925

TOUGHENING MECHANISMS FOR CERAMICS

R.H.J. Hannink
CSIRO, Division of Materials Science and Technology,
Normanby Rd, Clayton, Victoria 3168 Australia

The application of ceramics in industry has for a long time been limited due to their lack of toughness. Over the last twenty years a great deal of effort has been made to understand the mechanisms whereby the strength of ceramics may be maintained while the fracture toughness is enhanced. Considerable progress has been made in achieving these goals.

Two types of toughening approaches have been utilised; intrinsic and extrinsic. Intrinsic mechanisms, those which increase the lattice resistance to crack propagation within a single phase material, have been of very limited success. Extrinsic mechanisms on the other hand, whereby the passage of a crack is significantly influenced by the material's microstructure, have been very effective at increasing the fracture toughness. The extrinsic approach operates predominantly through crack tip shielding mechanisms such as zone and contact shielding.

Four distinct types of crack tip shielding mechanism have emerged as those predominantly used to toughen ceramics; these can be classified as crack deflection, crack bridging, microcracking and transformation toughening. In general it is recognised that more than one system is operating at any time.

This paper describes the salient features of each of these mechanisms and outlines some of the ceramic microstructures and composites which have evolved in an effort to maximise benefits from the various toughening mechanisms.

1. INTRODUCTION

Historically the use of ceramics in structural applications has been limited due to their poor strength in tension and catastrophic fracture behaviour, ie. structural unreliability. However, other physical and chemical features make ceramics very attractive for a large number of engineering applications. These features include hardness, wear resistance, high elastic modulus, chemical inertness, electrical properties, creep resistance, retention of strength at high temperature and high melting temperatures.

To increase structural reliability both strength and toughness must be addressed. For over half a century research on structural ceramics has concentrated on increasing the strength of ceramics. This increase has in general been achieved by reducing the size and population of the inherent flaws such as manufacturing defects, surface damage and grain size [1,2]. However, it is only in the last two decades that the emphasis has been concentrated on the toughness aspects of ceramics where the most significant improvements to structural reliability are to be achieved.

During the last twenty years a number of ceramic toughening mechanisms have been recognised. Basically the mechanisms operate by interacting with the passage of a crack such that the crack tip is "shielded" from the applied stress. This is achieved, as in metals, by the presence of a "process" zone around the crack tip or by the interaction of the stress with physical features in the crack wake [3]. Within the process zone at the crack tip or within the wake, microstructural features are present which enable a reduction of the stress intensity, K, at the crack tip to be achieved. This reduction in driving force requires an increase in applied stress to maintain crack propagation.

In this paper we shall present a general summary of the toughening mechanisms which have been developed to the present day and

indicate some of the materials developed to exploit these mechanisms with the aim of enhancing their industrial performance. Due to the unusual features of the transformation toughening mechanism, larger coverage has been given to this system. For greater detail concerning the theoretical developments of the toughening mechanisms the reader is referred to a number of excellent reviews [4-7].

2. BACKGROUND

2.1. Strength-Toughness Relationship

Before describing the toughening approaches it is appropriate to examine the parameters which relate the strength and toughness. These are generally related through the flaw size and fracture energy. The strength of an ideally brittle material is inversely related to the square root of the flaw size. As mentioned flaws can be considered as processing defects, grain boundaries or surface damage resulting from grinding or mishandling. The strength or fracture stress, σ_f, is most elegantly expressed through the Griffith equation:

$$\sigma_f = 2E\gamma/\sqrt{c} \qquad (1)$$

where E is the Young's modulus, γ the fracture surface energy and c the flaw size. To relate the strength to the critical stress intensity factor, K_{Ic} or fracture toughness, the relationship is modified to:

$$\sigma_f = K_{Ic}/Y\sqrt{c} \qquad (2)$$

where Y is a geometric parameter related to the flaw size and location, ~1.2 for surface cracks.

As has generally been recognised, improvements in strength are most easily achieved through a reduction in the flaw size, c, for example by polishing the surface after grinding, reducing the grain size or eliminating processing defects. This approach has limited practical impact on toughness as it is concerned more with prevention of crack initiation and has little influence on crack propagation once started. It is crack propagation through which a material's

toughness behaviour is exhibited. Thus the most effective method for improving the toughness is by crack tip shielding.

