STUDIES ON FRACTURE AND STRENGTH BEHAVIOUR OF A1-GLASS AND A1-FLYASH PARTICULATE COMPOSITES

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ABSTRACT

Mechanical properties such as ultimate tensile strength (UTS), 0.2% proof stress, 0.5% compressive proof stress, hardness as well as fracture toughness (K_{IC}) of Al-glass and Al-flyash composites are reported. It is found that UTS and 0.2% proof stress decrease while hardness and 0.5% compressive proof stress increase with increasing volume fraction of dispersoids. However, magnesium seems to increase K_{IC} of the composite as well as strength properties of hot worked composites. The fracture mechanism and observed strength properties of particulate composites are interpreted on the basis of microstructural and fractographic studies. It appears that Rice and Johnson theory of fracture based on microvoid coalescence does not hold good in the case of particulate composites probably due to bimodel distribution of particles which behaved as voids.

KEYWORDS

Particulate composites; Al-glass and Al-flyash composites; Liquid Metallurgy; 0.5% compressive proof stress; stress intensity factor; K_{IC}; post yield fracture; microvoid coalescence; debonding of particles.

INTRODUCTION

Aluminium base particulate composites consist of dispersion of graphite and ceramic particles which are typically of size 10 to 200 μm spaced over the distances of same order of magnitude in aluminium matrix. These metal matrix composites differ from dispersion hardened materials wherein particle size varies from 0.1 to 2 μm and inter-particle distance is few μm. Therefore, it may be expected that the strengthening mechanism in these two types of materials are different. However, particulate composites exhibites lower strengths. Metal-matrix particulate composites particularly those prepared by liquid metallurgy technique are gaining importance for non-aerospace applications since last decade due to their ease of fabrication, near isotropy of mechanical properties (Divecha and co-workers, 1981). In addition, these composites exhibit better machinability and superior tribological properties over the parent matrix materials and various conventional bearing materials (Badia and Rohatgi, 1969; Pai and co-workers, 1974; Pai and Rohatgi, 1974; Rohatgi and co-workers, 1976; Surappa and Rohatgi, 1978, 1981; Krishnan and co-workers, 1980; Rao and co-workers, 1980; Surappa and co-workers, 1982; Surappa, 1978; Krishnan, 1981; Murali, 1982; Banerji, 1982; Patton, 1972; Katsuhiro and co-workers, 1980a, 1980b; Deonath, 1979). Further, whiskers (diameters upto 1 μm and length 1 μm and above) reinforced composites produced by powder metallurgy technique exhibit moderate strength but possess good high temperature properties which is a distinct advantage over the precipitation hardened materials.

Another material property gaining importance is fracture toughness of these materials for fail-safe design in structural applications. Although considerable work has been done in respect of strength, tribological and other properties, very little work has been reported on the fracture toughness of particulate composites. Divecha and co-workers (1981) have reported fracture toughness of Al-SiC composite whose value varied from 13 to 34 MPa $\sqrt{m}$. Das and co-workers (1982) have reported fracture toughness of Al-4 Cu-Li alloy with dispersions of 1% graphite. They have concluded that graphite dispersions through two phase solidification improves the K_{IC} in the high strength alloys. Similarly, Mg enhances the K_{IC} value in Al-Zn-Mg alloys. Renganatha and Srinivasan (1982) have measured the fracture toughness of various cast Al-Cu, Al-Cu-Si and Al-Si alloys with and without dispersions of Al_2O_3 (2%). They have concluded that dispersions of Al_2O_3 in Al-12% Si alloy does not bring about appreciable change in the value of strength intensity factor (K_Q). But in all these studies, neither satisfactory explanation has been given for decreasing strength properties with increasing volume fraction of dispersoids, nor attempts made to understand the fracture mechanism in the particulate composites. ASTM handbook (1975) reports a toughness value of 43 MPa $\sqrt{m}$ for aluminium alloy (201 alloy).

