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Erosive Wear Behaviour in Duophase SiAlON Composites

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Abstract: This communication describes the erosive wear behaviour of sialon ceramics composed of different ratios of α- and β-phase in terms of material properties such as hardness and fracture toughness by means of a sand blasting method with SiC abrasive particles. Experimental results show that the erosion rate (Δv) of the α/β sialon composites decreases with increasing hardness (H) and fracture toughness (K_c) and can be expressed by $\Delta v = 6.53\ H^{-1.51}\ K_c^{-3.33}$. A correlation factor of 0.98 between the measured and the predicted values has been obtained, indicating that the erosion rate of the composites is predictable with high accuracy if the hardness and fracture toughness of the composite were determined. © 1998 Elsevier Science Limited and Techna S.r.l.

1 INTRODUCTION

Erosive wear behaviour of brittle solids by impacting particles is rather complex because it is controlled by a number of factors such as mechanical properties and microstructure of target materials,[1–3] angle of impingement, and particles properties such as size,[1] shape,[4,5] hardness and fracture toughness.[6] In view of these parameters, two elastic–plastic wear theories have been developed to account for the erosion of brittle solids. Both theories are based on the assumption that lateral cracks grow in a quasi-static manner as a result of residual induced by particle impact. The first is developed by Evans *et al.*[7] who considered a dynamic stress wave effect on the impact event and is expressed by

$$\Delta v \sim V^{3.2}\ R^{3.7} \rho^{1.3}\ K_c^{-1.3}\ H^{-0.25} \qquad (1)$$

where Δv is erosion rate, V is particle velocity, R is particle radius, ρ is particle density, K_c is the fracture toughness and H is the hardness of target materials. The erosion rate described by eqn (1) is inversely proportional to target material's toughness and hardness. The second was developed by Ruff *et al.*[8] who proposed a model based on a quasi-static approach by assuming that the kinetic energy of particles is absorbed completely by plastic flow when the particles impact the surface and they gave

$$\Delta v \sim V^{2.4}\ R^{3.7} \rho^{1.2}\ K_c^{-1.3}\ H^{0.11} \qquad (2)$$

Both eqns (1) and (2) display similar power-law correlatlon which relates the erosion rate (Δv) to particles (V, R, ρ) and target materials (H, K_{1c}). However, some exponent values for the same properties such as velocity, particle density, and hardness differ. Ruff and Wiederhorn have made a comparison between these theories and their experimental results are reasonably consistent with theoretical predictions with regard to the exponent values for velocity and particle size.[8] The difference in exponent for hardness has been critically reviewed by Wiederhorn and Hockey[1] and they have concluded that such difference is primarily due to the use of different parameters, i.e. penetration depth (negative exponent) and maximum load (positive exponent) obtained with an impacting particle event, in estimate of hardness effect.

Based on aforementioned equations, wear resistance has been related to a product of hardness and

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fracture toughness (K_{Ic}^{n} H^{m}). Recently, an extensive study on the erosion behaviour of brittle solids has been conducted by Wada.[6] He used a simple equation to characterize the effect of target material's properties, i.e. primarily hardness and fracture toughness, on erosive wear behaviour under the same conditions of erosion, i.e. V, R, and ρ are fixed. This equation gives

$$\Delta v = CH^{a} K_{c}^{b} \quad (3)$$

where C is a constant and the exponents a and b are constant which usually exhibit negative sign, and all of which can be derived directly from experiments. The correlation between eqn (3) and the experimental results of Wada is reasonably good for a number of brittle materials. Therefore, an attempt was made to use the same formulation [eqn (3)] in the current study to correlate the erosion behaviour with materials properties of the Sialon composites.

Two major erosion mechanisms, i.e. fracture mode and scratching mode, are frequently observed in erosive wear of brittle materials. Both mechanisms can be described in terms of material's properties and in general, for anti-erosion applications, target materials with sufficient mechanical properties such as hardness and fracture toughness are expected to exhibit desired anti-erosion performance. The purpose of this communication is aimed at investigating the erosive behaviour of the state-of-the-art materials, i.e. duophase SiAlON composites. The erosion behaviour of the composites, to the best of our knowledge, has not yet been available in literature and among numerous brittle solids including ceramic composites, α/β duophase sialon composites have demonstrated superior mechanical properties[9,10] and are expected to be a prime candidate material for advanced anti-erosion uses.

