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On the Machinability of Ceramic Compacts

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Abstract

Conventional machining of compacted ceramic preforms can be a cost-effective method of making small production runs of complex-shaped engineering ceramics. This paper discusses the factors influencing chipping resistance and edge retention for two spray-dried zirconia powders. The chipping resistance did not deteriorate significantly, neither did the swarf size change after pre-sintering the components at temperatures up to 1000°C despite a twenty-fold increase in compact strength. It is argued that machinability is only indirectly dependent on compact strength but is dependent directly on the initial defect size and cutting parameters. © 1997 Elsevier Science Limited.

1 Introduction

Complex-shaped ceramic components can be made in small numbers economically by conventional machining of a blank of compacted powder[1] which can be prepared either by die pressing[2,3] or isostatic pressing.[4] A similar technique can be automated for mass-production of spark plug insulators .[5] The compact can also be partially sintered to confer additional strength.[6] Much higher removal rates can be achieved in the unfired state compared with the diamond machining of sintered ceramics.

The mechanism of material removal is one of pulverization not unlike that which accounts for the ease of machining of mica-filled glass ceramics (so-called machinable ceramics). In these materials it is the cleavage of mica and the separation of the mica–matrix interface, together with crack-linking, which produces powder-like swarf.[7,8] The machinability thus depends on the degree of interlocking of the mica crystals which usually constitute 50–60% of the material.[7] In addition to machinable glass ceramics, some boron nitride composites and Wollastonite-based ceramics are also machinable in the dense state.[9–11] In these materials, material removal is enhanced at the expense of strength by deliberate tailoring of microstructural heterogeneity to give weak interphase boundaries which enable easy grain-scale cleavage or dislodgement and the formation of powder-like swarf.

The strength of a pressed ceramic body before firing is very low, typically below 2 MPa.[12] This may result in damage during clamping of the workpiece. In contrast, high strength before firing favours handling and clamping but is likely to result in poor chipping resistance and edge retention during cutting.[7,8] Thus, by studying powders which gave normal and very high compacted strength it was suggested that the unfired strength of the compact was influential in controlling edge retention in a milling trial which was used as an indication of machinability.[2] The resistance to clamping and chipping damage are thus the two competing attributes of a ceramic powder compact beside which other machining criteria, such as removal rates and tool wear, are of secondary importance.[13,14]

2 Experimental details

Two ultrafine zirconia powders were selected (Table 1). Both powders were received in the form of spray-dried agglomerates. The ultimate particles observed by TEM were deduced from transmission electron micrographs (Table 1). SY-Ultra has a very low agglomerate strength of 0·02 MPa and narrow strength distribution.[15] It has been found to break down well in shear mixing.[16] The agglomerates of a powder produced by an identical route to YZ5N have a wide range of strength in excess of 0·22 MPa and many agglomerates were found undeformed even at die-pressing pressures over 10 MPa.[15,16]

The powders were pressed in a rectangular die $50 \times 8{\cdot}5$ mm with average depth 6·8 mm. The pressed bars were tested on an Instron testing machine (Model TM-M, Wycombe, UK) for three-point flexural strength using a span of 30 mm and cross-head speed of 0·5 mm min^{-1}.

Table 1. Details of ceramic powders

Powder	*Average particle size (μm)*	*Average agglomerate size (μm)*	*Yttria addition (wt%)*	*Manufacturer*
SY-Ultra (5·2)	0·3	31	5·1	Z-Tech, Australia
YZ5N	0·07	20	3·5–5·0	Tioxide Specialties Ltd., UK

The machining was performed on a Harrison vertical milling machine using a 25·4 mm diameter insert tip cutter fitted with a single tip. The tips were triangular carbide inserts (Type S25M, ex Seco Tools AB, Sweden) and a fresh tip was used for each sample. A pressed bar was fixed with double-sided adhesive tape and was surrounded with a container to retain the swarf.

The machining conditions consisted of a cutter speed of 1500 rpm giving $2\,ms^{-1}$ tip speed and a feed rate of $2{\cdot}5\times10^{-3}\,ms^{-1}$. The top surface of a sample was machined to provide a reference plane for accurate measure of depth of cut [Fig. 1(a)]. A step was machined with a depth of cut of 2 mm [Fig. 1(b)]. A second step was then machined with 1 mm depth and width of cut [Fig. 1(c)]. Out of the three edges created as shown in Fig. 1(c) as I, II and III, edge II was machined with identical machining conditions as well as depth and width of cut for every sample. Comparison was therefore possible by examining with an optical microscope (Wild M3Z Heerbrugg, Switzerland) and by examining the swarf collected during the machining of this step using SEM (Cambridge S250, UK).

