

Fracture of Steel/Carbide Particulate Composites

ANN M. RITTER, MELVIN R. JACKSON and ROGER N. WRIGHT*

*GE Corporate Research and Development, P. O. Box 8,
Schenectady, New York 12301, USA*

**Department of Materials Engineering, Rensselaer Polytechnic
Institute, Troy, New York 12181, USA*

ABSTRACT

The strength, ductility and fracture behavior of stainless steel/carbide particulate composites were studied using tensile samples and bend bars with 0-40 vol/o carbide. Strength increased with increasing amounts of carbide, and the ductility decreased. The ductility was affected by the size distribution and type of carbide, and the specimen heat-treatment. Composite fracture was initiated by cracking of the carbides, followed by ductile tearing of the matrix. Composite fracture required a critical amount and spacing of cracked carbides, and apparently a critical level of strain localization in the matrix.

KEYWORDS

metal matrix composites; fracture criteria; particulate composites

INTRODUCTION

The addition of brittle particles to ductile metal matrices to form composites enhances the strength and reduces the ductility (Edelson and Baldwin, 1962). Such composites can be made by powder metallurgy techniques, including low pressure plasma deposition (Muehlberger, 1973; Siemers et al., 1985), which incorporates particulate in the matrix by melting both metal and brittle powders in a plasma gun and depositing the molten material on a substrate. This process, commonly used to produce monolithic or composite coatings (Rairden, 1980; Price et al., 1977), has also been used to fabricate structural deposits, from which bulk material properties can be obtained (Siemers et al, 1985; Jackson et al., 1981, 1988). It has been shown (Ritter, et al., 1988) that the variation of such properties as strength and ductility with the amount of particulate in the plasma-sprayed composites is similar to that of composites produced by other powder metallurgy techniques (Edelson and Baldwin, 1962). However, the effects of particle type, size and morphology on composite ductility are not clear in comparing data from some studies (Edelson and Baldwin, 1962; Gladman et al., 1971; Brindley and Lindley, 1972). In the present work, these effects will be detailed, and considered in relation to the composite fracture process.

EXPERIMENTAL PROCEDURE

The materials used were produced in a low-pressure plasma-spraying system using mixtures of 316 stainless steel and carbide powders. The carbide powders were TiC and NbC, with the former from three lots. One had a maximum particle size of about 70 μ m, and an average particle size of 3 μ m, and is designated -325 TiC. The second lot had a maximum size of about 40 μ m, and an average size of 3 μ m, and is designated as 3 μ m TiC. These lots had a broad particle size distribution. The third had a maximum particle size of 5 μ m and an average size 1.0 μ m, and is designated as 1 μ m TiC. The NbC powder lot had a maximum particle size of about 30 μ m, with very few particles of this size, and an average size of 2 μ m. The 1 μ m TiC and NbC powder lots had a narrow particle size distribution. Deposit thicknesses were generally 0.3-0.5". After plasma-spraying, the deposits were consolidated by hot-isostatic-pressing (HIP) at 1100 °C or heat treatment at 1200°C/4hr. The heat-treated samples were water-quenched, and the HIP'ed ones were cooled slowly in the chamber.

Microstructures were evaluated using light microscopy and transmission electron microscopy (TEM). Tensile samples and bend bars from the deposits were tested at room temperature, and the amount of carbide in each sample was measured (12). The fracture surfaces of the tensile samples were examined in the scanning electron microscope, and polished longitudinal sections were characterized by optical microscopy. Bend bars were polished metallographically and strained sequentially in three-point bending, with replicas made at each strain. The replicas were examined using optical microscopy and TEM.

The volume-percent of cracked carbides at each strain level on the bend bars and on polished longitudinal sections of the fractured tensile specimens was measured, as was the average size, defined as the maximum dimension, of the cracked carbides. From the average size of the cracked carbides, d_{cr} , the average spacing between the cracked carbides, D_{cr} , was calculated using(1):

$$D_{cr} = d_{cr} (1-f_{cr})^{1/2} / (2/3f_{cr})^{1/2}$$

where f_{cr} is the volume-percent of cracked carbides. The average length of the cracks in the carbides was also measured on metallographic sections.

