Wear of ceramic particle-reinforced metal-matrix composites

Part II *A model of adhesive wear*

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Wear experiments have shown that when the applied normal stress exceeds a critical value, a transition occurs from the regime of mild wear to that of adhesive wear. As a result, the wear rate of particle-reinforced aluminium composites increases by a hundred-fold. Based on dislocation and delamination theories, a criterion is proposed for determining the critical transition load, and a quantitative model is developed for adhesive wear in a wear system of a steel disc sliding against aluminium matrix composite pins. Experimental results of four kinds of metal-matrix composites confirm the validity of the criterion and the model.

1. Introduction

The mechanisms of abrasion and adhesion control most wear processes. Bowden and Tabor [1] discussed the adhesion theory of friction which attributes friction to the formation, shear and failure of adhesive junctions of surface asperities on a pair of rubbing surfaces. However, little attention was paid to the metallurgical structure of materials involved. Suh [2] developed a delamination theory of wear and introduced a simple dislocation model for the soft surface layer to replace the adhesion theory. Since then extensive wok has been done to confirm the delamination behaviour [3–7] and define the transition from mild to severe wear [5–7]. However, to date, neither a criterion for transition load, nor a practical model for describing adhesive wear, has been developed. This is particularly lacking in the case of advanced materials, such as particle-reinforced metal-matrix composites. In Part I of this study [7], the wear mechanisms in several ceramic-reinforced aluminium matrix composites have been studied. It is observed that a transition of mild wear to severe wear occurs when the applied normal load reaches a certain critical value. This transition behaviour is both load- and microstructure-dependent. In Part II we study the role of the critical transition load and develop a quantitative model to predict adhesive wear which includes the microstructure of the materials.

2. Experimental procedure

Wear tests were carried out on a Plint-Cameron pin-on-disk machine. Detailed descriptions of the machine were given in Part I [7]. Pins made from four kinds of 6061–Al matrix composites reinforced with either SiC or Al_2O_3 particles with volume fractions of 10 % and 20 %, respectively, were studied. All the materials were heat treated to T6 condition, i.e. solution treatment at 530 °C for 1.5 h, quenching into water followed by natural ageing for 20 h and then artificial ageing at 175 °C for 8 h. The mean diameter of the particles and Vickers microhardness of the composites measured with an indentation load of 5 N are listed in Table I of Part I [7].

Transitions from mild to adhesive wear for all the materials were observed when the applied stresses exceeded the corresponding critical values. The transition stress increases with increasing volume fraction and particle size as shown in Fig. 1. Because the relationship between the volumetric wear and sliding distance is linear, the wear rate could be defined as volumetric wear per sliding distance. The correlation between the wear rate and applied normal stress is shown in Fig. 2 for 20 vol % SiC-reinforced composites, in the adhesive wear regime. The determination of the critical transition load will be discussed in Section 3.

The debris produced in the adhesive wear regime was in the form of flake-like thin sheets as shown in Fig. 3. These wear sheets were identified to come from the metal-matrix composite pin by energy dispersive X-ray analysis (EDAX) [7]. However, the steel disc was always covered by pieces of adhered composite materials. Therefore, it would be reasonable to assume that no wear had occurred on the disc surface.

The dimensions of the wear sheets were measured by using optical microscopy and image analysis. Their

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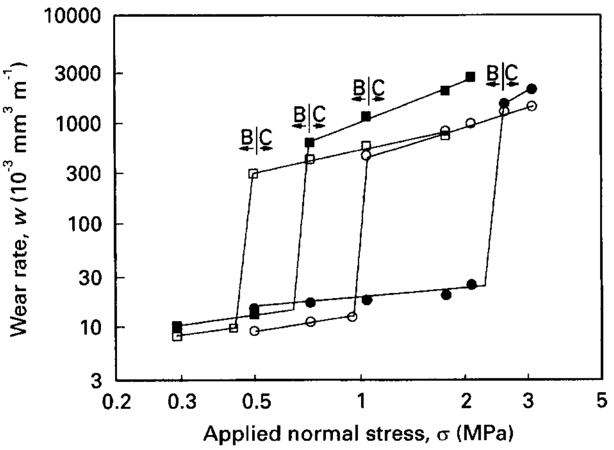

Figure 1 Transitional behaviour of four metal-matrix composites: (○) 10% Al_2O_3, (●) 20% Al_2O_3, (□) 10% SiC, (■) 20% SiC, B|C represents the division of the mild and severe wear regimes.