Following this procedure, the effect of compaction pressure and pre-sintering on compact strength and machinability were studied for compaction pressures of 25, 50, 75 and 100 MPa and for the SY-ultra powder at pre-sintering temperatures of 500–1200°C. The fracture surfaces of the pressed bars were examined using SEM (Cambridge S250, UK).

3 Results

The edge retention was examined by observing the degree of damage at edge II [Fig. 1(c)]. For powder SY-Ultra, there is no significant difference in edge retention between samples pressed from 25 MPa to 100 MPa as shown by Fig. 2. Powder SY-Ultra is an unusual one in that it is quite difficult to detect agglomerate relics in the fracture surface of compacts prepared at pressures above 25 MPa (Fig. 3). The fracture was almost entirely trans-agglomerate even at 25 MPa [Fig. 3(a)]. This is attributable to the low agglomerate strength previously reported.[15,16] The compacts presented a low flexural strength (Table 2) and in principle this should make it resistant to chipping during machining before firing.[2]

It was expected that edge retention would deteriorate as the compact strength was increased so attempts were made to increase the strength in several ways. Increasing the compaction pressure above 100 MPa produced problems of sample integrity during ejection for this powder and so was abandoned. An attempt was made to increase the unfired strength by saturating the compact with water and then allowing it to dry to cause redistribution of water-soluble organic matter at particle contacts; this had negligible effect on strength and no effect on machinability. The compacts were then heat-treated for 1 h before the machining trials and Table 2 shows the increase in compact strength with heat treatment temperature, particularly in the 850–1000°C region reflecting the onset of neck growth.

Surprisingly, no apparent effect was observed on edge retention due to the increase in compact strength by heat treatment and this is shown in Figs 4(a) and (b) which should be compared with Fig. 2. Even the sample sintered at 1200°C was successfully machined for a few millimetres and the cutting was accompanied by a dramatic production of sparks. In this case the machining forces were

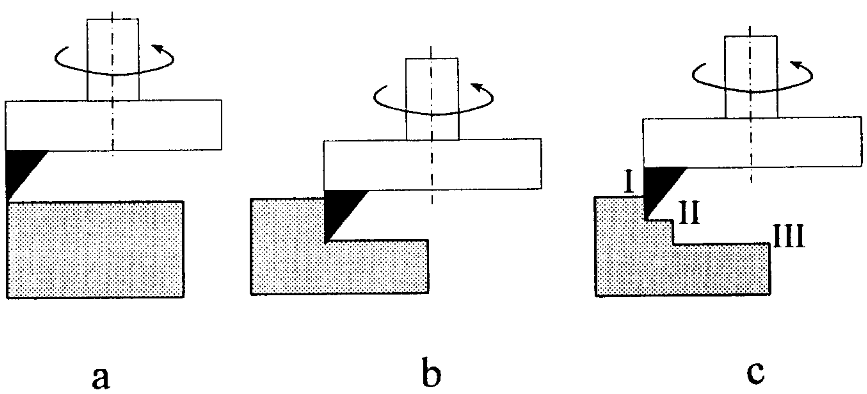


Fig. 1. Machining sequence for the steps: (a) machining of top reference surface; (b) machining of a 2 mm depth step; (c) machining of a step with 1 mm depth and width.

Table 2. Flexural strength of SY-Ultra compacts subjected to various compaction pressures and heat treatments (1 h at temperature)

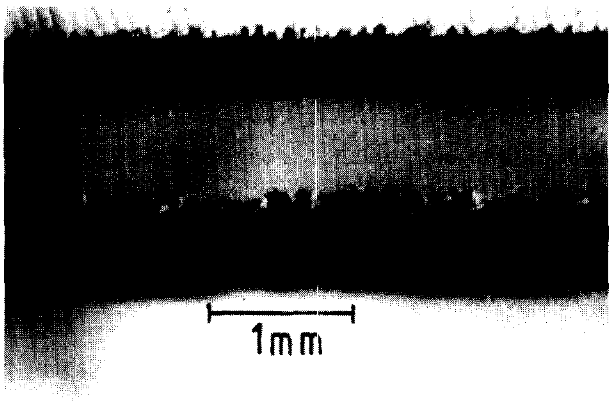
Compaction pressure (MPa)	*Heat treatment (°C)*	*Failure stress (MPa)*		
		Mean	*S.D.*[a]	n^b
25	—	0·4	0·0	5
75	—	0·5	0·1	6
75	500	0·8	0·1	5
75	600	1·0	0·1	5
75	700	1·4	0·1	5
75	750	1·4	0·3	4
75	850	4·0	0·4	4
75	1000	10·3	2·2	4

