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Review of the Manufacture and Properties of Ceramic Cutting Tools

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SYNOPSIS

The last fifteen years have seen major advances in the development of ceramic tool materials achieving high cutting speeds with long tool lives. These developments require rigid machine tools with higher power motors and a change in the way the tool tips are used. The major areas of application of these new tool materials is in the aerospace industries and probably somewhat less in the cast iron field. There is now an ever increasing number of grades (and trade names) available and this paper explains the background and development of the materials.

Briefly the materials may be classified as alumina or silicon nitride based and these base compositions then give rise to families of materials with alloying additions. The manufacturing routes are discussed as these have influence on both grain size and porosity of the finished material both of which influence the behaviour of the materials in use. Finally, the wear of these tools is examined against the resulting understanding of the microstructures.

1.0 Introduction

In metal cutting a chip is formed by plastic deformation and fracture of the workpiece material. It has been shown by simple continuum mechanics [1] that deformation in the region of 200-400% is required to form a continuous chip. With cutting speeds of 100 m/min deformation rates of about 10^4 sec $^{-1}$ are often observed in this process leading to high temperatures and requiring high forces for this to occur. The stresses on the tool when machining even soft materials such as 70/30 brass may be as high as 1800 MN/m², consequently tool materials need to withstand extremely high stress at temperatures up to 1000°C for high speed machining.

Ceramics were introduced in commercial quantities during World War II because of the scarcity of tungsten, the basic raw material for cemented carbide tools. The history of ceramics dates as far back as the upper paleolithic times when ceramic tools were used in the first simple machines [2]. There was no record of the development of ceramic tools after their early use until the early part of this century.

Ceramic tool materials exhibit very high hardness and wear resistance, high resistance to plastic deformation, chemical stability, etc., They presently constitute about 4-5% of the total estimated indexable insert market for metal cutting [3] and are used in the automotive industry predominantly for high speed machining of grey cast iron for producing brake drums, brake discs and flywheels. Ceramic tools are also used for high speed machining of superalloys, hard chill cast iron and

high strength steels. In each step of the advances from carbon steel tools to HSS, to cemented carbides, to ceramics, each produced new machining capabilities and spurred machine tool manufacturers to develop new, faster and more powerful machines. This development however has necessitated much more rigidity in the machine tool structure to ensure cutting progresses smoothly with little or no accidental impact to these more brittle tools. Higher cutting speeds now available also demand higher power motors and consequently the adoption of new materials in this market has always relied on manufacturers re-equipping the factory.

1.2.0. MANUFACTURING ROUTE

That ceramic tooling has only recently been adopted by the metal cutting industry can be attributed to the failure of the early ceramic tools by brittle fracture. Also these early tools failed to find wide acceptance because they had many failures as a result of improper application, unsuitable equipment or both. The low fracture strength was directly related to the presence of porosity in the pressed composite as a result of the conventional sintering techniques used. The tools produced with these early process routes also had large grain sizes, and hence were inherently weak in tension, impact and dynamic loading. These deficiencies restricted the application of the early ceramics to the continuous cutting of soft materials and cast irons at moderate speeds and feeds in sufficiently rigid machines. Development work [2-4] resulted in higher strength, more uniform and better quality tools through an understanding of the importance of microstructure in

controlling the mechanical properties, coupled with improved processing techniques. There are three major categories of ceramic tool materials available today; pure oxide, mixed oxide and nitride ceramics. Alumina (Al_2O_3) is predominant in the pure oxide and mixed oxide ceramics, while silicon is predominant in the nitride ceramics. It is therefore useful to classify ceramic tools into alumina and silicon nitride based materials.