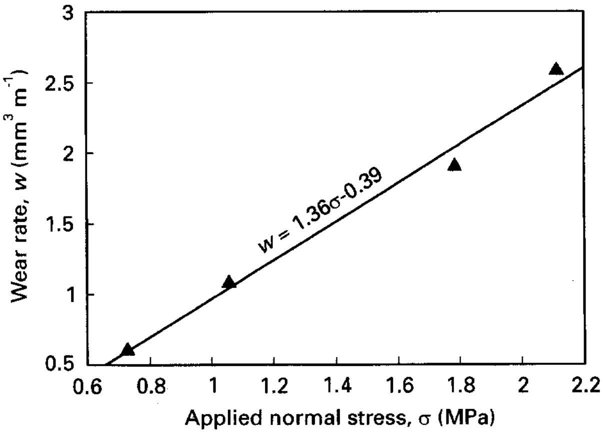

Figure 2 Relationship between wear rate, *w*, and applied normal stress, σ, for 20% SiC–Al composite.

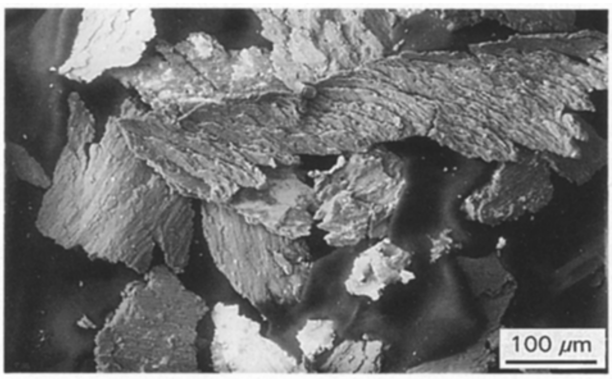

Figure 3 Flake-like wear sheets of 20% SiC–Al composite at a normal stress of 2 MPa.

sizes were from $50 \times 20 \times 4$ to $3000 \times 500 \times 70\ \mu m^3$ and an average wear sheet dimension was $600 \times 200 \times 8\ \mu m^3$ for 20 vol% SiC-reinforced composites when the wear tests were conducted under a normal stress of 2 MPa. The thickness of the sheets was found to increase with increasing normal stress. In addition, fine SiC–Al composites resulted in thinner wear sheets than coarse Al_2O_3–Al composites.

3. Problems in theoretical modelling

The adhesion theory proposed by Archard [8] was based on multiple asperities in contact. A wear equation is given by

$$V = KLS/3H \quad (1)$$

where V is the wear volume, L the normal load, S the sliding distance and H is the Vickers hardness of the surface. This equation states that wear rate is proportional to the inverse of the hardness of the worn material. However, it is not consistent with our experimental results with metal-matrix composites (see Fig. 1).

The delamination theory suggested by Suh [2] was based on extensive experimental observations on many different materials. It assumes that materials would wear layer by layer and each layer consists of N pieces of flaky sheets. To remove such a layer, a critical sliding distance, S_0, is needed. The wear volume, V, can then be calculated by

$$V = N_1(S/S_{01})A_1h_1 + N_2(S/S_{02})A_2h_2 \quad (2)$$

where h is the average thickness of delaminated sheets and A is the average area of the sheets. The subscripts 1 and 2 refer to the two materials sliding against each other. A schematic view of the material removal is shown in Fig. 4. For a metal without large hard particles, h can be obtained from dislocation theory [9]

$$h = \frac{Gb}{4\pi(1-\nu)\sigma_f} \quad (3)$$

where G is the shear modulus, σ_f the friction stress, $\boldsymbol{b}$ the Burgers vector and ν is Poisson's ratio. According to Gupta and Cook [10], the following relationship exists

$$NA \propto L \quad (4)$$

The wear volume can be obtained by substituting Equations 3 and 4 into 2

$$V = \frac{\boldsymbol{b}}{4\pi}\left[\frac{K_1G_1}{\sigma_{f1}S_{01}(1-\nu_1)} + \frac{K_2G_2}{\sigma_{f2}S_{02}(1-\nu_2)}\right]LS \quad (5)$$

where K_1 and K_2 are constants depending primarily on the surface topography.

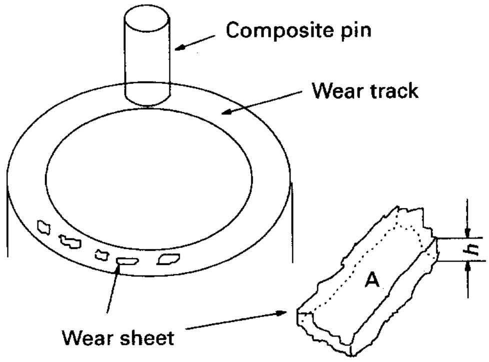

Figure 4 A schematic view of the material removal process in pin-on-disc wear tests.