In this paper, which forms part of an overall investigation on the effect of incorporating flyash and glass particles of size 10 to 150 μm in commercial aluminium with and without magnesium addition, fracture and strength properties of Al-glass and Al-flyash composites are reported. In addition, herein an attempt is also made to understand the basic mechanism of fracture and causes for the observed low strength in particulate composites.

EXPERIMENTAL

Table 1 lists the nominal compositions of composites used in the present investigation. These are prepared by liquid metallurgy technique wherein particles of glass/flyash were introduced into the molten aluminium by forming a vortex using a stirrer and then subsequently casting the same. Specimens of size 25 mm diameter and 15 mm height, 50 mm gauge length and 20 mm diameter, 30 mm height with l/d ratio equal to 1.5 were used for hardness, tensile testing and compressive testing respectively under the testing conditions mentioned in Table 1. In all average of three indentations on a specimen constitute one hardness reading while 5 specimens were tested in each case for tensile and compressive testing using Brinell hardness tester and Instron testing machine respectively.

TABLE 1 Mechanical Properties of Composites

Sl. No.	Composition	Casting condition/ process	0.2% proof stress MPa	U.T.S. MPa	% elongation	0.5% compressive proof stress MPa	Hardness Brinell	K_{IC}
1.	Al	Chilled cast	40.3	75.5	37	95	27	-
2.	Al-2.5 flyash	,,	43.9 $\sigma \pm 2.6$	79.6 $\sigma \pm 2.9$	12.3	119	33	-
3.	Al-1 Mg	,,	46.6	81.3	22	-	39	-
4.	Al-1 Mg-5 Flyash	,,	43.2 $\sigma \pm 3.1$	72.8 $\sigma \pm 1.7$	6.9	135	48	-
5.	Al-2 Mg	,,	57.3	95.12	16.6	-	49	-
6.	Al-2 Mg-10 Flyash	,,	47.3 $\sigma \pm 4.8$	72 $\sigma \pm 3.9$	4.1	167	71	-
7.	Al-3 glass	,,	53.6 $\sigma \pm 3.1$	83.6 $\sigma \pm 4.6$	14.6	124	31	-
8.	Al-1 Mg-5 glass	,,	47.7 $\sigma \pm 4.9$	75.1 $\sigma \pm 3.8$	8.1	135	47	-
9.	Al-2 Mg-10 glass	,,	41.6 $\sigma \pm 3.8$	69.3 $\sigma \pm 4.1$	5.1	143	65	-
10.	Al-2 Mg-8 glass	Hot pressed	-	196.3	7.6	-	-	16.63
11.	Al-2 Mg-10 flyash	,,	-	189.8	5.3	-	-	14.76
12.	Al-2 Mg-5 glass	Hot forged	-	208.6	6.7	-	-	16.60
13.	Al-2.5 flyash	,,	-	161.3	8.3	-	-	14.35
14.	Al-2 Mg-7.5 flyash	,,	-	208.0	5.8	-	-	17.61

Note: Throughout the investigation commercially pure aluminium is used.
In both tension and compression test cross-head speed was 0.005 m/min.

Hot extrusion of composites was carried out on billets of size 15 cm diameter and 25 cm in length heated to 480°C in a horizontal extrusion press of capacity 1250 tons with extrusion ratio of 15:1 and speed of run of 02 M/min.

Fracture toughness testing of Al-glass and Al-flyash composites was carried out on specimens of size 25 mm thick, obtained from 40 mm thick billets which was subjected to a compressive load of 1000 tonnes during extrusion. The specimens having other dimensions specified in ASTM E-399 were then used for precracking by fatigue testing at 800 Kg (max) and 80 Kg (min) load with a frequency of 40 C/sec. After the crack was extended to about 1.3 mm (this was measured using a microscope), the specimens were subjected to tensile testing for final fracture. Both precracking and final fractures were carried out using a MTS machine.

Metallograph and fractographic studies were carried out on composite samples after suitable preparation using a Leitz metalloplan and JEOL Scanning Electron Microscope respectively.