[a]Standard deviation
[b]number of tests

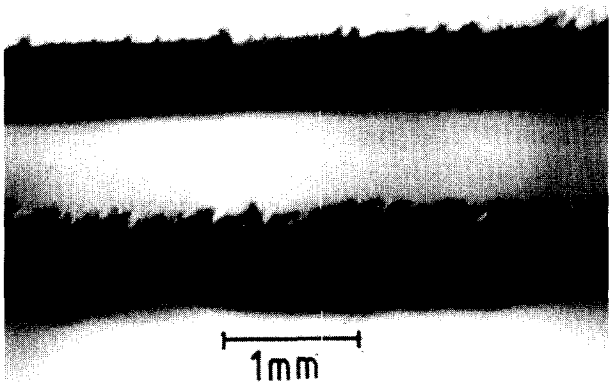
too great for the method of clamping and the sample came loose and fractured. The edge that was produced was of comparable quality to those produced in samples sintered at lower temperatures.

What is perhaps more surprising was that the size of the swarf generated by milling did not change significantly for these samples. Swarf collected during machining of edge II of compacts which had been subjected to different compaction pressures and heat treatments was examined by SEM (Fig. 5) which indicated that there was no significant difference in the size range of fragments. Thus compacts pressed at 75 MPa having a strength of 0·5 MPa produced similar edge retention and swarf size as compacts heat treated at 1000°C which presented strengths of 10 MPa for the SY-Ultra powder.

Powder compacts from YZ5N behaved very differently. The unfired strength (Table 3) had a comparable value to that of SY-Ultra at low compaction pressure but increased more rapidly when pressed at higher pressures. Figure 6 shows the excellent edge retention at lower pressures and a clear trend of deterioration as compaction pressure increased. The agglomerate strength of YZ5N powder was higher and had a wide distribution as

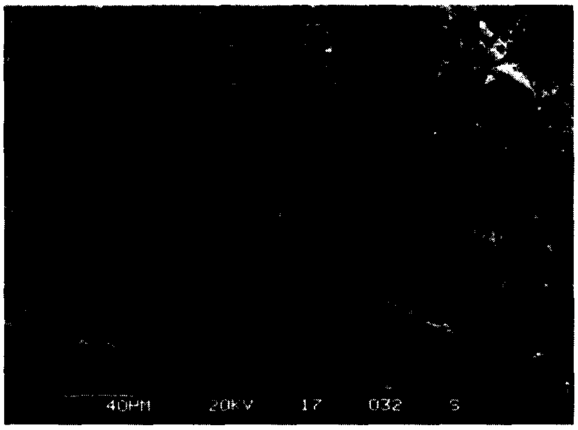


(a)

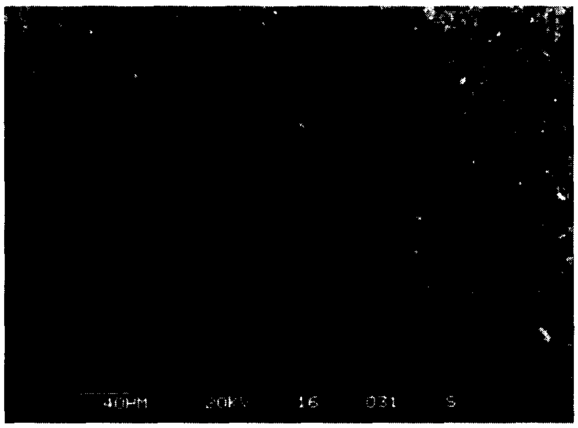


(b)

Fig. 2. Optical micrographs showing a similar degree of edge retention at edge II (the lower edge) for powder SY-Ultra at compaction pressures (a) 25 MPa and (b) 100 MPa.



