JOURNAL OF MATERIALS SCIENCE LETTERS 12 (1993) 366–368

Strength and microstructure of electrodischarge-machined titanium diboride

W. J. TOMLINSON, K. N. JUPE
School of Natural and Environmental Sciences, Coventry University, Priory Street, Coventry CV1 5FB, UK

Ceramics are characteristically hard and brittle, and conventional methods of cutting are limited in terms of the efficiency of the process, geometry of the workpiece and damage to the machined surface. In contrast, electrodischarge machining (EDM) is a thermal erosion process in which material is removed by an electrical discharge between the workpiece and the tool [1]. Temperatures during sparking have been estimated to be about 8000 °C, and the hardness and brittleness of ceramics does not affect the efficiency of the process or the geometric complexity of the workpiece. However, a major limitation is that the material must have a sufficiently high electrical conductivity, and this is the case in only a few ceramics. There are two types of EDM (wire cutting and die sinking) and work on ceramics have used mainly the wire cutting technique. Konig *et al.* [1] recently critically reviewed the fundamental material properties and machining processes, and the future steps necessary for EDM of ceramics.

Studies on the EDM of ceramics have been confined mainly to silicon carbide [2–8], titanium diboride [9–11] and various composites [12–14], and the effects of the machining parameters on the process, material removal rates, surface roughness and surface microstructure. The thermal shock sensitivity is an important parameter in determining the mechanism of material removal [1], and hence of the microstructure of the machined surface. When the ceramic is sensitive to thermal shocking, material is removed by spalling [9], whereas if the ceramic has a high resistance to thermal shocking, then material is removed by melting, vaporization and ejection, and the surface microstructure consists of melted and resolidified material [1]. If the ceramic contains a sufficient amount of a low-melting point phase, then liquid formation prevents spalling, and material is simply dislodged from the surface.

There have been very few studies on the effect of EDM on the strength of ceramics. The strength of siliconized silicon carbide has been reported to be similar under electrodischarged and conventionally ground conditions [1, 6–8]. Titanium diboride had a strength of 305 MPa after wire-cutting EDM, compared with a strength of 396 MPa after diamond grinding [5]. The composites ZrO_2–NbC [12] and ZrB_2-based [14] also have a lower strength in the EDM condition. The present work investigated the strength and microstructure of titanium diboride after EDM using the die-sinking technique and after conventional diamond grinding.

Titanium diboride powder (grade HCT-S) was supplied by Union Carbide. It had a mean particle size of 0.7–0.9 μm, surface area of 0.5 $m^2 g^{-1}$ and composition (in wt %): TiB_2 96.85, N 0.7, O 1.7, Fe 0.05, Si 0.15, Al 0.05 and others 0.5. Cold-pressed discs about 25 mm in diameter and 3 mm thick were isostatically pressed at 200 MPa, held in a bed of TiB_2 powder and sintered in argon at 2100 °C. From the measured density the porosity was calculated to be 4.4%. A surface grinder with a resin-bonded wheel (grit size D126, concentration C75) was used to diamond grind the fired discs. EDM used a Wickman Electropulse 1000 series 30 EDM1 die-sinking model fitted with a vertically operated servocontrolled head. The operating conditions were a brass electrode (negative), with a surface area matched to that of the discs, hydrocarbon dielectric (BP Energol HPO), periodic tool retraction to assist the removal of debris, square-wave pulse, pulse duration 10 μs, duty factor 50% and a discharge voltage of 40–60 V with currents of 1, 3 and 6 A. All of one face was eroded. Disc flexure tests used three equally spaced balls to support the specimen on the eroded surface, and a load was applied at a speed of 1 mm min^{-1} to a centrally placed ball on the opposite face [15]. Standard optical and electron microscopy, X-ray diffraction, microhardness and three-dimensional profilometry (Perthen Focodyn optical probe connected to custom-built amplifier and control system) were used to examine the material.