Alumina based materials

These include Al_2O_3, Al_2O_3 + ZrO_2, Al_2O_3 + TiC, Al_2O_3 + TiC + TiN + ZrO_2, Al_2O_3 + TiN and the recently developed Al_2O_3 reinforced with SiC whiskers (5). Pure oxide (Al_2O_3) ceramic was first considered for machining operations in Germany as early as 1905, 25 years before cemented carbides were introduced. This was a relatively high purity tool in which pure alumina was densified in the presence of grain growth inhibitors (such as MgO, TiO etc). Mixed ceramics were introduced in the 1950's as one of a range of different materials based on consolidate alumina in order to meet the stringent mechanical property requirements of metal cutting. They may be classified into metal bonded and alloy tools [6]. In metal bonded ceramics, alumina is bonded by one or more of the transition metals while in alloyed tools, various alloying components either result in secondary phases or remain in solid solution. The main alloying additions are Zirconia (ZrO_2), TiC and TiN.

Silicon Nitride Based Materials

These were developed in the late 1970's. There are two forms, the α and β': α- Si_3N_4 is harder than the β'-Si_3N_4 and both forms are hexagonal, but with slightly different lattice dimensions. The maximum theoretical density of Si_3N_4 cannot be achieved by conventional sintering techniques; instead two shaping methods, known as "reaction bonding" and "hot pressing", are used. The α-Si_3N_4 is formed during the nitriding of silicon at temperatures up to $1300^\circ C$ (reaction bonding). It has a smaller yttria (Y_2O_3) content and a higher aluminium content. The β'-Si_3N_4 is a covalent solid, which contains a negligible amount of oxygen and is a well formed particulate crystal in contrast to the whiskers sometimes observed in the α-Si_3N_4 [7].

Methods Used for Producing Ceramic Tools

The early ceramics were produced by sintering almost pure alumina which was then cut to size and shape and subsequently polished. Sintering aids are still used to achieve high-quality microstructures (ie to retain small grain size while achieving high densities). The sintering aids can be divided into three categories: those which promote grain growth, those which have no effect on grain growth, and those which retard grain growth (Table 1). Some sintering aids produce double functions during sintering [8]: those which promote grain growth (such as Ti, Nb) also promote sintering, while those which retard grain growth also retard sintering. Grain growth inhibitors must therefore be selected with care to to prevent

recrystallization and yet ensure full density. The properties

of a sintered tool also depend strongly on the time and

temperature of sintering. A high pore density and fine grain

powders (0.5 - 1 micron) are required to produce a good

sintered ceramic tool. Tools made from very fine powders

usually produce a coarser final grain size than those made from

coarser starting materials [2]. The high temperatures in

sintering progressively coarsen the grain structure, with

deleterious effect on the mechanical properties of the tool

material [9]. The grain boundaries tend to migrate towards

their centre of curvature resulting in further increases of the

large grains at the expense of the smaller ones which shrink.

This process also traps most of the residual porosity within

the grains, leaving only a small amount on the boundaries, and

making further densification extremely slow.

These early problems resulted in the development of hot

pressing, hot isostatic pressing, cold pressing, and reaction

bonding as alternative methods of manufacturing ceramic tools.

These processes achieved densification with reduced grain

growth.

The hot pressing technique was developed by Deeley and others

in 1961 to overcome the problems of solid state sintering to

full density of solids with covalent bonding.[10] Ceramics

have low self-diffusivity at temperatures below that at which

thermal decomposition is dominant. In this process sintering

aids such as magnesia (MgO) are used. The major difference

between sintering and hot pressing is the application of

pressure during the consolidation process. Hot pressing

ensures rapid densification and generally results in higher density and transverse rupture strength (475-700 MN m^{-2}) for alumina ceramics than are obtainable by conventional sintering (190-350 MN m^{-2})). Hot pressing produces specimens with fine grain sizes because the temperature and time required are lower than in conventional sintering. Structural changes during plastic deformation can result in residual stress in the hot pressed ceramic tool. In hot pressing, powder compacts undergo a succession of processes: repacking, plastic flow, grain rearrangement, stress enhanced diffusion and a final stage of stress enhanced diffusion related to a creep model of deformation. An oversized die made from mould graphite is used to allow for shrinkage during sintering. The rate of pressing should be controlled - too fast a rate will not allow trapped air to escape, and the preform will disintegrate during stripping. An organic binder is added to the powder to provide internal lubrication between the powder particles during pressing, and to make the pressed compact easier to strip from the die. The furnace atmosphere during hot pressing is necessarily reducing because of the presence of graphite in the die, unlike conventional sintering.