The advantage of this model is that it does not depend directly on hardness. Unfortunately, it is not related to the microstructure of materials and hence cannot predict wear rate quantitatively. In addition, the assumptions used for deriving Equation 5 are not appropriate for particle-reinforced MMCs.

4. Determination of the critical transition load

Subsurface shear cracking has been observed by many authors. Wang and Rack [6] indicated that when sliding occurred in the severe wear regime, isolated voids were linked into cracks below the surface. These cracks were observed on a cross-section of a composite reinforced with SiC particles worn in the severe wear regime.

Rosenfield [11] assumed in his shear instability model that there was no pre-existing crack in the material. However, a thin crack-shaped local region whose flow strength is lower than that of the surrounding material, is postulated. The model gives rise to the stress intensity, K_{II}, as

$$K_{II} = \frac{-2F}{\pi^{3/2}a^{1/2}} I(k, p, \mu) \quad (6)$$

where F is the normal force per unit thickness and a is the half length of the shear crack. I is the geometry-dependent integral which depends on parameters $p(= y/a$ the normalized depth of the crack plane below the surface), k (the offset of the centre of the crack from a point beneath the applied force) and μ (the friction coefficient between the rubbing surfaces; see Fig. 5). Here I refers to the location at $k = 0.4$, where the maximum stress intensity occurs [11].

The relationship between I and p for $k = 0.4$ with three different friction coefficients is shown in Fig. 6. It indicates that the driving force peaks at a distance below the surface (when $y/a = 1$). Hutchings [12] also pointed out that the maximum shear stress would occur at a certain depth below the surface according to the Hertzian analysis of the elastic stress field due to a spherical indenter on a flat surface. Both implied that delamination would most likely happen for the materials subjected to repeated sliding forces.

The material resistance to shear on the crack plane can be expressed in terms of the stress intensity as

$$K_c = S_f \pi^{1/2} a^{1/2} \quad (7)$$

where S_f is the subsurface flow stress of the softer material. The instability condition is $K_{II} + K_c = 0$.

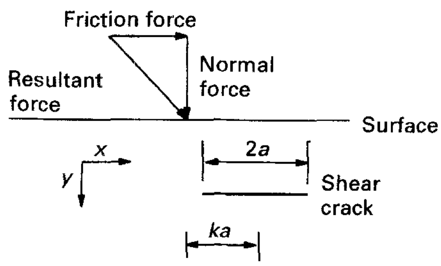

Figure 5 Geometry of the point force model (from Rosenfield [11]).

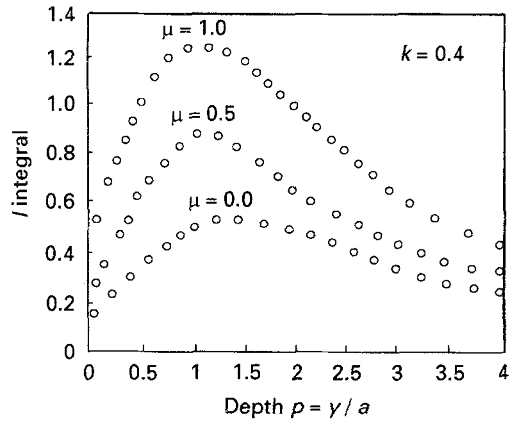

Figure 6 Relationship between I integral and friction coefficient, μ, and normalized depth, p, for $k = 0.4$.

The criterion of transition load can be obtained from

$$F_c = \frac{\pi^2 a S_f}{2I} \quad (8)$$

The variation of flow strength, S_f, with depth is not well established for particle-reinforced composites. Assuming the flow strength is proportional to the hardness, and the crack length, $2a$, is proportional to the particle size, d, the criterion for the transition load (stress) can be expressed by:

$$\sigma_c = \frac{q\pi^2 H_v D}{4AI} \quad (9)$$

where q is a factor which can be determined from experiments, A is the nominal contact area. This criterion is associated with the particle size, d, the hardness, H_v and I value. The peak I value can be obtained from Fig. 6. At this location, the normal load which leads to the delamination is minimum, and I is dependent on the friction coefficient. It is therefore practically applicable to particle-reinforced metal-matrix composites. The factor q should be different for SiC–Al and Al_2O_3–Al composites because of their different physical and mechanical properties. To determine q, take 10% Al_2O_3–Al and 10% SiC–Al composites as examples. The measured parameters are: (1) for 10% Al_2O_3–Al composite, $d = 4.5 \times 10^{-3}$ mm, $A = 16\pi$ mm^2, $H_v = 130$ kgf mm^{-2} and $\sigma_c = 1.06$ MPa; and (2) for 10% SiC–Al composite, $d = 1.8 \times 10^{-3}$ mm, $A = 16\pi$ mm^2, $H_v = 130$ kgf mm^{-2} and $\sigma_c = 0.53$ MPa. The friction coefficient in the regime of adhesive wear was about 0.5. The I value always peaks at $p = y/a = 1$. Therefore a peak value of $I = 0.9$ was taken for both Al_2O_3–Al and SiC–Al composites from Fig. 6. Thus q is determined to be 33 for Al_2O_3–Al and 41.6 for SiC composites, respectively. Using these q values and the appropriate values for H_v, d, A and I the transition loads (stresses) of the other two composites can be predicted from Equation 9 as 0.612 MPa (with 20% SiC) and 2.30 MPa (with 20% Al_2O_3). The experimental results of these composites are 0.79 and 2.65 MPa, respectively, which compare favourably with the theoretical estimates.

5. A new adhesive wear model

Because the experimental observations are consistent with the delamination theory. It is now used to predict the adhesive wear of particle-reinforced metal-matrix composites. A schematic illustration of the delaminated wear sheets is shown in Fig. 7. Because MMCs differ from metals and their alloys in many ways, the determination of some parameters, such as the thickness of the wear sheet (Equation 3) is inappropriate for the present composites. In addition, because the wear of the steel disc can be neglected in the adhesive wear regime, the second term in Equation 2 does not exist. Therefore, the wear of the composite pin would be similar to Equation 5

$$V = c_1 LS \tag{10}$$

where c_1 is dependent upon material properties and would vary whenever the pin material is changed. This is not practical, and modification of Equation 10 is needed.

In the adhesive wear regime, wear is controlled by propagation of subsurface cracks, thus the volumetric wear must relate to the resistance of the matrix deformation. The smaller the matrix area, the smaller is the resistance to cracking of the composites. Thus c_1 is assumed to be proportional to the inverse of the mean free path, λ, of the particles. According to Underwood [13], λ can be determined from the particle volume fraction, f_v, and average particle diameter, d. Equation 10 can be rewritten as

$$\lambda = d\frac{(1-f_v)}{f_v} \tag{11}$$

$$V = \frac{cLS}{\lambda} = \frac{cLSf_v}{(1-f_v)d} \tag{12}$$

where $c = c_1\lambda$ can be obtained by experiments on particle-reinforced aluminium matrix composites. Now c in Equation 12 is independent of particle size and volume fraction, which makes the formula useful for any given MMCs with a given type of matrix and particle materials. Equation 12 states that the wear volume is not only proportional to the sliding distance and applied load, but also increases with the particle volume fraction and decreases with particle size.

Because of the differences in both interfacial bond strength and particle properties of the SiC and Al_2O_3 reinforcement, wear varies accordingly. The factor c therefore differs as well. From wear results (Fig. 1) and the other known parameters of the composites, i.e. $V/S = 0.56\ mm^3\,m^{-1}$, $d = 1.8\times10^{-3}$ mm, $f_v = 0.1$ and $L = 106$ N, c is 0.086×10^{-3} for SiC–Al composites while for Al_2O_3–Al composite, c is 0.18 (calculated from the data of 10 % Al_2O_3–Al composite: $V/S = 1.2\ mm^3\,m^{-1}$, $d = 4.5\times10^{-3}$ mm, $f_v = 0.1$ and $L = 265$ N). Using Equation 12 and relevant values of c, L, f_v and d the wear rates are $1.26\ mm^3\,m^{-1}$ for 20 % SiC–Al and $1.37\ mm^3\,m^{-1}$ for 20 % Al_2O_3–Al composite, respectively. These predicted results are in good agreement with the corresponding experimental values of 1.08 and $1.42\ mm^3\,m^{-1}$.

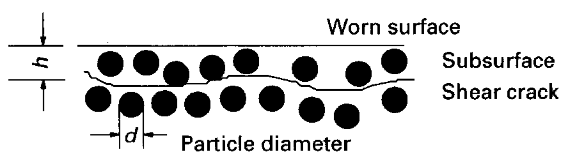

Figure 7 Schematic illustration of the delaminated wear sheets produced in particle-reinforced metal-matrix composites.