2 EXPERIMENTAL PROCEDURES

Powder mixtures containing various proportions of Si_3N_4 (Ube-E10), Y_2O_3 (Cerac. 1047Y), AlN (Tokuyama Soda, H-grade) and $A1_2O_3$ (Sumitomo Chemical, AKP50) were prepared. The powders were ball milled for 20 h with Si_3N_4 grinding media in ethanol solution. The powder slurry was oven-dried at about 100°C, followed by die-pressing under a uniaxial pressure of 50 MPa. The as-formed powder compacts were sintered pressurelessly in 1 atm N_2 gas at 1800°C for 2 h. The densities of the as-sintered specimens were determined by using water displacement technique. Phase content of α and β modifications in the sialon composites was calculated by a method based on determination of intensity ratios of the reflections (101) and (210) of the β phase and (102) and (210) of the α phase. The hardness (having a minimum-to-maximum deviation in the range of $\pm$1.6 GPa) and fracture toughness (having a deviation in the range of $\pm$0.35 MPam$^{1/2}$) of the duophase sialon were measured using Vickers hardness indentation (under a load of 300 g for 15 s) and single-edge notched beam technique (using a 4-point fixture through an Instron tester), respectively.

The erosion tests are carried out by means of a sand blasting method, shown in Fig. 1. The target specimens having dimensions of 20×20×5 mm^3 were placed in position with a fixed distance of 5.0 cm away from the B_4C nozzle having an inside diameter of 5 mm. Such nozzle-to-target separation make an impact area of about 50–60% of the surface area of target specimen. The 120-mesh SiC particle ($\sim$150 μm) are used as impacting abrasives, Fig. 2. Specimen surfaces were subjected to a fixed flux of the SiC abrasive grit under the same conditions of particle velocity (80 m s^{-1}), temperature ($\sim$25°C), and incident angle (90°). The particle velocity was determined by using a time-of-flight developed by Ruff and Ives[11] which will not be described in detail here. Briefly, two discs rotate on a common axis and are parallel to the direction of gas stream. For a fixed speed of rotation, the particle velocity can be determined by measuring the position eroded by particle impact on the second disc, relative to the position of the slit on the first disc. Erosion rate (Δv), obtained by averanging the data of four specimens for each composition, is determined by calculation of the volume loss of the Sialon divided by the volume of SiC abrasive used after a long-term exposure (approximately 10 min). The as-eroded surface was examined by scanning electron microscopy (Cambridge Instruments, S-360).

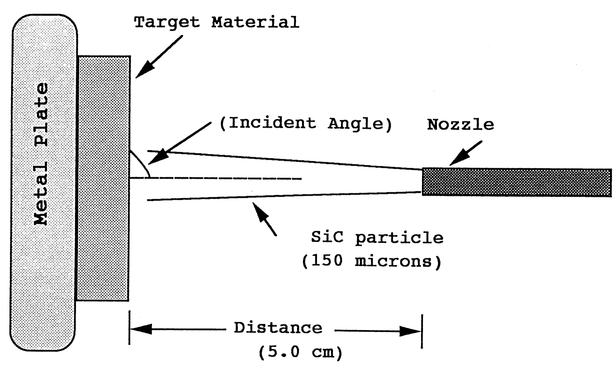

Fig. 1. Schematic diagram of erosive wear test apparatus.