Hot isostatic pressing (hipping) was originally developed for the fabrication of nuclear fuel components and materials not readily produced by conventional routes. The process makes use of inert gas pressure at elevated temperatures for the solid state diffusion bonding and joining of components of various metals and ceramics. The process involves an isostatic pressurizing medium where uniform pressure can be applied over the whole surface of the compact using hot pressurised gas

channelled through an expendable impervious container. This leads to the deformation of individual particles and the promotion of interparticle bonding. Lower pressure is required to consolidate most structural materials by hipping than with conventional hot pressing. High pressure is used in hot pressing since the material is pressed in one direction and pressure is often lost due to friction with the container side-walls unlike in hipping. The advantages of hipping have been summarised elsewhere [12], the main drawback is that it can be a much more costly production process than sintering.

Reaction bonding was first used in the 1950s on a small scale for producing Si_3N_4. The process consists of two steps: diffusion of the nitrogen-containing gas through a previously pressed porous compact of silicon particles, and subsequent chemical reaction between the gas and the particles to form Si_3N_4 in situ. Components produced by reaction bonding process have a lower modulus of rupture and impact strength than similar hot pressed components, largely because higher densities achievable by hot pressing. The reaction bonding process was however used owing to the difficulty of forming components by sintering Si_3N_4 compacts at atmospheric pressure, since densification will not occur below the material's decomposition temperature.

Production of Alumina Based Ceramics

High purity alumina ceramic tools are manufactured by two different methods: a conventional cold pressing operation similar to conventional powder metallurgy, or a hot pressing

operation which is very effective for sintering alumina to high density.

The alumina powder used is produced by chemical or thermal decomposition of an aluminium salt to produce pure alumina, which is then milled to a fine powder. The milled slurry is dried, and mixed with a waxy, temporary binder (e.g. resin, glue) which also serves as a lubricant during the pressing operation and makes it easier to strip the pressed compact from the die. Very fine alumina powders with an average grain size of 0.5 μm and a purity of 99.9% must be used in the initial mixture to ensure the production of fine grained components. Before hot pressing, the compacted powder is presintered at temperatures of 1500°C to 1700°C for one hour, followed by the cooling of the sintered body to room temperature. Prior to hot pressing, the compacted powder is presintered at temperatures of 1300°C-1520°C for about 30 minutes in a vacuum of 1.3Pa. Hot pressing itself is carried out at a pressure of about 20 MPa and temperatures ranging from 1500 to 1700°C for one hour, after which the sintered body is cooled to room temperature. If hipping is used rather than hot pressing, the compact is placed in an argon atmosphere at a temperature of about 1450°C and a pressure of 160 MPa for about 1 hour. Particles of sintered alumina produced by hot pressing tend to be arranged preferentially because the material is pressed in one direction, and this results in different mechanical and physical properties in planes parallel and perpendicular to the pressing direction [2[]

Mixed alumina ceramic (e.g. Al_2O_3 + TiC) tools are produced by mixing fine grained (0.5μm-1μm) and pure alumina with 20-30% of TiC powder and pressing at temperatures in the range 1500°C - 1800°C and pressure of between 10 and 40 MPa. The carbon content in the TiC should be between 12.5 to 20.05%, with a free carbon content of less than 1%. The powder size of both the alumina and TiC must be small in order to minimise particle growth during sintering [13].

The average grain size of the TiC should be less than 3 μm, with a size distribution range of 0.2-0.5 μm in order to avoid particle growth during sintering.