2.2 Toughening Concepts

With the limited toughening improvements achievable through utilisation of intrinsic mechanisms alone, ie. by increasing the inherent microstructural resistance (eg. by varying particle spacings, modifying bonding strengths or increasing matrix ductility), enhancement to toughening approaches have concentrated on the concept of crack tip shielding [3]. This mechanism relies on both intrinsic and extrinsic contributions, so that the final toughness can be expressed:

$$K_{Ic} = K_i + K_e \qquad (3)$$

where K_i is the intrinsic or matrix toughness and is generally quite low (see Table 1), and K_e is the extrinsic toughness as contributed by the shielding mechanisms.

Extrinsic toughening mechanisms in ceramics are most dominant in the wake of a crack, so that for maximum toughness to be achieved a crack must propagate to establish the full process zone and wake. This behaviour results in a crack-resistance or R-curve (eg. see ref. [8]) or more recently referred to as a toughness or T-curve by Lawn and co-workers [9]. The increasing or rising toughness, K_R, as the crack advances is depicted schematically in Fig. 1, where it can be seen that a material possesses some intrinsic toughness. When no R-curve exists, at a critical stress, the crack will extend with no additional increase in toughness and the material generally fails catastrophically. Materials containing extrinsic toughening mechanisms exhibit R-curve behaviour when the applied stress is at a value sufficient to "drive the crack" but below the fracture stress. Under these circumstances the process zone and shielding zones are developed resulting in continually rising toughness until a saturation level or critical stress intensity is reached where upon the crack propagates unstably and the material fractures.
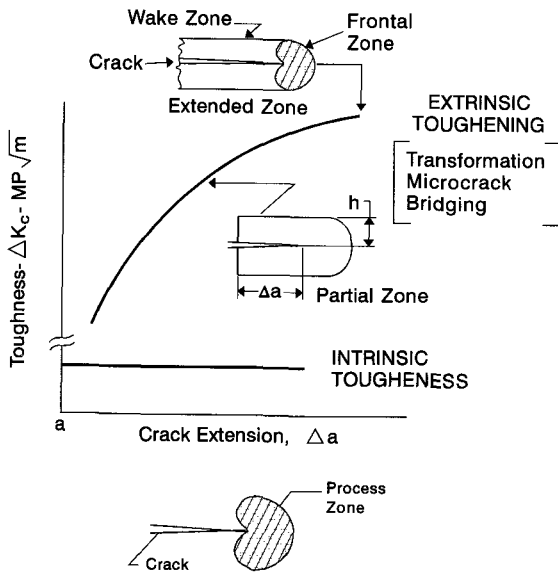
Figure 1. Schematic illustration fracture toughness versus crack extension plot for conventional ceramic material with "intrinsic" toughness and materials with microstructural design containing "extrinsic" toughening mechanisms which promote crack tip shielding.

R-curve behaviour has been well documented for metals [10] where it is recognised as a characteristic of the material, generally associated with shear lip formation resulting from increasing plastic flow at a crack tip. In general however it is anticipated R-curve behaviour will be displayed by any material which exhibits a non-linearity in its stress-strain behaviour and leads to a crack stabilising effect. Examples of this behaviour are found in viscoelastic solids or fibre reinforced composites. The energetics of a typical R-curve is shown in Fig. 2 where below the threshold toughness value, R_0, there is no crack growth even though a crack of size a_0 may be present. The energy release rate of a Griffith type crack, a

small crack in an infinite medium, is shown by the radial lines of increasing energy release rate, G, where $G = \pi\sigma^2 a/E^*$ and a is the crack length. A crack situated at a_0 will remain stationary until the applied stress has reached the minimum fracture toughness level, R_m. A further increase in stress, σ, will see the crack advance. The crack will advance stably while the rate of change of crack resistance is greater than the rate of change of strain energy release, ie. dR/da>dG/da. The crack becomes unstable when the situation is reversed, this is depicted at position a_1 for a critical Griffith stress of σ_c, when the crack resistance has risen to R_c and is referred to as the "tangency condition" for failure by fracture which defines the critical stress for the materials, σ_c. In this analyses the value R_c is less than the steady state value, R_{ss}. For a more comprehensive account of R- or T-curve characteristics the reader is referred to the recent precis by Lawn [9a].

One feature of the R-curve which is a consequence of the tangency condition, is that the strength of a ceramic is directly related to the steepness of the R-curve and not necessarily the length of crack extension. This facet of the strength-toughness behaviour is discussed for transformation toughened ceramics in section 3.2.1.

