Thermal-Shock Resistance and Fracture-Strength Behavior of Two Tool Carbides

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The thermal-stress resistance and fracture-strength behavior of TiC and WC after severe thermal shocking were investigated. The damage-resistance parameter, $(K_{Ic}/\sigma_f)^2$, for both carbides was determined over a wide range of loading rates and temperatures. The fracture-strength behavior of these two carbides in the quenching temperature range of 25° to 800°C follows Hasselman's model for an instantaneous strength loss at a critical quenching temperature. When WC was shocked at temperatures >800°C, it exhibited higher retained strengths because of its higher $(K_{Ic}/\sigma_f)^2$ values at these temperatures. The effects of specimen size and repeated thermal shocks on retained strength and critical quenching temperature for these carbides were also investigated.

I. Introduction

ALTHOUGH thermal-shock-induced fracture in carbide tools during cutting operations is not uncommon, it has received relatively little attention. Mai and Atkins[1] have successfully applied Hasselman's[2] unified theory of thermal-shock failure to examine the retained-strength (σ_a) behavior of some tool carbides after they have been shocked through their respective critical quenching temperatures (T_c). The σ_a was low because the preexisting small internal flaws in the carbides grew during the critical shock. Mai and Atkins[1] suggested that carbide tools could be compared according to both their resistance to crack initiation and their resistance to crack propagation (i.e. extent of damage). For crack initiation caused by thermal shock, the appropriate parameter is $k\sigma_f/E\alpha$, where k is thermal conductivity, σ_f tensile fracture strength, E Young's modulus, and α the coefficient of linear expansion of the ceramic under consideration. The larger the value for $k\sigma_f/E\alpha$ the lesser the chance that flaws will be initiated. Damage resistance or extent of crack propagation by thermal shock is best characterized by the magnitude of $(K_{Ic}/\sigma_f)^2$,† where K_{Ic} is fracture toughness. Ceramics can also be ranked by a more traditional method in which small specimens are successively subjected to thermal cycles with increasing shock severity. The number of cycles required to cause the appearance of surface cracks and final fracture, respectively, are recorded and the results are used to rate the ceramics.

The present paper reports the results of an experimental investigation on the thermal-shock resistance and fracture-strength behavior of two cemented carbides (a tungsten carbide and a titanium carbide). Both $(K_{Ic}/\sigma_f)^2$ and the retained strengths for these carbides are determined as a function of temperature (25° to 1000°C), and possible influences of loading rates and the effects of different surface finishes are investigated.

II. Experimental Procedure

(1) Preparation of Materials

The WC‡ was blended with 8.5% cobalt and has the composition WC/TiC/TaC/Co in the ratio (in wt%) of 72:8:11.5:8.5. The TiC§ was blended with Ni and Mo alloys. Table I compares some of their physical and mechanical properties. Both WC and TiC were prepared with two surface conditions, one the as-sintered condition and the other diamond-ground to ≈0.25 to 0.50 μm rms which approximates the surface finish of the actual tool inserts.

Table I. Mechanical and Physical Properties of Carbides

Property	TiC	WC
k (cal/s C cm)	0.04	0.11
σ_f (MN/m²)	500	1680
E (GN/m²)	446	560
ν*	0.20	0.23
α† (°C^{-1})	2.40×10^{-6}	5.70×10^{-6}
β‡	8.10	2.96

*Poisson's ratio; †coefficient of linear expansion; ‡Biot's modulus.

(2) Measurement of Temperature-Dependent and Retained Strength

Three-point bend specimens with the grinding direction parallel to the longitudinal axis were prepared by the manufacturers. The specimen edges were slightly rounded to eliminate chipping along these edges during the tests. Strength was determined between 25° and 800°C in air at a loading rate of 83.3 μm/s. The after-shock room-temperature retained strengths (σ_a) were determined by quenching from 200° to 1000°C; specimens were equilibrated in an electric furnace to the required temperature and quenched in a water bath at ≈20°C. The time required to transfer specimens to the quenching medium was ≈1 s.