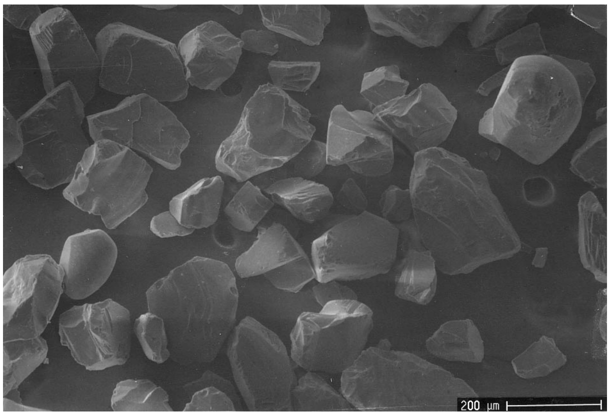

Fig. 2. SEM micrograph of SiC abrasive particles.

3 RESULTS AND DISCUSSION

It is well-known that α-sialon has a higher hardness than β-sialon[12,13] similar to that observed in Si_3N_4 ceramics.[14] An increased α-sialon content in the duophase composites, accompanied by an increase of composite hardness, is therefore a necessary condition for an enhancement of erosion resistance. Figure 3 shows the experimental results that the erosion rate decreases considerably with increase of α-sialon content up to 40 wt%; however, a significant increase of the erosion rate is observed when α-sialon content is as high as 90%. One of the major causes for the increased erosion rate at 90% α-phase content is due to the presence of porosity fraction of ~5% (while others have a porosity fraction over the range of 1–2%), and which is accompanied with a decrease in hardness, i.e. from ~24 GPa for 40% α-phase to 19.6 GPa for 90% α-phase. In fact, the hardness of the sialon composites at 90% is sufficiently higher than that for the composites with α-phase content < 5%. Accordingly, it should exhibit a better erosion resistance compared to that at lower α-sialon content. Therefore, it forces the attention toward the fracture toughness effect. By plotting the toughness values obtained against the erosion rate, Fig. 4 depicts a strong dependence of the erosion rate on the toughness. A value of 5.78 MPa $m^{1/2}$, which is the least tough composition in the present work, is obtained for the composites with 90% α-sialon. This appears to explain the anomalous erosion behaviour in Fig. 3.

The dependence of hardness of the sialon composites on the erosion rate is displayed in Fig. 5 by open circles, and the data of open square are cited from Wada[6] for comparison. The erosion rate of α/β sialon obtained by Wada is higher by approximately an order of magnitude than is obtained in this work, which may be due to different experimental techniques, e.g. particle (erodent) size/shape, particle velocity and flying distance, etc. Figure 5 reveals, in the scale of the plot, that the erosion rate of the duophase sialon composites shows a weaker dependence on the composite hardness, while a stronger hardness dependence is observed for Wada's work.

To correlate the hardness and toughness with erosion rate, a multiple regression analysis is employed to determine the constants C, a, and b in eqn (3) which gives

$$\Delta v = 6.53\, H^{-1.51}\, K_c^{-3.33} \quad (4)$$

for the sialon composites. The hardness exponent in eqn (4) is quite similar to that in Wada's work, i.e. $a = -1.73$. However, both exponent values of the composite hardness and fracture toughness are higher than theoretical predictions in eqns (1) and (2). In a first approximation, the greater dependence of erosion rate on the hardness and fracture toughness of the composites indicates both fracture

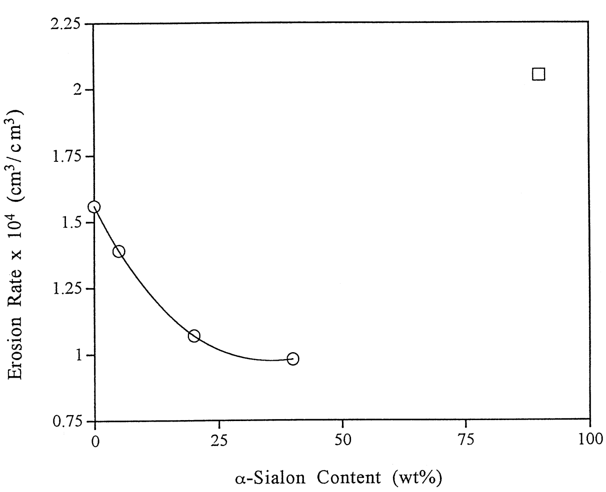

Fig. 3. Erosion rate of the sialon composites in terms of α-sialon content.