RESULTS AND DISCUSSIONS

Strength Properties of Composites

Table 1 lists the values of hardness, UTS, % elongation, 0.2% proof stress, 0.5% compressive proof stress and K_{IC} of Al-glass and Al-flyash composites studied in the present investigations. Fig. 1 shows the mechanical behaviour of these composites with and without hot extrusion as a function of volume fraction of dispersoids. It can be seen from Table 1 that in the case of both Al-glass and Al-flyash composites, tensile strength decreases while 0.5% compressive proof stress and hardness increase with increasing volume fraction of dispersoids. Also, strength properties of composites are improved considerably (~280% for 8% glass composite) by extrusion when compared to that of matrix (~180% increase only). Similarly, results were observed on hot rolling and forging of these two composites as well as for hot worked Al-SiC whiskers (Divecha and co-workers, 1981) and Al-glass composites prepared by powder metallurgy techniques (Bergmann and co-workers, 1979). Data on strength properties of Al-glass and Al-flyash composites can be compared with similar data for other aluminium based particulate composites such as Al-graphite and Al-shell char prepared by liquid metallurgy technique by other investigators. Figure 2 is a normalized plot of $\sigma_{composite}/\sigma_{matrix}$ versus volume fraction of dispersoids. Here the strengths of composites are normalised with respect to base matrix so as to plot the data of various investigators including the present study on the same scale irrespective of type of dispersoids. The solid line is a theoretical estimate of strength based on post yield fracture mechanics as discussed below assuming particles (dispersoids) behave as voids.

As a first order approximation, one can treat these particles as critical crack and determine the tensile strength from fracture equation given by

$$K_{IC} = \sigma \sqrt{\pi a} \qquad (1)$$

where K_{IC} is the fracture toughness of the materials and a is the radius of the particle. Equation (1) assumes linear elastic behaviour in the material. The fracture toughness of the composite is not known, but for these cast materials it could be quite low. Assuming a minimum K_{IC} of 15 MPa $\sqrt{m}$ and taking the radius of particles as 50 um, one arrives at a tensile strength of approximately 1200 MPa. This is much larger than the average value of strength which is around 200 MPa for these composites which clearly indicates

that plasticity aspects have to be invoked.

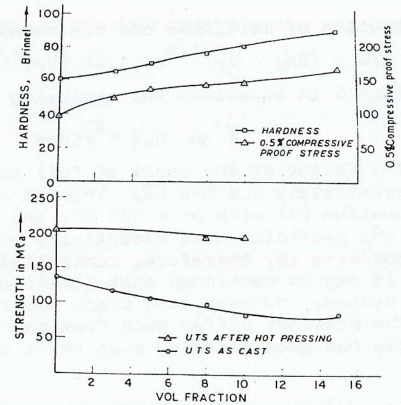

Fig. 1. Mechanical behaviour of composites with respect to volume fraction of dispersoids.

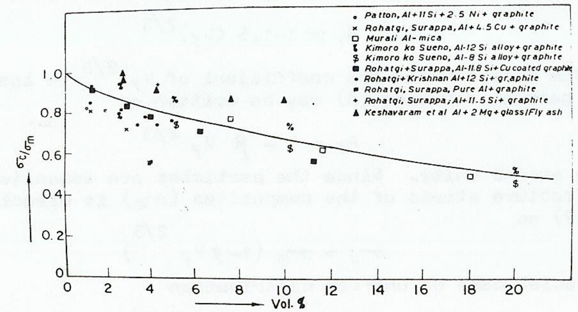

Fig. 2. Normalized plot of σ_c/σ_m vs volume fraction of dispersoids.