(3) Determination of Fracture Toughness

Fracture toughness values for both WC and TiC were determined using bend specimens containing semielliptical surface precracks, which were introduced in the test pieces by surface indentation using a Vickers hardness tester with a 40-kg applied load and a Knoop diamond indenter. This procedure gave very consistent precracks ≈0.70 to 0.80 mm long ($2a$) and ≈0.30 mm deep (b), which could be determined from the fracture surfaces (see Fig. 1). Success with this method of precracking cemented tungsten carbides has been reported previously by Kenny[6] and Ingelstrom and Nordberg.[7] The fracture toughness (K_{Ic}) under 3-point bend cracking may be determined by[7]:

$$K_{Ic} = {}^3/_2\,\frac{XLb^{1/2}}{BW^2}\left[1.96-2.75\left(\frac{b}{W}\right)+13.66\left(\frac{b}{W}\right)^2 - 23.98\left(\frac{b}{W}\right)^3+25.22\left(\frac{b}{W}\right)^4\right]\Phi^{-1} \quad (1)$$

when Φ, the elliptical integral, is given by

$$\Phi = \int_0^{\pi/2}\left(1-\frac{a^2-b^2}{a^2}\sin^2\theta\right)^{1/2}d\theta \quad (2)$$

In Eq. (1), X is the fracture load, L the length of span, B and W specimen thickness and depth, and b and a are the semiminor and semimajor axes of the surface precrack.

Where possible, K_{Ic} was also estimated by investigating the fracture surfaces of those bend specimens that were used for measuring σ_a (Section II(2)) and establishing the flaw size from which the final break occurred. The K_{Ic} results for both TiC and WC

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†$(K_{Ic}/\sigma_f)^2$ in fracture mechanics describes the size of the plastic zone at the crack tip (Ref. 3). Recent research also shows that this quantity characterizes the transition of quasi-static cracking to generalized yielding (Ref. 4) and that it can rank the machinability of materials (Ref. 5).

‡Carboloy 370, General Electric Co., Detroit, Mich.

§Titan 80, Adamas Co., Livonia, Mich.

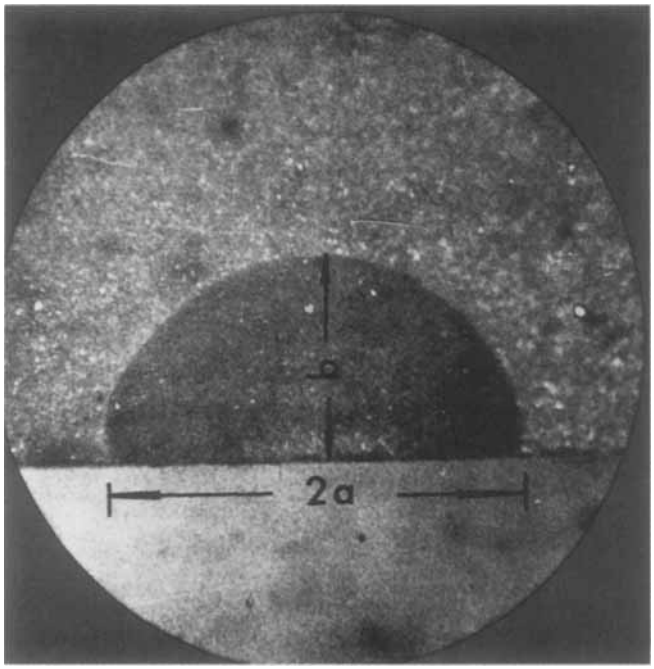

Fig. 1. A typical semielliptical surface crack introduced by the indentation technique. (TiC, ×80).

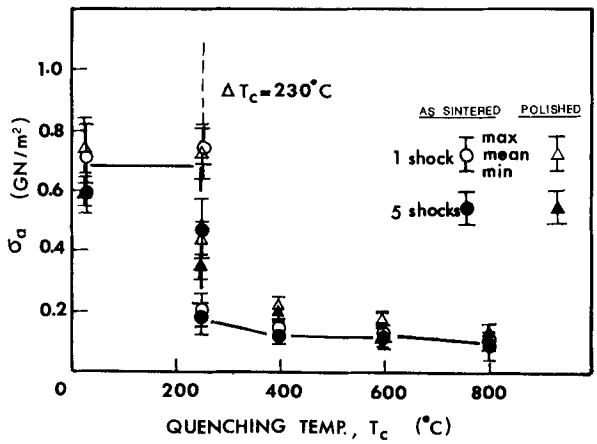

Fig. 2. Strength variation of TiC with quenching temperature.

were also obtained as a function of temperature and crosshead rate (0.05 to 25 cm/min).