It is now recognised that any mechanism which leads to stable crack growth through crack tip shielding will impart R-curve behaviour. Ritchie [3] has described the types of mechanism which may lead to direct crack tip, wake zone or contact shielding. These mechanisms fall into three general categories plus a fourth which is a combination and are defined as:

(i)   direct crack tip: deflection and meandering,

(ii)   wake zone shielding: which for ceramics includes transformation toughening, microcrack toughening and residual stress fields,

(iii)   contact shielding: includes grain wedging, sliding and whisker, fibre, elongated grain and ductile particle bridging, and
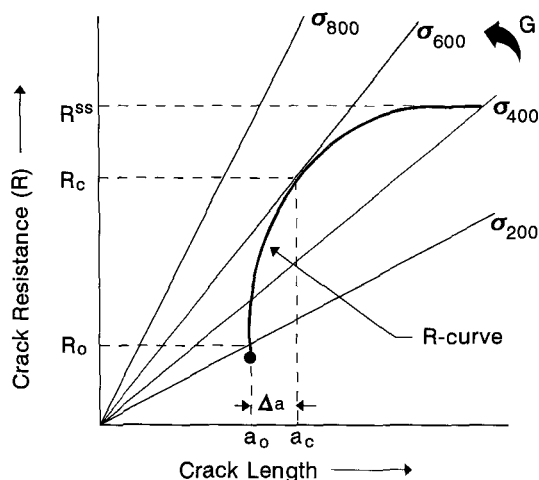
Figure 2. Energetics of a crack resistance- or R-curve developed due to crack tip shielding mechanisms. See text for details.

(iv)   combined zone and contact shielding.

An indication of the toughening contribution attributable to each mechanism is shown in Table 1.

We shall describe the salient features of each mechanism illustrating the material systems which have been developed to exploit the benefits of each.   Some systems have evolved which combine several of the shielding mechanisms and others have been specifically developed to exploit the particular mechanisms.

3. TOUGHENING MECHANISMS

While different shielding mechanisms have been identified and modelled, in practice it has been very difficult to separate the individual contributing mechanisms for the microstructural components of a system. For example it has been recognised that in optimally aged magnesia-partially stabilised zirconia materials (Mg-PSZ), as a result of the precipitate morphology, all the toughening mechanisms may be contribute to varying degrees [11].

It has been exceedingly difficult to isolate the contribution for each individual shielding mechanism to that of the total toughness of a system. Therefore, we shall present the possible contributions based on the identified mechanisms and highlight the salient features of each without attempting to indicate the fractional contribution for a particular ceramic microstructure.

3.1 Crack-tip interactions: deflection and meandering

This mechanism becomes operative when a crack tip is caused to deviate from its planar crack path by interacting with non-fracturing second phases, grain boundaries and/or residual matrix stresses. The deflection results in an increased toughening increment contributed by the reduced driving force on the deflected portion of the crack.

Faber and Evans [12] have reported a rigorous analysis for a system modelled on rod-like inclusions randomly dispersed in a matrix in which a propagating planar crack must tilt or twist. This propagation mode causes the crack to deviate from the normal to the crack plane, resulting in a reduction of the driving force or crack tip stress intensity.

While modelling of the propagating crack is highly complex the general conclusions, by Faber and Evans, have been that crack deflection and meandering can increase the toughness by upto ~4.0x, the toughening increment being dependent on the particle aspect ratio and independent of size. Figure 3 illustrates the toughening increment possible from various particle morphologies, and also shows the effect of aspect ratio variations for rod shaped particles in the range 3.5 to 12. A somewhat surprising

Table 1 - Optimal fracture toughness obtained through various crack shielding mechanisms (adapted from Ruhle & Evans [7]).

Mechanism: Primary Operating System	Optimal Toughness (MPa√m)	Typical Material	Maximum Operating Temp. (Limiting factor)
Ceramic Matrix	< 4	Al_2O_3, SiC, 5wt% MgO-ZrO_2	(Melting point)
Crack Deflection	8	β-Si_3N_4 TiO_2/Al_2O_3	1300°C (Interface effects)
Transformation: Single Phase	20	MgO-,CeO_2-,Y_2O_3-ZrO_2	~ 300°C
Composite	8	ZrO_2-Al_2O_3	(Tetragonal phase
Laminates	> 20	CeO_2-Al_2O_3/Al_2O_3	stability)
Microcrack	~ 10	Al_2O_3/ZrO_2	1100°C (Crack healing)
Fibres Matrix/fibre	≥ 30	LAS/SiC, Al_2O_3SiC SiC/SiC, C/C	~1100°C (Interface Diffusion)
Dispersed Metal	~ 25	Al_2O_3/Al, ZrO_2/Fe Al_2O_3/Ni, WC/10Co	1000°C (Oxidation, diffusion)

feature is that the mechanism saturates in its effectiveness at a volume fraction of roughly 10%.