Assuming the Dugdale (1960) and Bilby and co-workers (1963, 1964), model for small scale plasticity between the system of particles (acting as voids) and the crack opening displacement (δ) at the crack tips at the point when plasticity spreads throughout the system is approximately given by

$$\delta = 8/\pi \cdot \sigma_y/E \cdot a \cdot \ln\left[\operatorname{Sec}\left\{\pi/2\,(1-a/w)\right\}\right]\alpha \qquad (2)$$

where σ_y and E are respectively flow stress and modulus of the matrix, a is the radius of the particles (crack), 2w is the average centre to centre distance between particles (cracks) and α is a factor which takes into account the presence of other cracks whose value is approximately 2.6 for $a/w = 0.6$ (Bilby and co-workers, 1963, 1964).

For uniform distribution of particles one can obtain

$$a/w = (6/\pi \cdot V_f)^{1/3} \simeq 1.24\ (V_f)^{1/3} \qquad (3)$$

substituting equation 3 in equation 2 and employing the equation

$$\delta \simeq 0.5\ K^2/E\sigma_y \qquad (4)$$

the stress intensity factor at the onset of full scale yielding for $V_f = 0.1$ is obtained as approximately 2.6 MPa $\sqrt{m}$. This is almost of the same value evaluated using equation (1) with $\sigma = 200$ MPa and $a = 50\ \mu m$. Hence, until general yielding, the particles have essentially no effect on fracture. The strength of the composite is, therefore, controlled by post-yield fracture characteristics. It may be mentioned that equation (2) is based on one dimensional crack system. However, the crack opening displacement (COD) for penney shaped cracks does not differ much from one dimensional system. Hence the discussion above has good validity even for a two dimensional crack system.

Beyond the point of general yield, the COD rises linearly with strain. The fracture stress is, therefore, controlled by limit load on the ligament areas. If a is the radius of the particle, 2w is average centre to centre distance between the particles, then the fractional ligament area (A_f) is given by

$$A_f = 1-(a/w)^2 \qquad (5)$$

For a uniform particle distribution, from equation (3) we get

$$A_f \simeq 1-1.5\ (V_f)^{2/3} \qquad (6)$$

For a random distribution the coefficient of $V_f^{\ 2/3}$ is less than unity. Hence, in general equation (6) may be written as

$$A_f = 1 - \beta\ V_f^{\ 2/3} \qquad (7)$$

where β is around unity. Since the particles are essentially non-load bearing, the fracture stress of the composites (σ_c) is directly obtained from equation (7) as

$$\sigma_c = \sigma_m\ (1-\beta V_f^{\ 2/3}) \qquad (8)$$

For the special case of uniform distribution

$$\sigma_c/\sigma_m \quad (1-1.5\ V_f^{\ 2/3}) \qquad (9)$$

This is plotted as a solid line in Fig. 2.

Experimental data of Al-glass and Al-flyash composites fall above this line. Hence, it may be concluded that bonding between particles and matrix in these composites is much better than that in other Al-base composite systems studied so far. This is evident from the Scanning electron microscopic studies. Figure 3(a) shows typical fractured surface of Al-flyash composite revealing fractured to be essentially by micro-void coalescence. Figure 3(b) and 3(c) are fractographs of Al-flyash and Al-glass composites respectively taken at higher magnifications indicating that some particles do crack while most of them actually debond from the matrix. This further suggests that the bonding between particles and matrix is weak. Thus at low strains the particles seem to behave as voids, excessive damage to these particles and low bonding between the particles and matrix will lead to low strength properties of the particulate composites.

The increase in strength after extrusion may be attributed to breaking up of dendritic structure, work hardening of the material and closing up of voids in composites. It is evident from Fig. 4 which is a typical photomicrograph of Al-glass composite/Al-flyash composites in as-cast (Fig. 4a & b) and extruded condition (Fig. 4c & d) that fibrization (in the case of Al-glass composite)/formation of stringers (in the case of Al-flyash composites) are formed in the direction of extrusion. These result in reduction of particle size, improved bonding between particles and the matrix and probably dislocation structure around particles, contributing to the increase in strength of Al-glass and Al-flyash composites. Presence of texture effect around particles in the matrix is not analysed though it is also a contributing factor for the increase in strength of composites after hot extrusion. Similar results have been obtained by Bergman and co-workers (1979) and Divecha and co-workers (1981) who have forecast a good future for particulate composites on mechanical working for use in structural applications.