III. Results and Discussion

(1) Effect of Thermal Shock, Temperature, and Loading Rate on Bend Strength

Figures 2 and 3 show the retained room-temperature strength (σ_a) plotted vs the shocking temperature interval for TiC and WC in both the as-sintered and polished surface conditions. The TiC obviously follows the Hasselman[2] model, which shows a distinct discontinuity with considerable strength loss at a critical quenching temperature (T_c). As shown by Manson,[8] this T_c may be obtained by equating the maximum surface stress during thermal shock to the strength of the carbide in the shock environment. Thus, we obtain[9]

$$\Delta T_c = (T_c - T_w) \simeq \frac{\sigma_f(1-\nu)}{3E\alpha\psi} \qquad (3a)$$

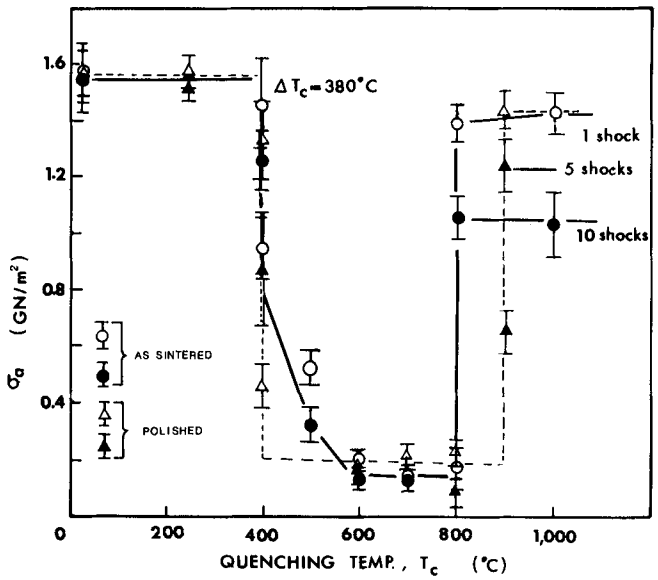

Fig. 3. Strength variation of WC with quenching temperature.

and

$$\psi^{-1} = 1.50 + \frac{3.25}{\beta} - 0.5 \exp -\frac{16}{\beta} \qquad (3b)$$

where ν is Poisson's ratio, α is the coefficient of linear expansion, T_w is the temperature of the water bath, β is the Biot modulus ($=dh/k$); d is half the thickness of the test piece, h the heat-transfer coefficient, and k thermal conductivity. Although h may be dependent on the quenching temperature difference (ΔT), values of 1 to 2 cgs units are not unreasonable.[9,*] Thus, when the values for σ_f, E, α, and ν as given in Table I are substituted into Eq. (3*a*), ΔT_c for TiC is ≈250°C when $h=1$ cgs unit and 208°C when $h=2$ cgs units. These values agree well with the experimental result (230°C) shown in Fig. 2.

The σ_a for WC below a quenching temperature of 800°C can also be described by Hasselman's model. From Fig. 3, the discontinuity in the plot of the retained strength vs shocking temperature occurred at ≈400°C. This result compares fairly well with the theoretical calculations which give $\Delta T_c = 375$°C for $h=1$ cgs and 310°C for $h=2$ cgs.

It should also be noted that, for both WC and TiC, T_c does not appear to depend on the surface-finish condition (see Figs. 2 and 3), thus indicating that the original internal flaw sizes are the same. This similarity in flaw sizes is also supported by the fact that the unquenched strengths of both carbides, as shown in Figs. 2 and 3, are similar for the two surface finishes.

Experiments have shown that loading rates within the range 8.33×10^{-6} to 2.5 m/s have negligible effects on σ_f for both the as-sintered and polished WC and TiC. Three-point unshocked bend strengths† of TiC and WC (for specimens 6.5 by 5.6 by 38 mm) in the range 25°C (*RT*) to 800°C in both the as-sintered and polished condition remained constant to ≈600°C. The strength of TiC was 480 to 500 MN/m², independent of surface finish, whereas the strength of WC was ≈1.5 and ≈1.7 GN/m² in the as-sintered and polished conditions, respectively. At 800°C, both carbides display some strength loss as a result of subcritical crack growth caused by plastic deformation within the metal phase; nonlinearities were observed in the load/time traces before final fracture.