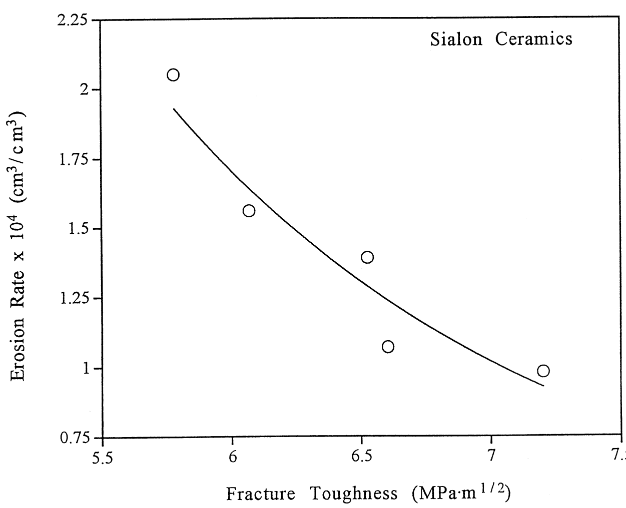

Fig. 4. Dependence of erosion rate on the fracture toughness of the sialon composites.

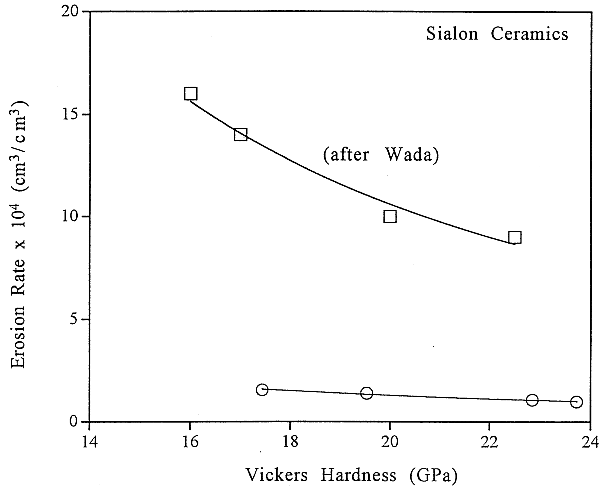

Fig. 5. Dependence of erosion rate on the hardness of the sialon composites with the data cited from Wada for comparison.

and scratching modes are operative simultaneously during the impact. Examination of the eroded surface shown in Fig. 6 (a) and (b) for pure β-sialon and the composite with 90% α-sialon, respectively, reveals a considerable difference in surface morphology. For pure β-sialon [Fig. 6(a)], the material loss is likely to be produced by both fracture and scratching modes, whilst for the composite [Fig. 6(b)] brittle fracture appears to be responsible for material loss. A higher magnification of Fig. 6 as depicted in Fig. 7 reveals that the eroded surface of the composite [Fig. 7(b)] appears to be smoother than that in the pure β-phase sialon [Fig. 7(a)]. This, although not a definite conclusion, suggests that material loss in the duophase sialon could be caused by lateral cracking. This may therefore provide some clues on the explanation of the greater fracture toughness dependence of erosion rate. In fact, it is improper to make a direct comparison between the current result and those deduced from theories, i.e. eqns (1) and (2), primarily because both theories are applicable to monolithic brittle solid. Although an equation was derived recently by Wayne and Buljan to describe the erosion behaviour of two-phase brittle composite materials[15] by taking a microstructure parameter, λ the average spacing of discrete particulate phase into consideration and has received satisfactory results, this parameter cannot be determined in our duophase composites and therefore fails to make any further correlation.