(a)

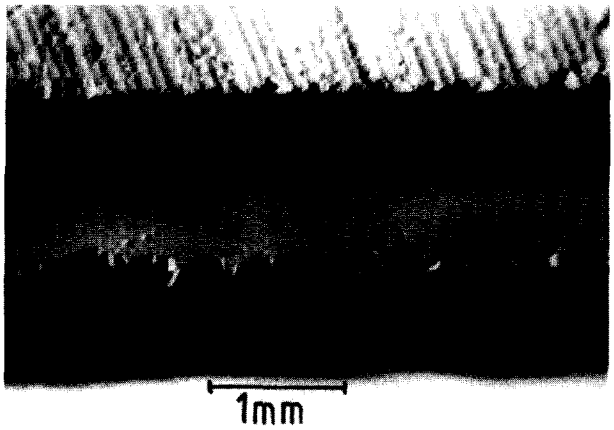


(b)

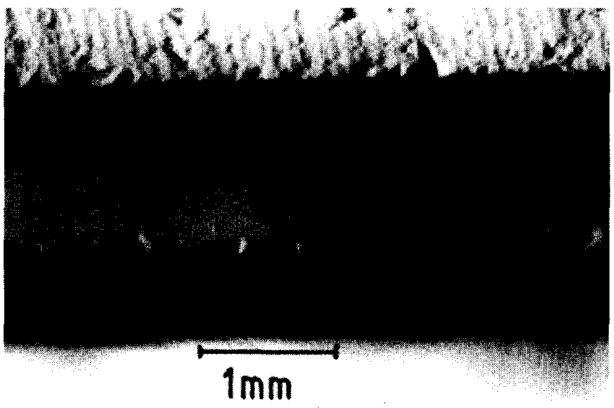
Fig. 3. SEM of fracture surface of SY-Ultra compacts pressed at (a) 25 MPa and (b) 100 MPa.

Table 3. Flexural strength of YZ5N compacts subjected to various compaction pressures

Compaction pressure (MPa)	*Failure stress (MPa)*		
	Mean	*S.D.*	*n*
25	0·35	0·08	4
50	0·55	0·09	5
75	0·61	0·13	5
100	0·70	0·15	4

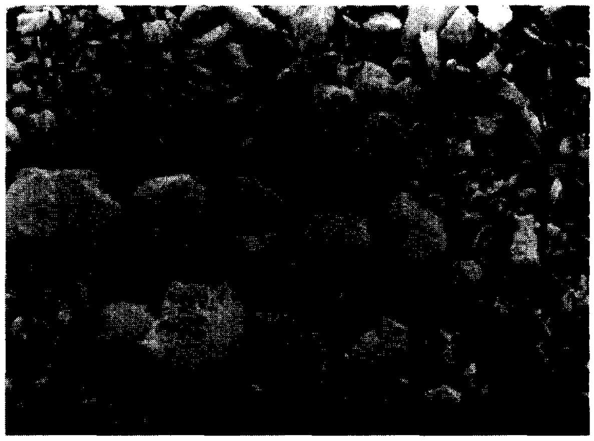


(a)



(b)

Fig. 4. Machined edges from SY-Ultra compacts pressed at 75 MPa and heat treated for 1 h at (a) 750°C and (b) 1000°C.



(a)

(b)

Fig. 5. SEM of swarf collected during machining of edge II of SY-Ultra compacts, (a) as pressed to 25 MPa and (b) as pressed to 75 MPa and sintered for 1 h at 1000°C.

indicated by tests on a different batch of powder produced by the same route[15,16] and confirmed by the very different characteristics of the fracture surfaces as shown in Fig. 7 compared with those in Fig. 3. Agglomerates were only slightly compacted at low pressure [Fig. 7(a)] and failure was clearly dominated by inter-agglomerate fracture. A mixture of trans-agglomerate and inter-agglomerate fracture was found at medium pressure [Fig. 7(b)] and trans-agglomerate fracture prevailed at high pressure [Fig. 7(c)].

The features of the swarf collected from compacts of YZ5N correlated well with the edge retention studies. Fine swarf which retained the characteristics of the original agglomerates resulted from inter-agglomerate fracture during cutting [Fig. 8(a)]. Indeed Fig. 8(a) shows some whole undamaged agglomerates which have been abstracted from the compact. The deterioration of the machined edge with increase of compaction pressure was accompanied by a higher proportion of trans-agglomerate fracture and the size and shape of swarf changed from equiaxed to large irregular chips with sharp edges [Figs 8(b) and (c)].

4 Discussion

The problem now is to interpret these results in the context of previous studies and of the general observation that a high compact strength results in poor resistance to chipping.

The results for SY-Ultra sever the direct causal connection between chipping resistance and compact strength since heat treatment allowed strength to be increased by a factor of 20 without deterioration in edge retention and without changing the size and shape of the swarf. On the other hand, the use of compaction pressure to increase compact strength in powder YZ5N supports such a connection.