6. Discussion

The criterion for determining the critical transition load by Equation 9 is related to particle size and intrinsic properties (H_v and I) of the composites. For a given average particle size, the higher the particle volume fraction, f_v, the higher the hardness of the composite. Therefore, the transition load also increases with increasing particle volume fraction as seen in Fig. 1. Also, the transition load increases with d for a given f_v, therefore, particle size influences the transition load more pronouncedly than particle volume fraction. This can also be seen from Fig. 1 for Al_2O_3–Al composites.

The subsurface flow strength, S_f, is associated with the thickness of the delaminated layer. Wang and Rack [9] suggested that friction-induced thermal softening may play a role in wear transitions at high sliding speeds over $1.2\ m\,s^{-1}$. Previous study [14] has, however, shown that high-temperature deformation resistance of aluminium alloys increases with increasing SiC reinforcement. Therefore, the shear strength of such a composite is less likely to be weakened by frictional heating compared with unreinforced alloys and sparsely reinforced composites. The deeper the subsurface crack produced from the worn surface, the higher is the S_f value, and a higher transition load is needed to shear the material. Higher transition loads are expected for a thicker delaminated layer. So the effect of the reinforced particles on delaying the transition load is mainly due to the enhanced shear strength at both room temperature and elevated temperature and different thickness of the delaminated layer caused by particle size.

The adhesive wear model, Equation 12, clearly shows that microstructure parameters such as f_v and d play important roles in determining wear in the composites. The volumetric wear decreases with particle size and increases with particle volume fraction. Our measurements show that the thickness of the wear sheet is one to several particle diameters. The wear sheets of 20 vol % SiC–Al composite are thinner than that of 20 vol % Al_2O_3–Al composite, indicating the effect of particle size on the thickness of delaminated layer. According to Everett and Arsenault's investigation [15], the dislocation density, ρ, in SiC particle-reinforced Al composites can be expressed as

$$\rho = \frac{Bf_v\varepsilon}{b(1-f_v)d} \tag{13}$$

where B is a geometric constant and ε the misfit strain due to the difference of thermal expansion of SiC particles and matrix. It can be seen from Equation 13 that the dislocation density also increases with the particle volume fraction and decreases with particle size. As long as the transition load is reached, the matrix with a high dislocation density does not resist adhesive wear, but accelerates it. In other words, the wear rate increases with increasing dislocation density after the transition to severe wear. The classical adhesion theory and delamination theory have been further improved by this model. Nevertheless, because the wear process is so complex, the wear rate may change with working conditions such as sliding speed and counterface. The proposed model is therefore valid for the present experimental and similar conditions.

7. Conclusion

The critical transition load is determined by subsurface shear and confirmed by the experimental results. Increase in particle size is more effective than increase of volume fraction to raise the transition load for the metal-matrix composites studied here. A new model for adhesive wear has been developed based on dislocation and delamination theory and it relates to the microstructure of the composites. Quantitative predictions of adhesive wear in materials agree well with experimental results.

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References

1. F. P. BOWDEN and D. TABOR, "The Friction and Lubrication of Solids" (Clarendon, Oxford, 1954).
2. N. P. SUH, *Wear* **25** (1973) 111.
3. S. JAHANMIR, N. P. SUH and E. P. ABRAHAMSON, *ibid.* **32** (1975) 33.
4. J. CLARKE and A. D. SARKAR, *ibid.* **82** (1982) 179.
5. J. ZHANG and A. T. ALPAS, *Mater. Sci. Eng.* **A161** (1993) 273
6. A. WANG and H. J. RACK, *ibid.* **A147** (1991) 211.
7. Z. F. ZHANG, L. C. ZHANG and Y-W. MAI, *J. Mater. Sci.* **30** (1995).
8. J. F. ARCHARD, *J. Appl. Phys.* **24** (1953) 981.
9. J. P. HIRTH and J. LOTHE, "Theory of Dislocations", (McGraw-Hill, New York, 1968).
10. P. K. GUPTA and N. H. COOK, *Wear* **20** (1972) 73.
11. A. R. ROSENFIELD, *ibid.* **116** (1987) 319.
12. I. M. HUTCHINGS, "Tribology: Friction and Wear of Engineering Materials", Metallurgy and Materials Science Series (Edward Arnold, London, 1992).
13. E. E. UNDERWOOD, "Quantitative Metallography, Quantitative Stereology" (Addison-Wesley, 1970) pp. 124.
14. T. G. NIEH, *Metall. Trans.* **15A** (1984) 139.
15. R. K. EVERETT and R. J. ARSENAULT, "Metal Matrix Composites: Mechanisms and Properties", (Academic Press, Boston, 1991)

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