The roughness parameters of the machined specimens are given in Table I. Values sampled over an area 2 mm × 2 mm are given, and these are typically higher than the values taken along a line. ARa is the centreline-average, ASk is the skewness and AKu is the kurtosis of the profile. In brief, a negative ASk indicates a pitted or scratched surface, and the more negative the value of ASk is, the greater the number of pits to peaks present in the profile. AKu is a

TABLE I Roughness of titanium diboride after EDM and after diamond-grinding

Parameter[a]	Electrodischarge machining			Diamond-grinding
	1 A	3 A	6 A	
ARa (μm)	1.9	2.7	4.0	0.33
ASk	−1.5	−0.8	−0.05	−2.3
AKu	11.3	6.3	4.2	16.3

[a] Areal values, see text for details.

measure of the number of pits on the surface [16]. A Gaussian surface has values of ASk = 0 and AKu = 3. Thus, it is seen that increasing the current increases the roughness of the surface but reduces the number of pits present and, although the diamond-ground surface has by far the lowest surface roughness (ARa = 0.33 μm), it has the greatest number of pits.

Microstructural details of the surface are shown in Figs 1 and 2. Grinding produced large grooves and many microcracks (not evident in Fig. 1a). EDMed surfaces showed a brittle intergranular fracture with a few cleaved grains (Fig. 1b–d).

Smeared grains occurred in specimens machined at 1 and 3 A, and the number of grains with a smeared surface increased as the current decreased. Small cracks were sometimes present in the smeared surfaces (Fig. 1c). Deep cracks (>30 μm) occurred in the ground specimens (Fig. 2a). The EDMed specimens had fewer cracks, but there was a damaged heat-affected zone about 15 μm thick (Fig. 2b). The microhardness [H_v(1.0, 30 s)] of the ground and and EDMed (1 A) surfaces were 2505 ± 360 and 2380 ± 770, respectively, indicating that no major change in structure had occurred. The surfaces of the specimens EDMed at 3 and 6 A were