Production of Silicon Nitride Based Ceramics

This type of ceramic can be produced only by hot pressing or reaction bonding. The Si_3N_4 - based ceramics were initially produced by hot pressing about 4 - 12% yttrium (Y_2O_3) and 96 - 88% Si_3N_4 powders at a pressure between 7.6 and 17.8 MPa and at a temperature in the range 1650 - 1775°C, until a density of at least 3.25 x 10³ kg mY-³ is obtained [14]. The hot pressed compact will consist mostly of β'- Si_3N_4. Without the yttria addition it would have a low bend strength at high temperatures; the yttria also produces a more compact sinter and a uniform structure. This and other additives (such as Cr_2O_3, TiO, or MgO) accelerate or aid densification without significantly impairing the high-temperature creep resistance of the final ceramic material. Hot pressing of Si_3N_4 results in a reduction of porosity to less than 0.1%, which leads to an increase in strength of the compact. (A detailed description

of the hot pressing equipment is given elsewhere [15]).

It has been recommended [15] that α-Si_3N_4 powder be used as a starting material in order to achieve complete densification of the end-product.

The starting material for reaction bonding is silicon which then forms a mixture of α- and β'-Si_3N_4 (Ref. 7) The conversion of silicon to Si_3N_4 is accompanied by a volume increase of approximately 22%, but this increase does not change the overall dimension of the compact as it is accommodated in the pore space of the original compact. The microstructure of the compact, therefore undergoes a considerable change as nitriding proceeds with attendant decrease in permeability. The green density of a compact needs to be reduced as its size increases in order to maintain sufficient permeability to allow the reaction to proceed at the centre of the compact. If the density of the compact is not reduced, then a central core of unreacted silicon may occur. A recent development in the manufacture of Si_3N_4 based ceramic tools is to coat the inserts with at least one hard, adherent coating (about 1-10 μm thick) of refractory metal nitride to improve their mechanical and chemical properties. These metal nitride coatings, applied by chemical or physical vapour deposition include the nitrides of Ti,V,Cr,Zr,Nb,Mo,Hf,Ta, and W. Carbonitrides of these materials can also be used as coating materials [16].

Some problems with hot pressed or reaction bonded Si_3N_4 led many workers to carry out further examination of the structure of Si_3N_4. The stronger materials could only be obtained by hot

pressing, which is very expensive, time consuming and limits the end product to fairly simple shapes. The reaction bonded materials were easy to make but are porous and not strong enough for many applications. Sialon ceramics are the result of discovery made independently by Oyama and Kamigaito [17] and Jack and Wilson [18] that oxygen (O^{2-}) may be substituted by nitrogen (N^{3-}) in the $\beta' - Si_3N_4$ crystal provided that aluminium (Al^{3+}) is simultaneously substituted for silicon (Si^{4+}) to maintain charge neutrality. Sialon ceramics have the same crystal structure and similar physical properties to the $\beta' - Si_3N_4$ but better chemical properties because of this chemical substitution [19]. Si_3N_4, AlN, Al_2O_3 and Y_2O_3 powders are the starting materials for producing sialon tools [20]. They are milled together, dried, pressed to shape and sintered at a temperature of about $1800^\circ C$ before being allowed to cool gradually. Y_2O_3 reacts with Si_3N_4 to form a silicate which is liquid at the sintering temperature. The liquid solidifies after sintering to a glassy phase bonding together the fine grained (about 1 μm) hexagonal $\beta' - Si_3N_4$ crystals.

Properties of Ceramic Cutting Tools

A tool material for accurate and efficient machining must be strong and have high wear resistance. It should also, amongst other requirements [21], be able to resist brittle and plastic failure. Shaw [22] considers that the best tool material may not necessarily be the one which gives the longest life or is the cheapest, but rather that which performs a given task to the required accuracy and efficiency at minimum cost. Ceramic cutting tool materials exhibit the following properties: high

compressive strength, high resistance to plastic deformation, high hardness and wear resistance, and chemical stability. These properties enable them to be used for high-speed machining, in which high temperatures are generated. Ceramic tools exhibit a compressive strength which varies little with temperature, unlike cemented carbides which show a rapid drop in the compressive strength at elevated temperatures [23] (Figure 1). Ceramic tools must have negative rake angles, or rounded or chamfered edges to compensate for their low tensile and shear strengths and to take advantage of their high compressive strength and wear resistance. Negative geometries are recommended because positively raked inserts cannot withstand the mechanical and thermal shock of entry and exit from the workpiece. A negative geometry places the ceramic tool tip under compressive loading and suppresses tensile crack formation. It has been reported recently that these negative geometries can lead to favourable residual compressive stress and long fatigue lives of components machined by ceramic materials [24].