RESULTS

Microstructures of typical samples are shown in Figure 1, with angular unmelted carbides in the 316 stainless steel/TiC material, and more rounded particles in the NbC-containing composites. Some melting of the TiC particles was occasionally observed, resulting in a lamellar structure typical of plasma-sprayed materials (Siemers et al., 1985; Jackson et al., 1988).

The addition of TiC and NbC to the stainless steel resulted in significant strengthening (Fig.2). This effect is due to the decreasing mean free path between particles as the volume fraction of carbide was increased, as discussed in an earlier paper (Ritter et al., 1988). The different amounts of strengthening arising from the TiC and NbC particles appear to be related to localized differences in the mean free path (Ritter et al., 1988), attributed to variation in particle morphology (Kelly, 1972). Increasing amounts of carbide had an adverse effect on the ductility (Fig.3).

Cracking of the carbides was seen on the fracture surfaces of tensile samples (Fig.4a). Fracture of the matrix regions between the cracked carbides occurred by ductile tearing. There was no lateral growth of voids (Cox and Low, 1974; VanStone and Cox, 1976) associated with the cracked carbides, and no debonding of the carbide/matrix interfaces (Fisher and Gurland, 1967). In bend bars, cracking was first observed in the largest particles at low applied strains. As the applied strain increased, cracking was then observed in the small carbides, and so the average size of all cracked carbides decreased. Multiple cracking was seen in the large carbides (Fig 4b), and the cracks tended to be perpendicular to the applied stress. The amount of cracked carbide in samples with a broad TiC size distribution increased systematically with strain, as seen in Figure 5. In samples with a narrow particle size distribution, most of the carbide cracking occurred within a small strain range (Fig. 5). Straining these materials to fracture did not result in more carbide cracking.

In all samples, as the amount of cracked carbide increased, and the average size, defined as the maximum dimension, of the cracked carbide decreased, the spacing between the cracked carbides decreased. In tensile samples, composite fracture occurred at critical levels of cracked carbide, f_{cr} , as shown in Table 1.

Table 1. Cracked Carbides at Composite Fracture

Total Vol/o Carbide	Carbide Type	Ave. Size of Cracked Carbides	Vol/o Cracked Carbide
7 - 28 vol/o	-325 TiC	2.9 - 9.9 μ m	5.3 \pm 1.6 vol/o
6 - 27	3 μ m TiC	3.8 - 5.7	4.9 \pm 0.8
6 - 11	1 μ m TiC	1.4 - 1.8	1.7 \pm 0.3
9 - 28	2 μ m NbC	3.2 - 5.2	6.7 \pm 1.9

The average spacing between the cracked carbides at composite fracture, D_{cr} , was calculated from these data. This is shown as a function of the average length of the cracks in the carbides in Figure 6.

DISCUSSION

Fracture in the steel/carbide particulate composites was initiated at the cracked carbides, as has been seen in other particulate systems (Cox and Low, 1974; Gangulee and Gurland, 1967), and occurred by matrix tearing between the cracked carbides. A criterion for fracture is the presence of a critical amount of cracked carbide, which corresponds to a critical spacing between the cracked carbides. A critical level of strain localization in the matrix is apparently also needed for composite fracture to occur, as inferred from the data from the NbC and 1 μ m TiC samples. In these, a large amount of carbide cracking was seen in a narrow strain range, with essentially no more carbide cracking as the strain increased above that range, as seen in Figure 5. However, additional applied strain was necessary for composite fracture, indicating that strain localization was critical to the fracture process, as has been shown for ductile fracture in other systems (Gurland and Plateau, 1963; Curry and Pratt, 1979).