*It has been suggested (Ref. 10) that, for a water quench, $h=0.1$ cgs unit. However, for this value of h, the critical quenching temperature differences for TiC and WC are 600° and 1700°C, respectively. Since these results are far greater than those observed experimentally, $h \simeq 1$ to 2 cgs units (as suggested in Ref. 9) was chosen in the present ΔT_c calculations.

†Note that in smaller specimens (5 by 2.8 by 30 mm for WC and 2.5 by 2.5 by 25 mm for TiC) the unshocked bend strengths are 25% and 100% stronger for WC and TiC, respectively.

As shown in Figs. 2 and 3, σ_a for WC and TiC is independent of the number of repeated shocks for 25°C$<T<$800°C. However, at ≈800°C, thermal fatigue effects caused by repeated shocks were observed in WC since internal flaws grew during each shock. The retained strength must, therefore, decrease with the number of repeated shocks, as shown in Fig. 3.

(2) *Fracture Toughness Results*

As pointed out by Ingelstrom and Nordberg,[7] valid K_{Ic} results can be obtained for the two carbides only when the residual stresses around the indentation precrack are relieved. Thus, the specimens were annealed at 800°C for 30 min and fractured in 3-point bending in the range 25° to 1000°C. Values for K_{Ic} thus obtained for WC and TiC (both surface finishes) were ≈13 MN/m$^{3/2}$ and 7.5 MN/m$^{3/2}$, respectively, and remained relatively constant throughout the range 25° to 800°C. Much higher K_{Ic} values were measured for both carbides above 800°C, e.g. at 1000°C, K_{Ic} was 10 and 15 MN/m$^{3/2}$ for TiC and WC, respectively. It seems that such increases in K_{Ic} are caused either by crack-tip blunting resulting from plastic deformation in the metal binders or by multiple cracking caused by separation of grain boundaries.

Useful estimates of K_{Ic} were also obtained from the σ_a experiments for the thermally shocked carbide specimens. When the final crack size (r_f) and geometry (i.e. part-through or through crack) from the fracture surfaces of the shocked specimens were noted and the corresponding σ_a in 3-point bending was used, K_{Ic} was obtained by plotting $\sigma_a Y$ vs $(\sqrt{r_f})^{-1}$. Note that Y is a correction factor for the particular crack geometry of the after-shock test sample and may be found from the work of Brown and Srawley.[11] For instance, Fig. 4(*A*) shows such a plot for the as-sintered WC; the slope of this line gives a toughness value of ≈13.33 MN/m$^{3/2}$. Figure 4(*B*) shows 2 typical fracture surfaces from which r_f can be measured. Similarly, the K_{Ic} value for TiC deduced from σ_a experiments is ≈7.2 MN/m$^{3/2}$. These K_{Ic} results agree very well with those obtained previously from specimens free of residual stress effects. The K_{Ic} values for both the as-sintered and polished TiC and WC were not affected by the loading rates used in the testing machine.*

(3) *Relation Between Strength Loss and $(K_{Ic}/\sigma_f)^2$ Parameter*

According to the analysis of Hasselman,[2] the fractional σ_a of a rectangular beam subjected to critical thermal shock is approximated by[12]

$$\frac{\sigma_a}{\sigma_f}=\left[0.32\pi(1-\nu^2)\frac{N}{V}\right]^{1/4}\left(\frac{K_{Ic}}{\sigma_f}\right)^{3/2} \qquad (4)$$

where N/V is the number of cracks per unit volume of the stressed specimen. Since N/V is difficult to measure, an approximate relation in which $N/V\simeq N_s^{3/2}$ is used, where N_s is the area density of cracks measurable from the shocked samples. This approach was used to measure N_s for TiC and WC at their respective T_c; the resultant values were converted to N/V, which was 7.6×10^7 cracks/m^3 and 1.5×10^8 cracks/m^3, respectively. Hence, in conjunction with appropriate $(K_{Ic}/\sigma_f)^2$ values at their corresponding ΔT_c from Table II, σ_a/σ_f for TiC and WC is 0.23 and 0.12, respectively. These results should be compared with the observed values of 0.20 to 0.25 for TiC and 0.08 to 0.16 for WC.