Ceramic tools have a higher hardness than cemented carbides and are much harder than tool steels at both room and elevated temperatures (Table 2). Their hardness explains why they resist abrasive wear more than carbides and tool steels, and can machine materials like castings with a long tool life provided fracture can be suppressed.

Additives/Alloying Elements in Ceramics and their effect
on the Properties of the Tool Materials

Alumina Based Materials

The bonding in alumina is ionic rather than metallic, and
consequently, is an electrical insulator with poor thermal
conductivity. The low toughness and tensile strength of
alumina ceramic tools make them less able to withstand rapid
fluctuation of temperature and stress during cutting. The
addition of additives such as chromium, titanium, or nickel
oxides or refractory metal oxides to pure alumina leads to a
significant improvement in mechanical properties [3].

Conventional alumina ceramics are usually highly susceptible
to fracture when used for machining superalloys. This tendency
is reduced by adding zirconia which helps to retard crack
propagation by transforming from a metastable state to a stable
state when a crack is initiated, thereby increasing the
toughness of the cutting tool. During cooling the zirconia
would normally undergo phase transformation. However, the
alumina causes constraint on the zirconia particles which
inhibits its transformation. This introduces compressive
forces into the structure (25). When a crack forms in a tool
in use, the metastable tetragonal zirconia transforms to the
stable monoclinic form, with an associated volume change
causing compressive stress at the crack tip and preventing
propogation [26]. These stresses effectively increase the
fracture toughness of the material by about 20-25%. The
microstructure and crystal structure of pure oxide (Al_2O_3 + <

1% ZrO_2) ceramic tool are shown in Fig. 2a and b respectively. The zirconia phase can be seen as the dark triangular phase at the grain boundaries (Figure 2B). The addition of zirconia also enables the tool to withstand high cutting temperatures helping to prevent plastic deformation or oxidation wear. The improved fracture toughness makes a pure alumina ceramic tool suitable for some interrupted cuts and other difficult machining applications it could not perform without the zirconia addition. Rapid temperature changes at the start or finish of a cut or brought about by use of coolants, can cause fracture by inhomogenous thermal expansion at the cutting edge, (thermal shock).

The thermal shock resistance of ceramic tools can be improved by introducing a metallic phase; TiN and TiC are added to provide adequate edge strength and high resistance to thermal shock. It is shown in Figure 3 that the thermal conductivity of TiN is about twice that of TiC at temperatures of 1000° - $1200^\circ C$, typical of ceramic chip/tool interfaces. The high thermal shock resistance of mixed ceramics enable them to be used for effective machining with or without coolants. The microstructure and crystal structure of a mixed oxide (Al_2O_3 + 30wt% TiC) ceramic tool are shown in Fig. 4a and b, respectively. The addition of TiC also results in a significant increase in the hardness of mixed oxide ceramics. The addition of SiC whiskers in the recently developed Al_2O_3 + SiC ceramic results in higher strength and an improvement in the fracture toughness of the brittle alumina matrix (5). The properties of the pure alumina (Al_2O_3 + ZrO_2), mixed oxide (Al_2O_3 + TiC) and the nitride (Sialon) ceramics are given in Table 3.