The sequence of events leading to composite fracture in the steel/carbide composites is as follows: (1) the largest carbides crack at low strains, and more

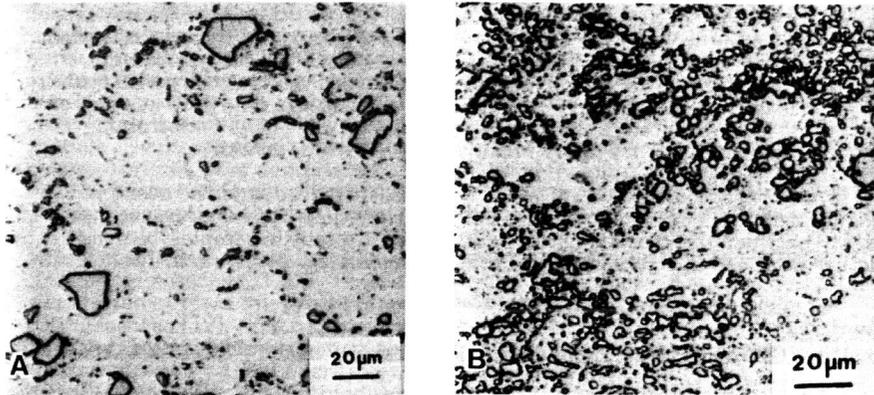


Figure 1. (a) 316ss/10vol/o TiC (-325) deposit, with angular unmelted carbide. (b) 316ss/10vol/o NbC deposit with rounded unmelted carbide.

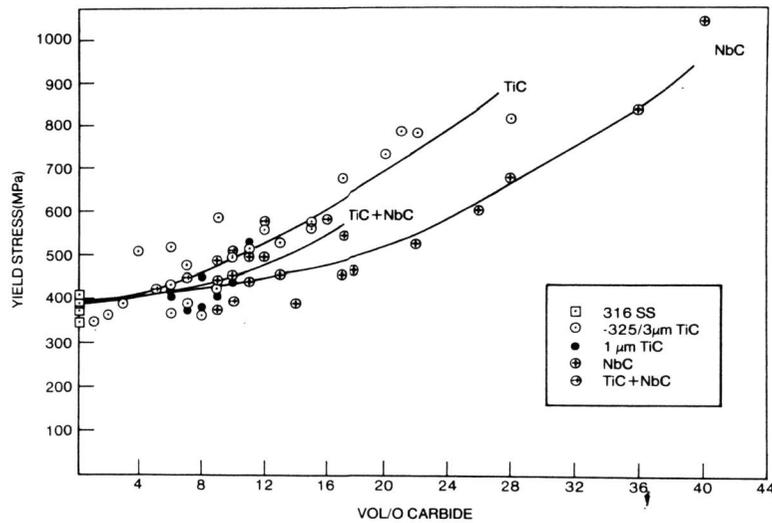


Figure 2. Room temperature yield stress versus volume-percent carbide.

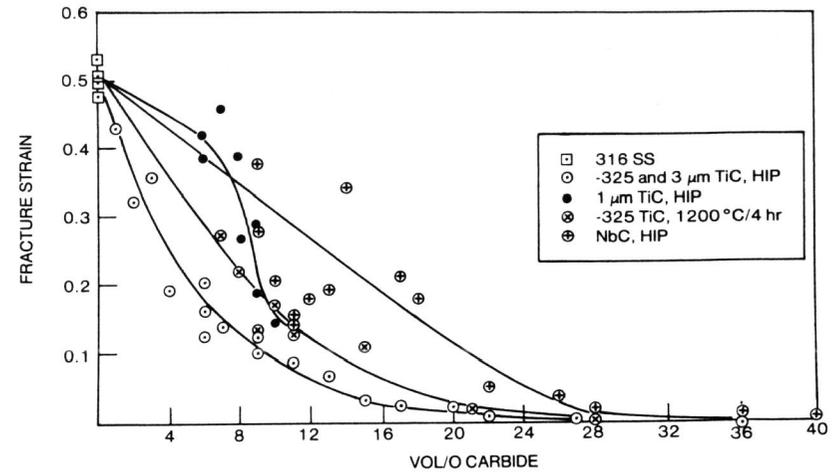


Figure 3. Room temperature true fracture strain versus volume-percent carbide.