The increase in compressive strength of Al-glass and Al-flyash composites with volume fraction agrees with that of hardness data for these composites. It is believed that the voids associated with particles in the composite may be due to air entrapment. Therefore, these voids open up in tension and accelerate yielding by stress concentration whereas under compression such stress concentration effects are negligible. The particles do impe e large scale yielding. Unfortunately, it is not realised in tension.

Fracture Toughness of Composites

Figure 5 shows typical load deflection curve obtained in fracture toughness testing of Al-glass and Al-flyash composites. It can be seen from the figure that there is a stable crack growth beyond the point of maximum load. This feature was also observed during experimentation where the crack was observed to propagate slowly during overload fracture. In fact, the crack propagation could have been stopped by stopping the actuator displacement of the MTS. The load (P_Q) located on the curve by drawing a line whose slope is 2% higher than the linear part of the curve is in accordance with ASTM E-399. Then, the stress intensity factor (K_Q) was calculated using the formu

$$K_Q = P_Q/BW^{\frac{1}{2}} \quad 29.6\ (a/w)^{\frac{1}{2}}-185.5\ (a/w)^{3/2}+655.7\ (a/w)^{5/2}- 1017\ (a/w)^{7/2}+638.9\ (a/w)^{9/2} \qquad (10)$$

The value of K_Q thus obtained was taken as equal to K_{IC} since all the conditions for plain strain as given in ASTM E-399 are satisfied. The toughness which is typically around 15 to 17 MPa $\sqrt{m}$ which is comparable to the toughness of cast aluminium alloys is listed in Table 1. It is also observed that Mg seems to have increased the fracture toughness of Al-glass and Al-flyash composites. Similar results were observed in the case of other Al-alloys reported elsewhere (Das and co-workers, 1982).

The K_{IC} value for composites could not be compared with that of the matrix, since the required size of the matrix specimen to satisfy plain strain condition was too large. The only comparison with cast composite can be Al 201 which is reported as 43 MPa $\sqrt{m}$ (ASTM, 1975). Thus, the particles do decrease K_{IC} of the material. This is also obvious from SEM (Fig.3) where particles act as potent nucleating sites for voids.

It is also observed that Mg increases K_{IC} and this is in agreement with the fact that Mg increases wettability between the particles and matrix and helps in the dispersion of particles as observed in other systems (Surappa, 1978, 1980; Pai and co-workers, 1976; Krishnan, 1981; Murali, 1982; Banerji, 1982; Renganathan and Srinivasan, 1982). Therefore, Mg does seem to inhibit void nucleation, i.e. they act as better bonding agents.

From the fracture toughness data one can estimate the distance between voids using famous Rice and Johnson's theory of fracture based on micro-void coalescence. From K_{IC} one can obtain crack opening displacement δ given by

$$\delta = K^2/2E\sigma_y \qquad (11)$$

The COD should be the order of spacing between voids. It is evident from the scanning electron micrographs (Fig. 3) that inter-void spacing is approximately 20 to 30 μm. Therefore, it is evident that Rice and Johnson's analysis does not hold good for these composites though they fit in very well with dispersion hardened aluminium alloys. It is important to note that not all voids are associated with particles as it is evident from the Fig. 3b & c. Hence, it seems that larger voids are nucleated at particles because of poor bonding and then the remaining ligaments, by formation of void sheet connecting the particles. It is probably because of these bimodel distribution of voids that the Rice and Johnson's analysis does not seem to hold good. However, further studies are in progress to confirm the above analysis.

ACKNOWLEDGEMENT

The authors wish to thank Mr. Brahma, Ph.D. Scholar for helping in fracture toughness measurements, Mr. S.G.K. Pillai and Mr. Peter Koshy for microscopic work, Mr. K. Sukumaran for the help in measurement of strength using Instron testing machine and Mr. R.N. Pillai for excellent typing work.