Silicon Nitride Based Materials

These tool materials have many good characteristics at high temperatures (1200° - 1400°C) such as good oxidation resistance, good mechanical strength, chemical inertness and high hardness in comparison to alumina based ceramics. The high thermal shock resistance of Si_3N_4 based ceramics is a result of their good thermal conductivity and low coefficient of thermal expansion (Table 3). These two factors reduce the stress set up between the hotter and cooler parts of the insert. The Si_3N_4 based ceramics have very good edge strength because of this. However, the chemical stability and wear resistance of Si_3N_4 based ceramic tools are somewhat lower than alumina based ceramics. Of the ceramics available Si_3N_4 is one of the toughest. The high fracture toughness make it less prone to catastrophic failure and makes it possible to machine at higher feed rates that can be done with alumina based ceramics. Reaction bonded Si_3N_4 has lower strength because of the higher pore space in the compact. Hot pressed Si_3N_4 components are theoretically fully dense, this improves their properties considerably. However, the pressing operation at high temperature limits them to fairly simple shapes. The advantages of sialon materials over conventional Si_3N_4 are improved resistance to oxidation, creep, and abrasion [19,27] and a pressureless sintering technique. The microstructure and crystal structure of a silicon nitride (sialon) based ceramic tool are shown in Figures 5a and 5b. The crystal structure consists of the $\beta'Si_3N_4$ cemented by a glassy phase. The interlocking nature of elongated β' Si_3N_4/Sialon grains contributes to the toughness of nitride ceramics. However the

fracture toughness of Si_3N_4/Sialon materials does not approach that of cemented carbides.

"Whisker Reinforced Ceramic/Ceramic Composites (30 wt % SiC + 70 wt % Al_2O_3

Ceramic materials are inherently brittle and thus difficult to use in manufacture. Allied to the inherent brittleness is a tendency towards porosity and thus any particular piece of ceramic have a greater likelihood of a flaw and hence properties are often given with a Wiebull modulus a higher figure for this modulus indicates a more predictable material a low figure of about 10 is more commonly seen with ceramics.

Whisker reinforcement is the product of recent development in ceramic tools in an attempt to improve the toughness, particularly of alumina (Al_2O_3) based ceramics by mechanical rather than chemical means. Increased toughness has been achieved through the reinforcement of the brittle alumina matrix with extremely strong, stiff silicon carbide (SiC) single crystals commonly called "whiskers" (or fibres). These "whiskers" are grown under carefully controlled conditions and due to their high purity and lack of grain boundaries approach the theoretical maximum strength obtainable (about 7000 MNm^{-2}) [28]. The rod like SiC whiskers are less than 0.5μm in diameter. The matrix consists of pure alumina of a fine grain size. The SiC whiskers when dispersed in the alumina

matrix functions in much the same way as fibres in fibreglass. The resulting materials after reinforcing the brittle matrix with SiC whiskers have a fracture toughness (K_{IC}) of $8MPam^{-\frac{1}{2}}$ well above hot pressed alumina composites (about $4 MPam^{-\frac{1}{2}}$) and silicon nitride ($6.8MPam^{-\frac{1}{2}}$ for Sialon) based ceramics. This improvement is possible because the SiC whiskers act as microcrack deflectors.

The reinforcement of alumina ceramics with SiC whiskers leads to a 40% increase in the thermal conductivity of the composite. This reduces thermal gradients during machining and improves the ability of the tool material to withstand thermal shock. The improved resistance to thermal shock enables the use of coolants when cutting and also permits the use of the ceramic composites for interrupted cutting without fear of catastrophic failure from thermal cracking.

Other important physical properties of the SiC reinforced alumina ceramics include, high hardness, strength and a two to three fold increase in the Weibull modulus in comparison to hot pressed composites and sialon ceramics [28]. The increased edge strength means that whisker ceramics can be offered as standard inserts without honed edges and with a positive rake thereby allowing their use in finishing operations on aerospace alloys. The absence of edge honing on ceramic inserts tend to eliminate the "smearing" problems common when machining with ceramic materials.

SiC reinforced alumina ceramic tools are mainly used for machining a range of difficult to cut exotic materials such as

nickel based alloys (including Inconel 718, Waspalloy, Inconel 903, Rene 41 and 95, Hastelloy X and Parallcy D2), hardened steels, chilled irons, and tool steels at high metal removal rates. The cost of "whisker" reinforced alumina ceramic insert is high when compared to previous ceramic and carbide inserts used for machining similar aerospace materials. This high cost will, however, be compensated with the overall improvement in productivity envisaged by using the whiskered ceramics.