The considerable strength regained by thermally shocked WC samples above 800°C was suggested previously to be a result of the oxidation effect.[1] However, it is now believed that since, as shown in Table II, the $(K_{Ic}/\sigma_f)^2$ values above 800°C are much greater than the corresponding value at room temperature, the cracks propagated after shocking at these high temperatures must be much smaller than those shocked through lower temperatures (where the $(K_{Ic}/\sigma_f)^2$ values are lower). Consequently, these shocked specimens display higher σ_a, as shown in Fig. 3.

(4) *Effects of Size on σ_a and ΔT_c*

Three major effects are associated with the size of the ceramic specimens: (1) Smaller specimens give higher σ_f values, presumably a result of a statistical distribution of flaws (Weibull's[13] weakest-link theory). (2) When quenching conditions are equal, smaller specimens—because of their smaller Biot modulus (β)—will yield larger ΔT_c values. Table III shows such results on ΔT_c

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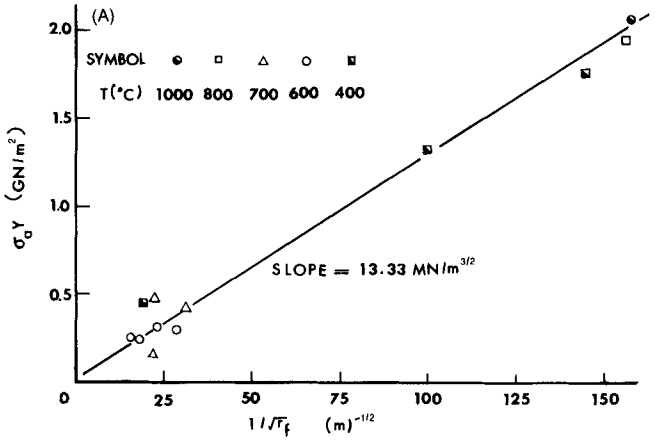

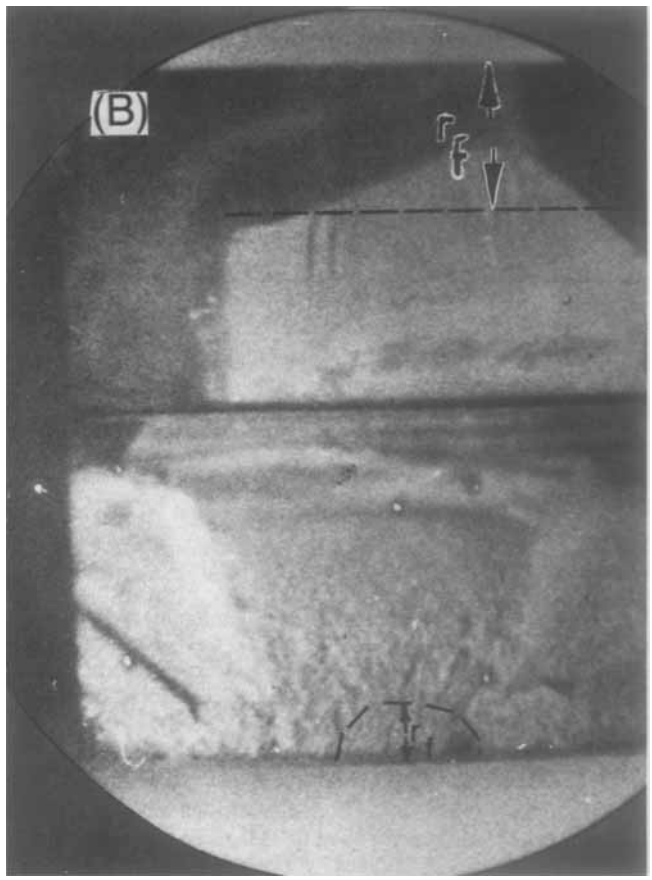

Fig. 4. (*A*) "Corrected" stress $\sigma_a Y$ plotted vs $1/\sqrt{r_f}$ for as-sintered WC. (*B*) Typical fracture surfaces with thermal-shock-induced cracks.