The presence of organic matter, such as dispersant added for spray drying or binder added as a compaction aid, had no direct effect on edge retention of SY-Ultra compacts. Multiple loss on ignition of SY-Ultra suggested the initial presence of 0·3wt % water and 0·9wt % organic matter; clearly by 500°C organic additives had departed. It is arguable that the small amount (typically $<0{\cdot}1$wt %) of inorganic ash in polymeric additives may

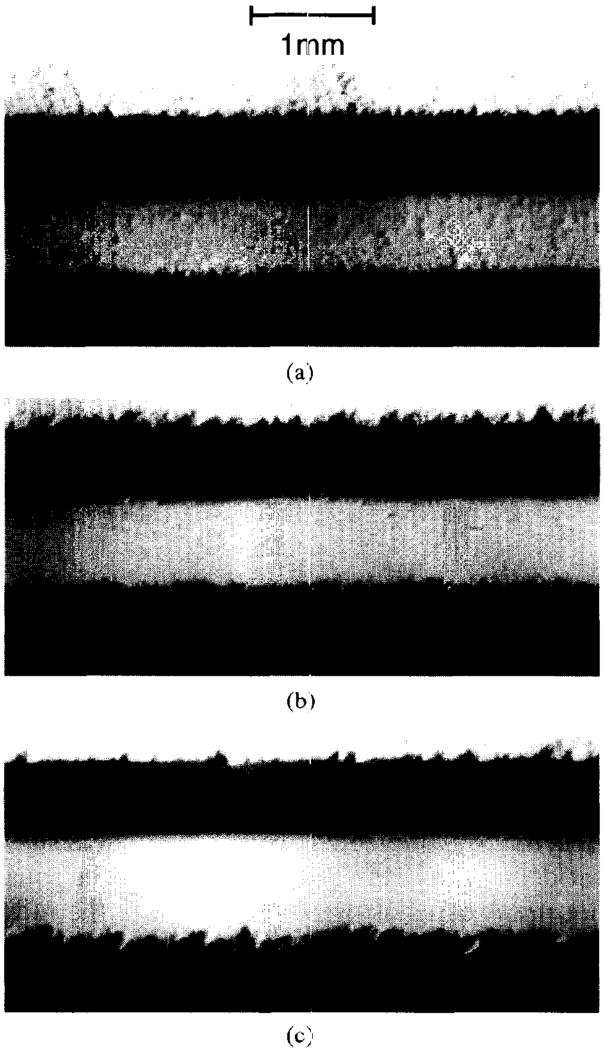


Fig. 6. Machined edges II (the lower edges) from compacts of YZ5N pressed to (a) 25 MPa, (b) 50 MPa and (c) 100 MPa.

have a slight effect on interparticle adhesion but the removal of 1wt % polymer which corresponds to 5 vol% had no effect on machinability. This is not to say that such additives did not, at an earlier stage, have a pronounced effect on deciding the structure of the spray-dried agglomerate by influencing dispersion in the slurry nor that they did not influence structure in the compact through die wall and interparticle lubrication. It is to claim that their role, although initially influential in deciding the structure of the compact, is no longer relevant after pyrolysis.

At first sight, these results seem to contradict our previous study of machining in which different powders with low and high compact strength were compared.[2] However, several factors influence the strength of a compact, among them agglomerate strength, agglomerate boundary strength and the size of intra- and inter-agglomerate flaws. In the present series of experiments, different strength is obtained in the SY Ultra powder by pre-sintering and the sintering schedule is sufficiently gentle not to have a great deal of influence on the initial critical defect size.