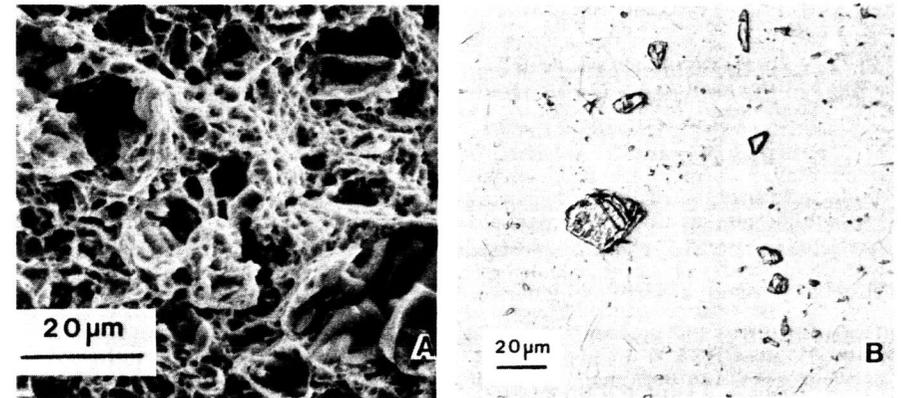


Figure 4. Cracked TiC particles on (a) fracture surface of room temperature tensile sample, and (b) surface of bend bar, with multiple cracks in large carbide

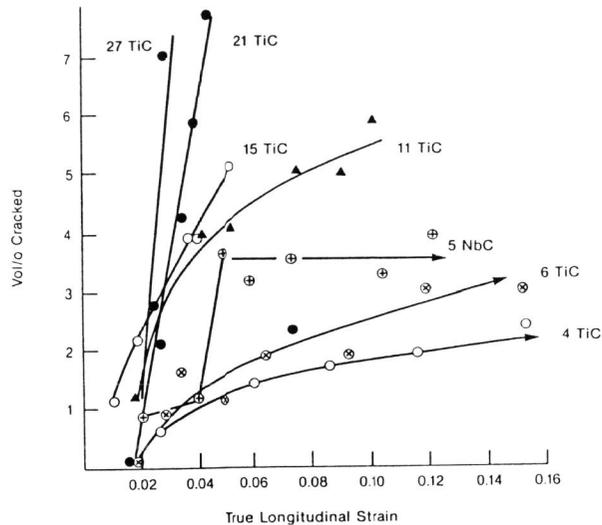


Figure 5. Amount of cracked carbide as a function of strain in bend bars. Samples containing 4-6 TiC and 5 NbC strained to higher levels than shown, samples containing 11-27 TiC fractured at strains on graph.

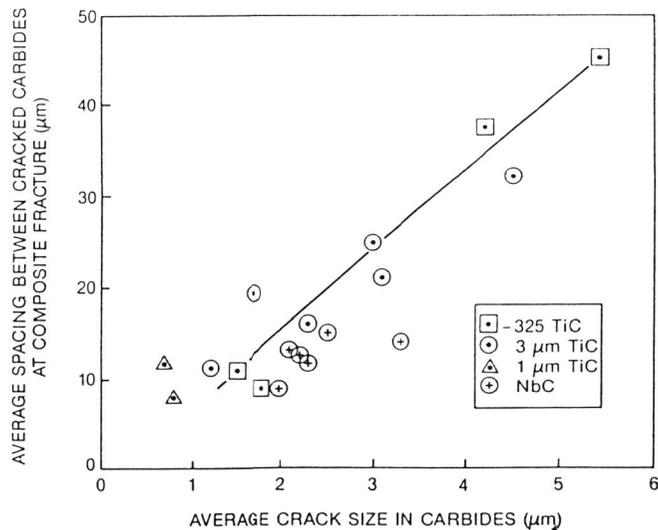


Figure 6. Average spacing between cracked carbides at composite fracture as a function of the average length of cracks in the carbides. Line drawn through data for TiC-containing samples.

carbide cracking occurs as the applied strain increases; (2) the average spacing between the cracked carbides decreases; (3) strain localization in the matrix occurs at the tips of the cracks in the carbides; (4) a critical spacing between the cracked carbides and a critical level of strain localization are reached; and (5) matrix regions between the cracked carbides fail by ductile tearing, and the composite fractures.

