1.96 N normal load and 280 and 150 grit SiC roughly correspond to the critical condition in which $R(X_{Si})$ is independent of X_{Si}, whereas the same normal load condition with 600 grit SiC gives rise to a tendency of increasing $R(X_{Si})$ with X_{Si}. Under these test conditions, just like those with negative $dR(X_{Si})/dX_{Si}$ (Fig. 8), the wear-surface morphology depicts an extensive brittle fracture of the pro-eutectic Si particles (e.g. Fig. 6). In fact, it is practically impossible to foresee whether $dR(X_{Si})/dX_{Si}$ is positive or negative merely from observing the severity of brittle cracking on the wear surface. According to the concept mentioned in the previous paragraph, this is because the α_c value of the Al–Si test alloys is very small, which gives rise to an enhanced brittle fracture even with positive $dR(X_{Si})/dX_{Si}$.

To characterize the nature of $dR(X_{Si})/dX_{Si}$, subsurface observation is effective. As indicated in Fig. 9, $dR(X_{Si})/dX_{Si} < 0$ provided that apparent subsurface fracture is discernible.

6. Conclusions

The following conclusions are drawn based on this investigation:

(a) During steady-state plastic grooving with constant wear-surface profile, the linear rule of mixtures leads to $R(X) \propto (1 - X)H_m + XH_s$. In the case of the Al–Si test alloys, the constant wear-surface profile depicts a protrusion of pro-eutectic Si particles above the eutectic matrix.

(b) For those involving both ductile and brittle mechanisms, the modified equation $R(X) \propto H_{eff} \equiv (1 - X)H_m + \alpha X H_s$ can be adopted, where $1 > \alpha > 0$. The parameter α decreases with increasing severity of brittleness effect, leading to a reduced effective hardness H_{eff}, and hence a reduced $R(X)$.

(c) According to the above modified equation, $R(X)$ is linearly proportional to X, where $dR(X)/dX \propto \alpha H_s - H_m$. The Al–Si alloys with X_{Si} ranging from 0.023 to 0.219 are in good agreement with this linearity under the test conditions of this study.

(d) The modified equation predicts a zero $dR(X)/dX$ when $\alpha = \alpha_c \equiv H_m/H_s$. For the Al–Si test alloys, α_c is about 0.05. This α_c value is rather small, implying that the test alloys can tolerate a substantial brittleness effect while $dR(X_{Si})/dX_{Si}$ remains positive. In fact, since extensive brittle fracture can occur, it is hard to discern from the wear surface whether $dR(X_{Si})/dX_{Si}$ is positive or negative.

(e) For materials with a negative $dR(X_{Si})/dX_{Si}$, brittle cracking of the pro-eutectic Si particles is apparent in the subsurface region.

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