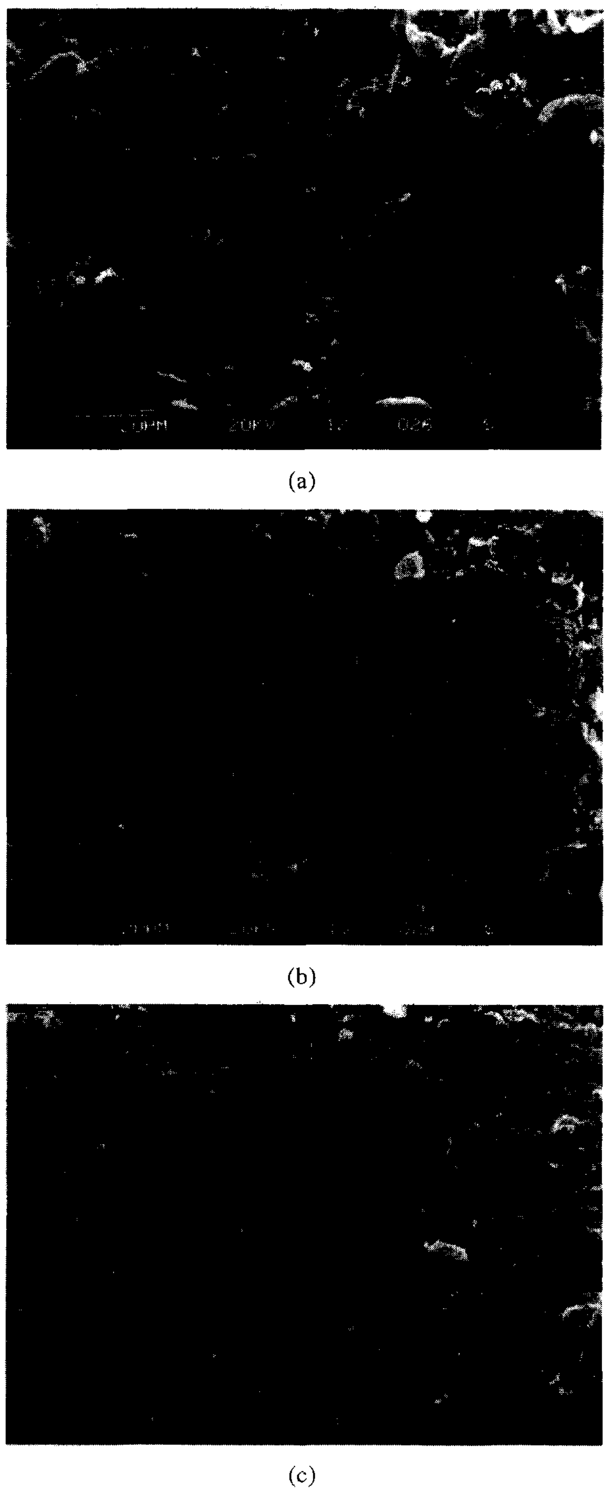


Fig. 7. Fracture surfaces of YZ5N compacts pressed to (a) 25 MPa, (b) 50 MPa and (c) 100 MPa.

In the second powder studied, strength is changed by increasing compaction pressure which, as shown by fractography, has a pronounced effect on critical defect size. It is argued below that under certain conditions, final crack length and hence machining quality and swarf size are influenced principally by initial crack length. The fact that initial crack length is one of the factors influencing strength leads to an apparent link between strength and machinability.

The size of the swarf is related to the crack propagation area during cutting, which should be restricted to avoid formation of large chips so that intricate shapes and details can be machined. The stress state in a compact near the tip of the cutting tool during machining is very complex and is discussed below in terms of a much simplified model which draws on an interesting aspect of Griffith's theory.

A general feature of Griffith's theory for unstable crack propagation is that final crack length is inversely related to initial crack length. A low initial defect size confers high strength which means more stored strain energy is available at the point of propagation. This idea has been developed by Hasselman[18–20] in the analysis of thermal shock damage. Hasselman considered that the material contained N circular and uniformly distributed microcracks of identical initial radius, l_o, per unit volume. The body is uniformly cooled with the surfaces rigidly constrained. Simultaneous propagation of all cracks occurred with negligible interaction between the stress field of neighbouring cracks.

The final crack length, l_f, is related to initial crack length, l_o, by

$$l_f \alpha (l_o N)^{-1/2} \tag{1}$$

but not to any other material property apart from Poisson's ratio.[20] High strength favours resistance to the onset of fracture but low initial strength reduces the extent of thermal shock damage by reducing the final crack length.

In the derivation of eqn (1), it was assumed that $l_f >> l_o$ and that the stored elastic strain energy was negligibly small once the final crack length, l_f, was reached. A similar result can be obtained directly from Griffith's theory under certain conditions.

Consider a central crack of initial half-length C_o in a thin plate with unit thickness and area A which is subjected to a uniaxial stress σ perpendicular to the crack. For a crack which is initially small, the final crack length C_f can be estimated by balancing the energies of the system before and after propagation as given by[17]

$$\frac{\sigma_o^2}{2E}(A + 2\pi C_o^2) + 4\gamma C_o = \frac{\sigma_f}{2E}(A + 2\pi C_f^2) + 4\gamma(C_f - C_o) \tag{2}$$