Wear Characteristics

Ceramic tools fail mainly by wear on the flank face caused by the movement of the newly cut surface of the workpiece against the cutting tool. The rapid flank wear is often caused by the individual particles being dislodged from the matrix of the tool by localised stress concentration during the machining operation. The high temperatures generated during machining may also encourage the development of an uneven stress region in the tool which lowers the cohesive strength of the ceramic bond. The severe wear on the flank face of the cutting tool can lead to the elimination of the clearance angle, and the flank face thus becomes a heat source increasing the temperature and compressive stress at the nose, resulting in the fracture or catastrophic failure of the tool. Flank wear is also a result of the inherent brittleness of ceramic tool materials which encourages chipping/plucking of tool particles at the cutting edge (i.e. attrition wear). Chipping can occur if there are hard spots or inclusions in the workpiece.

Plucking of tool particles may also occur if the temperatures generated at the cutting edge are high enough co weaken the interparticle bonds or when sufficiently high stresses result from the cutting action. The chipped or plucked tool particles may travel down the flank face (or less likely over the rake face) causing increased flank wear [29].

Rake face wear (cratering) occurs but it does not limit the tool life of ceramic materials. Cratering is caused by chips flowing over the tool surface. The chemical stability/inertness of ceramic tools at high temperatures ensures that there is only a slight weakening of the interparticle bonds and minimal diffusion, resulting in the small amount of cratering. Alumina ceramics show less crater wear than the mixed oxide (Al_2O_3 + TiC) ceramics when used to machine steel or materials with high iron content [30], this is probably because the mixed oxide ceramics contains TiC which has relatively more affinity for iron than the Al_2O_3.

Ceramic inserts can also fail by plastic deformation, fracture and notching. Notching at the tool nose and the end of depth of cut may be caused either by chemical reaction at the periphery of the tool/chip interface where sliding conditions are dominant, or by work hardening of the workpiece material as a result of the high pressures at the tool/workpiece interface [29,31]. Notching is a very critical wear process when machining heat resistant steels, nickel and titanium alloys with ceramic tools since these work materials generate segmented chips whose edge makes a intermittent contact with the tool, and also generate segmented chips whose edge makes

intermittent contact of the tool, and also generates fluctuating stresses [32]. This condition leads to rapid intermittent seizure of the chips and tool (some several thousand times a second). The release of the chip after the momentary seizure may lead to small fragments of the tool material being pulled out.

Premature failure or fracture of ceramic tools occurs mainly during cutting at lower speeds, as a result of poor toughness and transverse rupture strength. This failure mode can also occur when cutting at high speeds, following reduction of the chip/tool contact length and the uneven stresses acting at the edge. This is disadvantageous since a relatively small area will be heated up during the machining operation leading to the weakening of the tool and resulting in its premature failure. Extensive research work has been carried out on the failure modes and wear mechanisms of sialon materials when cutting various work materials [32,33,34]. Notching and flank wear are the major failure modes when cutting various materials (e.g. nickel and titanium alloys, steel, and cast iron). The tendency to notch, mainly when cutting superalloys, can be minimised by using sialon tools with the appropriate geometry and adopting careful machining practices: An approach angle of 45° and a clearance angle sufficient to prevent the tool from rubbing the workpiece and dwelling in the cut. Wear mechanisms in which the sialon material with the atmosphere, particularly nitrogen, have been proposed to explain the occurrence of notching when machined high-nickel alloys. The diffusion of tool and workpiece materials to form spinels, which are easily mechanically removed, has been proposed to explain flank wear.

Plastic deformation of sialon tool materials can result from the high compressive stresses occurring during machining and this can lead to cracks because of the tensile stresses around the cutting edge. These cracks have a tendency to open up very quickly as cutting proceeds resulting in the catastrophic failure of the tool edge. Sialon materials however are not subject to the same catastrophic failure modes often seen with Alumina based materials and can be utilised with coolants. These materials have found application in machining nickel based superalloys and cast irons but have not had success with machining steel. This may be because iron dissolves in the glassy phase of the ceramic, lowering its glass transition temperature and thus altering the mechanical properties.