By comparing eqn (4) to all the experimental data, a linear relationship as depicted in Fig. 8 is obtained, which has a correlation factor of 0.82. If considered only for duophase sialon composite, similar to those treated by Wada,[6] the correlation factor is increased to 0.98, indicating that the erosion behavior of the duophase composites can be

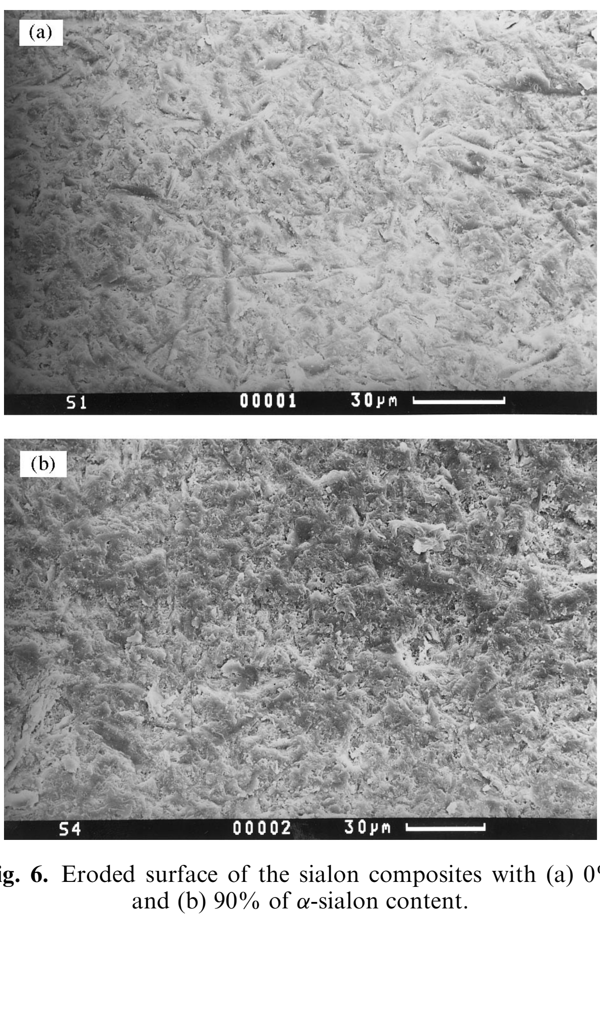

Fig. 6. Eroded surface of the sialon composites with (a) 0% and (b) 90% of α-sialon content.

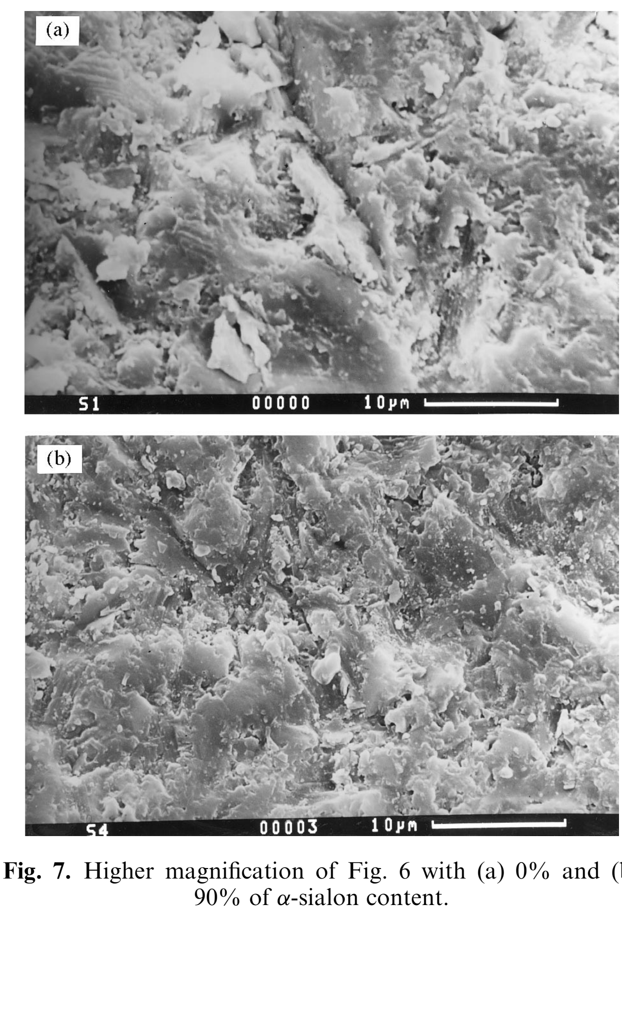

Fig. 7. Higher magnification of Fig. 6 with (a) 0% and (b) 90% of α-sialon content.