Here σ_o is the initial fracture strength corresponding to the initial crack length C_o given by Griffith's equation

$$\sigma_o = \left(\frac{2E\gamma}{\pi C_o}\right)^{1/2} \tag{3}$$

The stress σ_f however is undefined. Its lower limit is zero corresponding to negligible stored strain energy after propagation as assumed by Hasselman.[20] Its upper limit is the failure stress associated with the final crack length C_f. Thus

$$0 < \sigma_f < \left(\frac{2E\gamma}{\pi C_f}\right)^{1/2} \tag{4}$$

Substituting these bounds into eqn (2) and assuming $C_f >> C_o$ gives for C_f respectively:

$$C_f = \frac{A}{4\pi C_o} + \frac{C_o}{2} \tag{5a}$$

and

$$C_f = \frac{A}{6\pi C_o} + C_o \tag{5b}$$

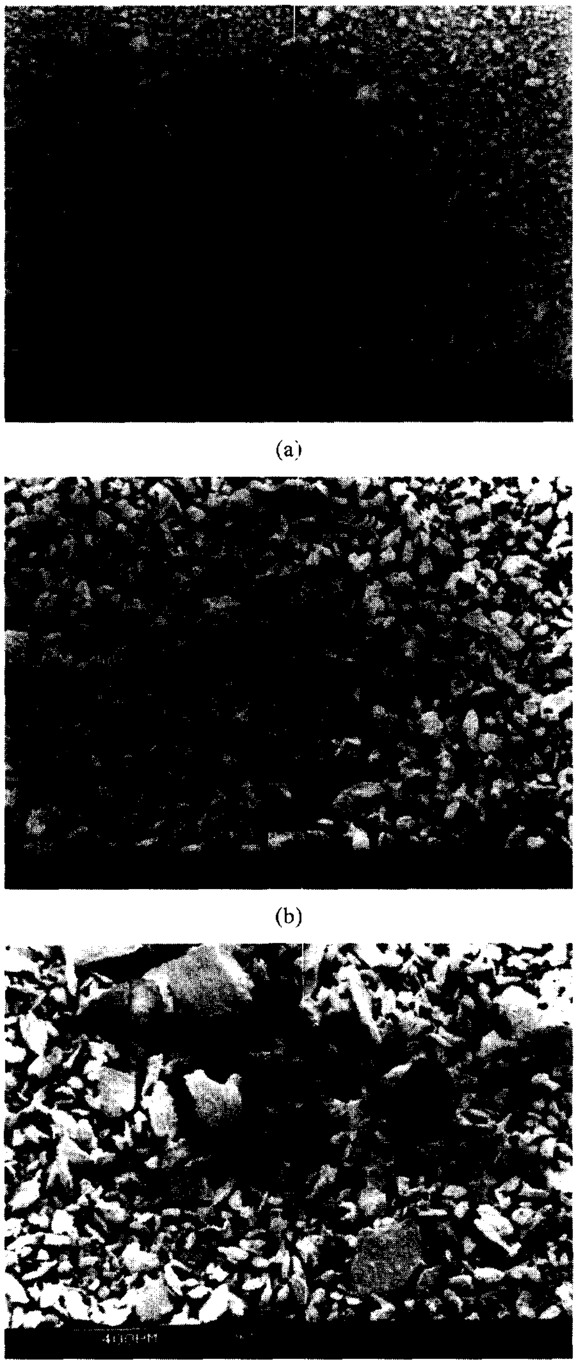
In each case the final crack length is inversely proportional to initial crack length and other material properties, notably E and γ do not influence it for the conditions stipulated. This is the same result as that deduced by Hasselman noting that the square root in his eqn (8)[20] arises because l_o is the radius of a circular crack rather than the length of a crack of unit width.

The machining situation differs from thermal shock damage because the cutter adds stored strain energy to a small volume of material at a time rather than to a thin region over the entire surface concurrently. As with Hasselman's analysis, isolated cracks are considered. In the case of machining, the density of microcracks and hence crack linking, are undoubtedly important features of chip production. The micrographs (Fig. 7) show a high density of cracks giving, typically, a volumetric crack area of $3 \times 10^5 \mathrm{m}^{-1}$ for inter-agglomerate boundaries alone. Generally both the critical defect size and the defect density decrease with increasing applied compaction pressure as observed directly by Uematsu *et al.*[21] This means that more elastic strain energy is stored in front of the cutting tip before propagation occurs. Once cracks start to propagate they tend to travel further to dissipate stored energy before linking to form a chip. This results in the creation of larger chips.

However, although sintering increases elastic modulus and fracture energy, it has little effect on the size of the critical defect or the defect density. This is certainly the case in this work where the sintering schedules involve small linear shrinkages (Table 4). Thus while the strength changed with pre-sintering temperature, the chipping damage was hardly affected.