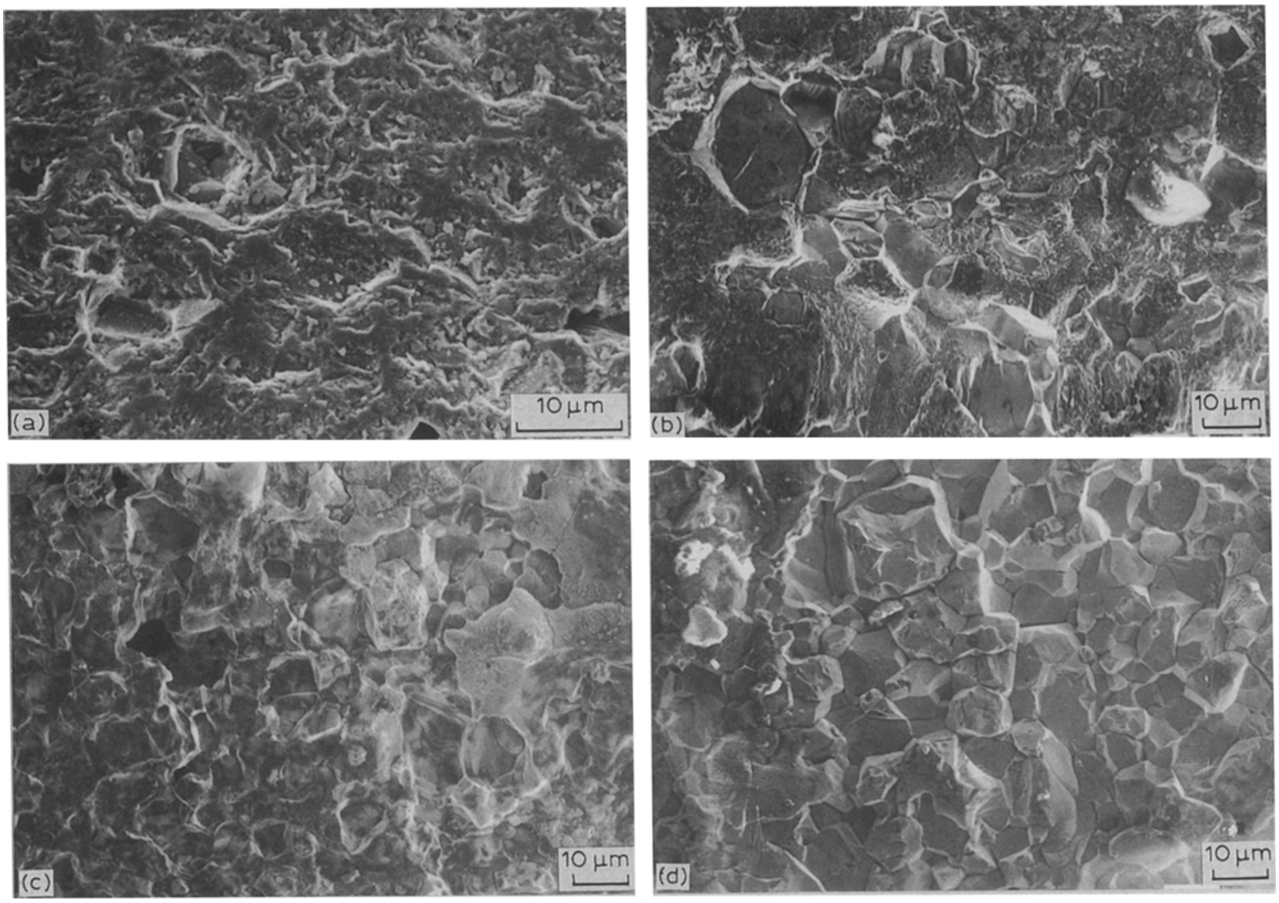

Figure 1 Microstructures (scanning electron microscopy) of the machined surfaces: (a) diamond-ground, and EDMed at (b) 1 A, (c) 3 A and (d) 6 A. Note the clean fracture and the absence of smeared grains in (d), the presence of a few smeared grains in (c) and the large number of smeared grains in (b).

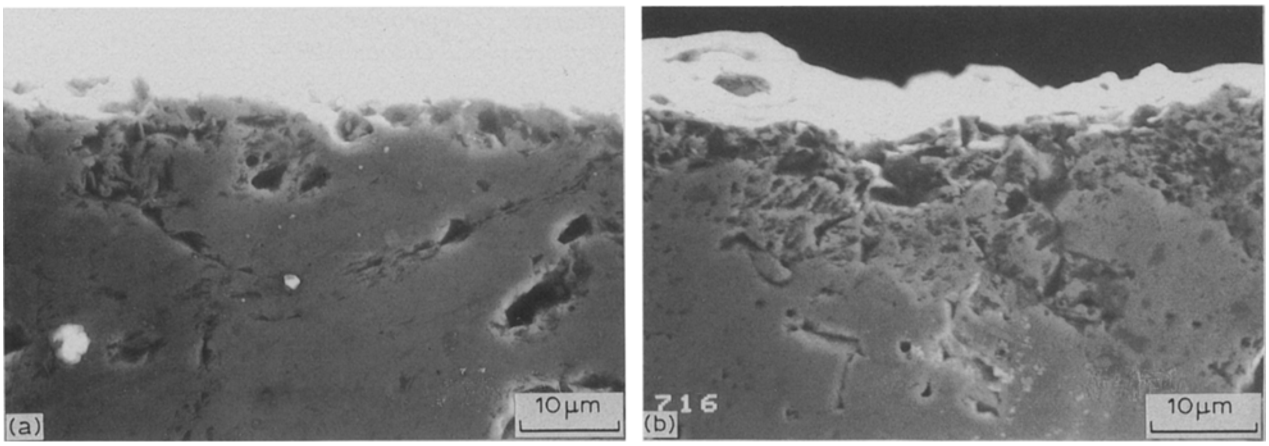

Figure 2 Polished sections (scanning electron microscopy, unetched) of the machined surfaces: (a) diamond-ground and (b) EDMed at 6 A.

too rough to allow microhardness measurements. X-ray diffraction also showed that no compositional changes occurred on the EDMed surfaces.

It is clear that the mechanism of material removal is by spalling [9]. What is less clear is the origin of the smeared regions observed at low sparking currents (Fig. 1b and c). It may be that the pulse energy at these low currents was insufficient to produce a level of thermal stress to cause fracture, and some locally melted material smeared over the grain and resolidified. The small cracks occasionally observed in the smeared regions (Fig. 1c) could be solidification cracks, and hence lend support to the present ideas.