Summary

The development of ceramic cutting tool materials based on aluminium oxide and silicon nitride is rapidly expanding, with microstructural and property improvements and developments in manufacturing methods proceeding concomitantly. Research indicates that while the toughness of these materials is still lower than that of the conventional materials, successful application is being found. This can be enhanced by reinvestment in modern metal cutting facilities. The metallurgist entering this field therefore needs a broad appreciation of the problems and potentials associated with all sides of these developments.

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Table 1 :

Increased growth	No effect	Retarded growth		
Ti	Gu	F	Sr	V
Nb	Y	Cl	Ba	Mg
Mn	P	Br	La	
Cu	Fe	I	Cr	
Ge	Th	Sb	Si	
	Ce	K	Su	
	Zr	Na	Ca	

Table 2 :

Material Properties		Hardmetals P10	Ceramic Al_2O_3
	$20^{o}C$	1400	2200
Hardness (HV) -	1000	600	1500
Tensile α B Mp_a	20	800	200
Strength	1000	600	200
Coeff of μ	20	0.6	0.15
Friction -	1000	-	-
Oxidation - -	20	no	no
	1000	severe	no

Table 3 :

Grade	Pure Oxide	Mixed Oxide	Nitride
Nominal Composition (vol %)	$Al_2O_3 > 99$	$Al_2O_3 = 70$	Sialon
	$ZrO_2 < 1$	TiC = 30	
Density (g/cm³)	3.99	4.29	3.26
Hardness: VHN-1Kg (Kg/mm²)	1800	2230	1870
Hot Hardness (1000°C) VHN-18Kg load	800	900	1230
Toughness, K_{1c} (MPa · m ³/₂)	4.3	4.5	6.5
Young's Modulus, E(GPa)	390	416	304
Thermal Conductivity (cals/cm s°K) Room Temperature	0.0708	0.0517	0.1284
1000°C	0.0181	0.0236	0.0203
Thermal Expansion Coefficient (10^{-6}/°C) Room Temperature - 1000°C	8.2	8.6	3.1
Bend Strength (MPa)	700	910	750
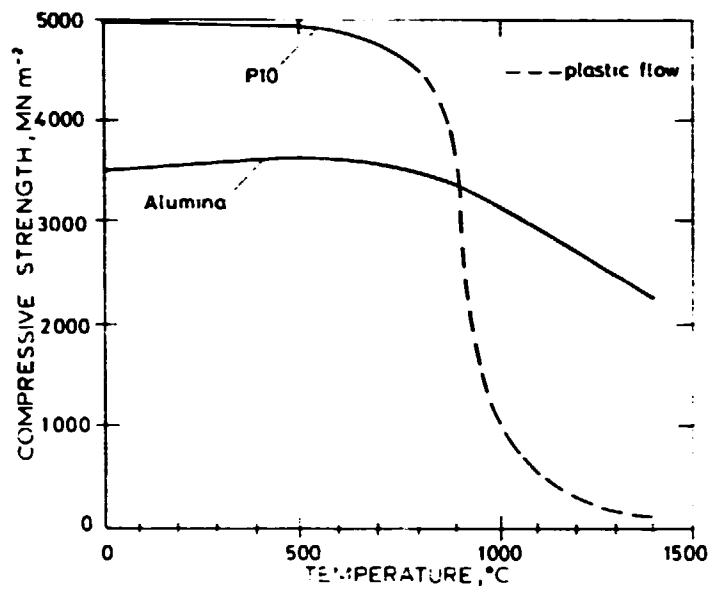
30KV X5300 10U 033 82355 KRD

ZrO₂

THERMAL CONDUCTIVITY
cal/(cm²sec)(°C/cm)

10PM   20KV   02   041   S

1μm

4PM   10KV   ·02   040   S

0·2μm