Table 4. Density of SY-Ultra compacts

Compaction pressure (MPa)	*Heat Treatment Temperature (°C)*	*Density* (kgm^{-3})	*S.D.*	*n*	*Relative density*
25	—	2306	17	5	0·384
50	—	2467	28	4	0·411
75	—	2629	35	6	0·438
100	—	2642	33	6	0·440
75	500	2686	46	5	0·448
75	600	2705	23	5	0·451
75	700	2727	16	5	0·455
75	750	2719	28	5	0·453
75	800	2702	—	—	0·450
75	850	2700	7	3	0·450
75	1000	2806	5	5	0·468
75	1200	3854	—	1	0·642



(a) (b) (c)

Fig. 8. SEM of swarf collected during machining of YZ5N compacts pressed to (a) 25 MPa, (b) 50 MPa and (c) 100 MPa.

The requirements for machinability are thus less contradictory than previously thought. High strength is desirable to resist clamping damage and favours low critical defect size. Low strength was thought to be necessary to resist chipping damage. In fact it appears to be high critical defect size and defect density which is necessary to resist chipping damage.

Inter-agglomerate boundaries which are extant after compaction provide suitable interlinking defects[2] and the existence of weak inter-agglomerate boundaries favours edge retention. They depend on the compaction pressure as well as powder agglomerate strength and relative strength of adhesion between agglomerates and particles as demonstrated above for powder YZ5N. For powders with low agglomerate strength, where trans-agglomerate fracture prevails, defects depend on the interparticle packing which may be influenced by particle size, shape[22,23] and the use of binder as a lubricant.[24]

A low cutting tool feed rate or depth of cut is preferable for finer chipping and hence better edge retention. Figure 9 shows the difference in edge quality for two depths of cut. Cutting tools with ultrasonic vibration have been successfully used in

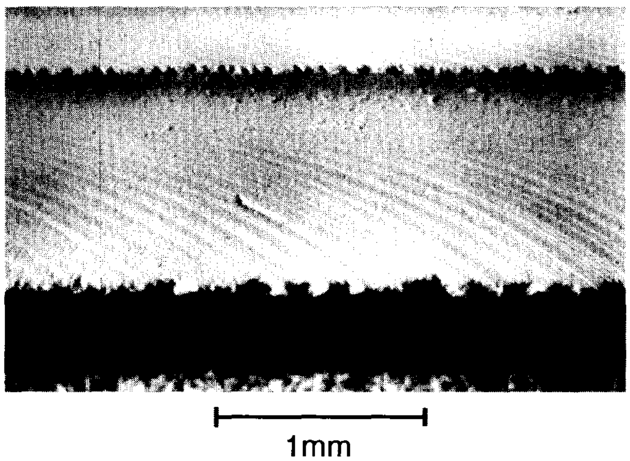


Fig. 9. Edge quality at two cutting depths: 1 mm for the upper edge and 2 mm the lower edge (YZ5N pressed to 50 MPa).

machining of intricate grooves or slits with sharp inside corners where the effective depth of cut was less than 20 μm[25]. It is also important to keep the cutting tool sharp to introduce stress concentration.

One way to increase compact strength to resist clamping damage is by using high strength binder systems[3] which can also be modified by plasticisers. Fine powders, especially with inorganic surface coatings tend to give higher compact strength.[3,26] Use of vacuum arbors and rubber-lined jaws or fixtures are recommended whenever possible to minimise clamping damage.[27]

Compaction pressure affects both compact strength and the critical defect length. In general, an optimum compaction pressure is selected to obtain a balance between edge retention and clamping resistance for a given powder. The final strength of a component prepared by compaction is frequently limited by the failure of compaction to consolidate agglomerates because such defects are not healed by sintering. In cases where sintered strength is less important than edge retention, these results show that relatively low compaction pressure can be used and accompanied by pre-sintering, if necessary, to enhance clamping ability.

5 Conclusions

Observations of machined edges, fracture surfaces and swarf of machined powder compacts indicate that the edge quality is determined by the original critical defect length in the compact preforms and by the cutting conditions. Weak inter-agglomerate boundaries are favourable for better edge retention.

Heat treatment may strongly affect the elastic modulus and cleavage energy and hence have a dramatic effect on compact strength but it has little effect on edge retention because it does not change the critical defect size or defect density much. Similarly, the removal of organic matter initially present at the 5 vol% level had no direct influence on edge retention but it may influence, by acting as a dispersant and a lubricant, the packing of a power and hence the defect size.

High compaction pressure increased compact strength giving better resistance to clamping damage but decreased the original critical defect length and hence caused deterioration in the edge retention. An intermediate pressure should be selected for resistance to damage in clamping and machining. Low compaction pressure can be used in situations where final sintered strength is less important compared with edge quality and pre-sintering may be used to raise strength without reducing edge retention.

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