The mean strength of the material is shown in Table II. Thus, EDM at 3 A reduced the strength of titanium diboride by 12%. This compares with previous work on wire-cut machining, of a reduction of 23% from 398 to 305 MPa for titanium diboride [5] and a reduction of 3% from 396 to 385 MPa for a ZrB_2-based composite [14].

If the reduction in strength is due entirely to the surface roughness, then with a Griffith surface crack of the form [17]

$$\sigma_f = (X/Y)(K_{Ic}/c^{1/2})$$

the reduction in strength would be approximately $(0.33/2.7)^{1/2} = 0.35$, i.e. 65%. Here we have assumed that the crack length (c) is proportional to the roughness, and that all other quantities (the constants X and Y and fracture toughness K_{Ic}) remain constant. The reduction in strength is far less than that calculated. If the reduction in strength was due to the depth of the subsurface damage or cracks, then the diamond-ground surface should be weaker. This is not the case. Thus, it appears that the reduction in strength is due to factors other than the surface roughness or subsurface cracks. One such factor is the internal stress generated upon cooling the thermally eroded surface. Ricci *et al.* [5] measured the internal stress and showed that in general the surface layers were in tension. This would weaken the material. However, the extent to which the surface roughness, subsurface damage, internal stress and other factors are responsible for reducing the strength of EDMed surfaces is, at present, unknown.

TABLE II Strength of titanium diboride after EDM and after diamond-grinding

Property	Electrodischarge machined, 3 A	Diamond-ground
Number of specimens	6	14
Mean strength (MPa)	320	362
Standard deviation (MPa)	39	47
Weibull modulus	8.4	8.3

Acknowledgements

The authors thank, at T and N Technology, Cawston, R. Wordsworth for providing the materials and facilities, and the help of L. Farthing, G. Nurse, C. Brennan, R. Wagstaff and D. Holme, and at Coventry University, the help of R. Phezey.

References

1. W. KONIG, D. F. DAUW, G. LEVY and U. PANTEN, *Ann. CIRP* **37** (1988) 623.
2. L. F. GREEN, MSc thesis, University of Manchester Institute of Science and Technology (1978).
3. A. J. AJMAL, MSc thesis, University of Manchester Institute of Science and Technology (1981).
4. C. F. NOBLE, A. J. AJMAL and A. J. GREEN, in International Symposium of Electro-machining, ISEM-7, Birmingham, 12–14 April 1983 (IFS/North-Holland, Amsterdam, 1983) p. 305.
5. W. S. RICCI, W. R. BLUMENTHAL and H. A. SKEELE, US Army Materials Technology Laboratory, Report AD-A186 989/0/GAR (September 1987).
6. D. F. DAUW, 1988, (reported in ref. 1).
7. W. KONIG and V. PANTEN, 1988 (reported in ref. 1).
8. L. WERNHONER, 1988 (reported in ref. 1).
9. G. LEVY, I. BELTRAMI and F. MAGGI, 1987 (reported in ref. 1).
10. N. F. PETROFES and A. M. GADELLA, *Adv. Mater. Manufact. Processes* **3** (1988) 127.
11. A. M. GADALLA and H. S. BEDI, *Adv. Mater. Manufact. Processes* **6** (1991) 125.
12. M. NAKAMURA, K. KANAYAMA and Y. HIRAI, *ibid.* **4** (1989) 425.
13. A. M. GADALLA and N. F. PETROFES, *ibid.* **5** (1990) 253.
14. M. NAKAMURA, I. SHIGEMATSU, K. KANAYAMA and Y. HIRAI, *J. Mater. Sci.* **26** (1991) 6078.
15. D. J. GODFREY, *Mater. Sci. Technol.* **1** (1985) 510.
16. K. J. STOUT, *Mater. Engng* **2** (1981) 260.
17. D. W. RICHERSON, "Modern ceramic engineering" (Marcel Dekker, New York, 1982).

Received 17 September
and accepted 8 October 1992