The details of the fracture process can be used to explain the effects of variables, such as carbide content, size distribution and type on the composite fracture strain. The first point to be considered is the relation between the amount of carbide and composite ductility. It is well known that increasing particle content causes decreased ductility, due to strain localization in the matrix associated with the particles (Edelson and Baldwin, 1962). In the present study, this effect was quantified. At a specific level of applied strain, the amount of cracked carbide increased as the total carbide content of the sample increased, as shown in Figure 5. This is a result of the increased flow stress in the higher-carbide samples (Fig. 2). Because of the higher carbide cracking rate, the critical amount of cracked carbide needed to initiate composite fracture will be reached at a lower applied strain.

The effect of variables such as particle size and type on the fracture process will now be considered. In samples containing -325 and 3 μ m TiC, there were very large particles, which cracked at low applied strains. The magnitude of the stress and strain localization at a crack tip increases with crack size. Therefore, the critical level of strain localization needed for composite fracture would be reached at lower applied strains than for samples containing mostly fine particles. This can be seen in Figure 3, which shows the 1 μ m TiC and NbC samples to be more ductile than the -325 and 3 μ m TiC samples. In addition, fracture by ductile tearing of matrix regions between very widely-spaced cracked particles may occur if the particles are large, due to the high strain localization associated with the cracks. If the cracked carbides are small, the critical distance between these particles necessary for fracture to propagate is small. This is seen in Figure 6, which shows that the critical spacing between cracked carbides at composite fracture varies linearly with the average size of the cracks in the carbides.

Differences in strain localization around particles may also arise from differences in particle morphology. Morphology affects the ease of matrix slip near the particle by changing the mean free path between particles (Kelly, 1972). The angular shape of the TiC may result in higher strains near these carbides, relative to the more rounded NbC particles. Higher local strains associated with the angular shape would enhance the local strains developed at cracked particles. The net effect would be a reduction in the amount of applied strain needed for tearing of the matrix. This may be a factor in the generally lower ductilities seen in samples containing TiC.

Since a critical amount of carbide must be cracked for fracture to occur in these materials, delay of carbide cracking to high applied strains results in increased ductility. This is seen in Figure 3, where, for samples containing from 7-15 vol% TiC, heat treating at 1200°C followed by a water quench resulted in higher ductilities relative to HIPed samples. This is presumably because higher residual compressive stresses due to thermal expansion mismatch between particles and matrix were present in carbides in the rapidly-quenched samples. This allows the carbides in the quenched samples to resist cracking to higher applied strains, an effect verified by experiments on bend bars (Ritter et al., 1988).

CONCLUSIONS

- 1) Fracture in the steel/carbide composites occurred when a critical amount and spacing of cracked carbide were reached. A critical level of strain localization was apparently also necessary for fracture.
- 2) Composite ductility decreased as the carbide content increased, due to rising flow stresses which caused the fracture criteria to be reached at lower strains.
- 3) In samples with a broad particle size distribution, the presence of large particles, which cracked at low strains, resulted in reduced composite ductility.
- 4) Carbide shape affected composite ductility, with the higher strain localization associated with angular morphologies resulting in decreased ductilities.
- 5) Carbide cracking in quenched samples was delayed to higher strains, apparently by changing the residual stress state in the particles. This enhanced composite ductility, since the critical amount of cracked carbide needed for composite fracture was reached at higher applied strains.

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