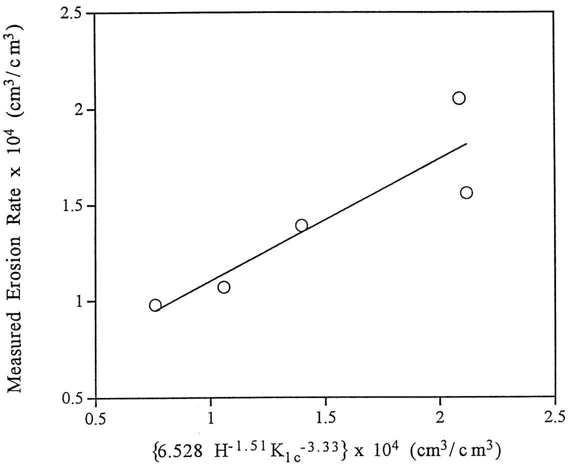

Fig. 8. Correlation between eqn (4) and the experimental data.

predicted with relatively high accuracy once the values of hardness and fracture toughness of the composites are determined by means of eqn (4).

4 SUMMARY

The erosive wear behaviour of duophase sialon composites composed of different ratios of α- and β-sialon is investigated using a sand blasting method with SiC as abrasive particles. An erosion wear model associated with the hardness and fracture toughness of the duophase sialon composites, $\Delta v = 6.53H^{-1.51}K_c^{-3.33}$, has been developed and demonstrated reasonable correlation with the experimental results. This equation indicates that higher values of hardness and fracture toughness are expected to improve the anti-erosion property of the sialon composites.

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REFERENCES

1. WIEDERHORN, S. M. & HOCKEY, B. J., *J. Mater. Sci.*, **18** (1983) 766–780.
2. RITTER, J. E., *Mater. Sci. Eng.*, **71** (1985) 195.
3. RITTER, J. E., STRZEPA, P., JAKUS, K., ROSENFELD, L. & BUCKMAN, K. J., *J. Am. Ceram. Soc.*, **67** (1984) 769–774.
4. KNIGHT, C. G., SWAIN, M. V. & CHANDHRI, M. M., *J. Mater. Sci.*, **12** (1977) 1573.
5. CHANDHRI, M. M. & BROPHY, D. A., *J. Mater. Sci.*, **15** (1980) 345–352.
6. WADA, S., *Key Eng. Mater.*, **71** (1992) 51.
7. EVANS, A. G., GOLDEN, M. E. & ROSENBATT, M. E., *Proc. Roy. Soc. London Ser.*, **A361** (1978) 343.
8. RUFF, A. W. & WIEDERHORN, S. M., In *Treatise on Materials Science and Technology*, Vol. 16, ed. C. M. Preece. Academic Press, New York, 1979, pp. 69–126.
9. EKSTROM, T., KALL, P. O., NYGREN, M. & OLSSON, P. O., *J. Mater. Sci.*, **24** (1989) 1853.
10. KISHI, K., UMEBAYASHI, S. & TANI, E., *J. Mater. Sci.*, **25** (1990) 2780.
11. RUFF, A. W. & IVES, L. K., *Wear*, **35** (1975) 195.
12. EKSTROM, T. & NYGREN, M.,. *J. Am. Ceram. Soc.*, **75**(2) (1992) 259.
13. ISHIZAWA, K., AYUZAWA, N., SHIRANITA, A., TAKAI, M., UCHIDA, N. & MITOMO, M., *Yogyo-Kyokai-Shi*, **94**(1) (1986) 183.
14. GREKOVICH, C. & GAZZA, G. E., *J. Mater. Sci. Letters*, **4** (1985) 195.
15. WAYNE, S. F. & BULJAN, S. T., In *Friction and Wear of Ceramics*, ed. S. Jahanmir. Marcel Dekker Inc., New York, 1994.