Table II. Variation of $(K_{Ic}/\sigma_f)^2$ with Temperature

Material	Temperature (°C) 25–400	600	800	1000
	$(K_{Ic}/\sigma_f)^2$ (μm)			
WC				
Polished	52.0	53.0	55.0	110
As-sintered	59.5	60.2	69.6	117
TiC				
Polished	136	136	250	870
As-sintered	136	136	250	692

Table III. Size Effects on Experimentally Observed Values for ΔT_c and σ_a for Titanium and Tungsten Carbides

Material	Quantity	Specimen size* A	B	C
WC	ΔT_c (°C)	†	550	380
	σ_a (MN/m²)	†	500	120–260
TiC	ΔT_c (°C)	400	450	230
	σ_a (MN/m²)	150	120	120–200

*A = 2.5 by 2.5 by 25 mm; B = 5 by 2.8 by 30 mm; C = 6.5 by 5.6 by 38 mm.
†"Continuous" strength behavior, no instantaneous drop observed.

Table IV. Evaluation of Thermal-Shock Resistance for WC and TiC

Material	Thermal cycles (to temp. (°C) specified) completed before failure 260	300	400	600	800	1000	Thermal-shock parameters N_f	$k\sigma_f/E\alpha$ (cal/cm⁻¹s⁻¹)	$(K_{Ic}/\sigma_f)^2$ (μm)
TiC	10	10	10(5)*	10	10	4	54	20.6	136
WC	10	10	10	10(1)	10		50	54.7	52

*Parentheses indicate cycles required for initiation of surface cracks. The specimens were 6.5 by 5.6 by 38 mm.

for TiC and WC for 3 test piece sizes. (3) Because N/V increases with decreasing specimen size, σ_a may also increase, as shown by the σ_a results in Table III. As a result, it is also plausible that the σ_a behavior can change qualitatively from catastrophic to gradual. For example, the "continuous" fracture mode observed in much smaller WC specimens* (size A in Table III) after thermal shock[3] has changed to a "discontinuous" behavior in sufficiently large specimens.

(5) Comparisons of Thermal Stress Resistance

The thermal-shock stress resistance of these two carbides may be compared using one of several methods: (1) traditional evaluation of thermal-shock stress resistance, (2) determination of T_c at which thermal cracks initiate, and (3) evaluation of the percentage of strength retained after shocking through ΔT_c. The first method has been described in detail in Ref. 1. Results for TiC and WC are given in Table IV. Apparently, these results can be correlated with the parameters $k\sigma_f/E\alpha$ and $(K_{Ic}/\sigma_f)^2$ for crack initiation and crack propagation, respectively. Based on crack initiation, WC is better because of its higher $k\sigma_f/E\alpha$ value, but based on final fracture, TiC is superior because of its higher $(K_{Ic}/\sigma_f)^2$ value. With method (2), where resistance to crack initiation is compared, the experimental results (see Figs. 2 and 3) show that WC is preferable to TiC because of its higher T_c values (i.e. $T_c \simeq 400$°C as opposed to $T_c \simeq 250$°C in TiC). This result is not unexpected since WC has much greater σ_f/E and k values. However, when method (3) is used as a criterion for comparison, TiC is preferable to WC for the quenching temperature range of 400° to 800°C. (See results in Figs. 2 and 3.) The two surface finishes do not appear to affect the thermal stress resistance for both TiC and WC in the present considerations.

IV. Conclusions

The $(K_{Ic}/\sigma_f)^2$ parameter for both titanium and tungsten carbides has been determined and found to be independent of both loading rate (8.33×10^{-6} to 2.5 m/s) and temperature (<800°C). The fracture-strength behavior of both carbides in the thermal-shock range of 25° to 800°C may be explained in terms of Hasselman's model. Instantaneous strength loss at T_c calculated using $(K_{Ic}/\sigma_f)^2$ values at T_c agree reasonably well with experimental results.

*This continuous fracture strength behavior in WC after thermal shocking is shown in Fig. 1 of Ref. 3.

The considerable strength regained by WC samples shocked through temperatures above 800°C is attributed to its much higher $(K_{Ic}/\sigma_f)^2$ values resulting from plastic deformation in the metal phase at these temperatures. The effects of specimen size and repeated shocks on ΔT_c and σ_a of the thermally shocked carbides have also been investigated.

Although, in the evaluation of the overall performance of ceramic and carbide cutting tools, resistance to thermal-shock fracture is only one of many pertinent considerations, it represents a very practical aspect in design and material selection if tool failure is to be minimized. The present work provides some useful information and procedures which may be adopted to investigate thermal shock behavior of other ceramic and carbide cutting tools, e.g. see Ref. 14.

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