REFERENCES

ASTM Handbook (1975), Taylor Leyman (Ed.), American Society for Metals, Metals Park, Ohio, Vol.1, 8th ed., p. 823.
Badia, F.A., and P.K. Rohatgi (1969). Trans. Soc. Automotive Engg., 76, 1200.
Banerji, A. (1982). M.Sc. Thesis, Kerala University, Trivandrum.
Bergmann, H.W., G. Frammoyar, and G. Wassermann (1979). Z. Metallk., 70(19), 802.
Bilby, B.A., A.H. Cottrell, and K.H. Swinden (1963). Proc. Roy. Soc. (London), A272, 304.
Bilby, B.A., A.H. Cottrell, E. Smith, and K.H. Swinden (1964). Proc. Roy. Soc. (London), A279, 1.
Das, N.K., R.K. Mahanti, C.S. Sivaramakrishnan, and R. Rajendra Kumar (1982). Ind. J. Techn. 20(11), 434.
Deonath (1979). Ph.D. Thesis, Indian Institute of Science, Bangalore.
Divecha, A.P., S.G. Fishman, and S.B. Karmarkar (1981). J. Metals, 33(9), 12.
Dugdale, D.S. (1960). J. Mech. Phys. Solids, 8, 100.
Katsuhiro, Y.K., M. Suwa, and S.D. Masato (1980). Japan kokai Tokkyo, Koho, 80, 24, 950.
Katsuhiro, Y.K., M. Suwa, S.D. Masato, and K. Takashi (1980). Japan kokai Tokkyo, Koho, 839.
Krishnan, B.P., N. Raman, K. Narayanaswamy, and P.K. Rohatgi (1980). Wear, 60, 205.
Krishnan, B.P. (1981). Ph.D. Thesis, Indian Institute of Science, Bangalore.
Murali, T.P. (1982). M.Sc. Thesis, Kerala University, Trivandrum.

Pai, B.C., P.K. Rohatgi, and S. Venkatesh (1974). Wear, 30(1), 117.
Pai, B.C., and P.K. Rohatgi (1974). Trans. Ind. Inst. Metals, 27(2), 97.
Pai, B.C., Subrat Roy, K.V. Prabhakar, and P.K. Rohatgi (1976). Mater. Sci. & Engg., 24(1), 31.
Patton, A.N. (1972). J. Inst. Metals, 100, 197.
Renganathan, S., and M.N. Srinivasan (1982). Bull Mater. Sci. 4(1), 29.
Rao, N.A.P., S. Biswas, P.K. Rohatgi, A. Santhanam and K. Narayanaswamy (1980). Tribology Int., 13(4), 171.
Rohatgi, P.K., N. Murali., H.R. Shetty, and R. Chandrasekhar (1976). Mater. Sci. & Engg., 20, 115.
Surappa, M.K., and P.K. Rohatgi (1978). Trans. Ind. Inst. Metals, 31(6), 488.
Surappa, M.K. (1978/1980). M.Sc./Ph.D. Thesis, Indian Institute of Science, Bangalore.
Surappa, M.K., and P.K. Rohatgi (1981). J. Mater. Sci., 16(6), 1599.
Surappa, M.K., S.V. Prasad, and P.K. Rohatgi (1982). Wear, 77(3), 295.

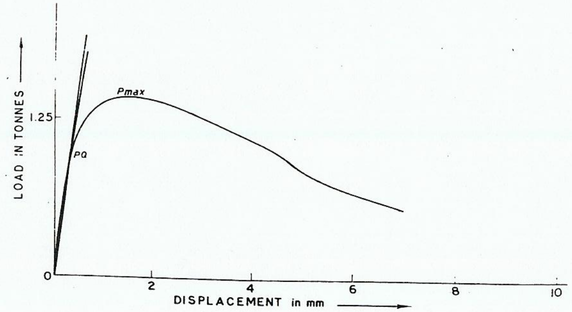

Fig. 5. Load-deflection curve of composites.