The crack tip deflection and meandering mechanism is essentially independent of temperature and particle size, but depends on particle aspect ratio. However temperature dependence is most becomes most pronounced when the mechanism is operative in systems where thermal mismatch stresses cause compressive stresses at grain boundaries. The deflection and meandering concept has been employed in the design of a large aspect ratio β-phase in Si_3N_4 [13]. Figure 4 shows typical examples of the elongated β-SiC grains grown from a α-SiC material.

Being a crack tip mechanism, when deflection and meandering occurs it is also generally accompanied by other crack tip shielding mechanisms such as zone and contact shielding, as for example in partially stabilised zirconia [6] and WC-Co composites. More recently for β-Si_3N_4 the dominant toughness contribution has also been attributed more to contact shielding, see 3.3.

3.2 Wake-Zone Shielding

3.2.1 Transformation Toughening

This mechanism was first reported to occur in ceramics by Garvie et al, in 1975, [13] who coined the term "Ceramic Steel" for zirconia based alloys. The term was used to describe the features zirconia alloys have in common with iron based alloys such as three allotropes, martensitic
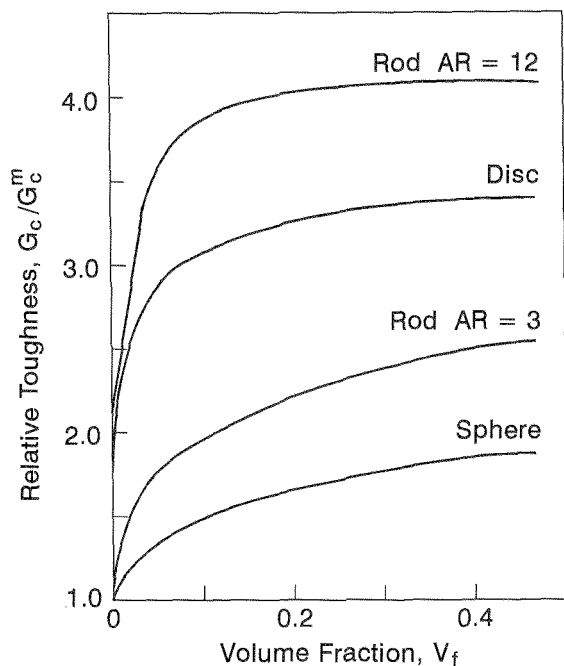
Figure 3. Toughening contributions achievable through crack deflection and meandering mechanisms operating in matrices containing, (a) inclusions of different morphologies and rods of different aspect ratios (AR).

Figure 4. Large aspect ß-SiC grains growing in a α-SiC polycrystalline matrix.

transformations and metastable phases together with thermal expansion very close to steel. Also similar to the behaviour of iron alloys, Garvie et al showed the tetragonal (t) to monoclinic (m) phase transformation which occurs in zirconia may be used to enhance the fracture energy/toughness of a material through careful alloying and microstructural control.

The transformation toughening (TT) phenomenon for ceramic systems, centred around the t→m transformation in ZrO_2, has been studied and reported in considerable detail [6,8,14,15]. Martensitic and shear transformation have also been identified in other ceramics as possible candidates for TT systems [16]. However, none have achieved the status of
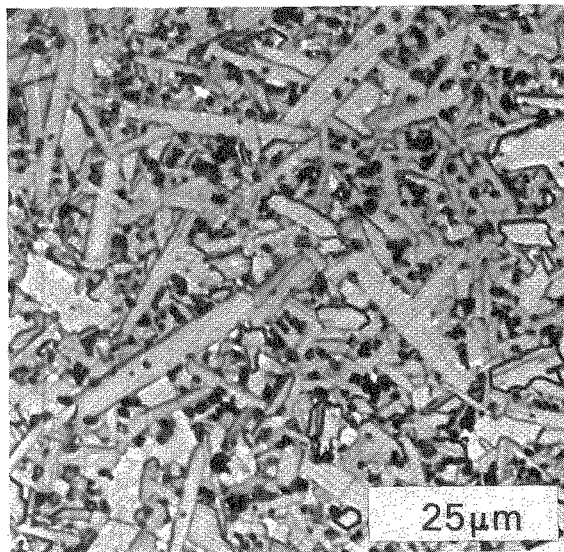
commercially viable products, therefore, we shall confine our remarks to zirconia toughened ceramics (ZTC).

Pure zirconia (ZrO_2) occurs as one of three polymorphs depending on temperature. These phases are monoclinic (m) up to ~1240°C, then tetragonal (t) to ~2380°C and beyond this cubic (c). The main features of the t→m transformation are its martensitic nature, i.e. diffusionless, a volume dilation of ~4.5% and a deviatoric shear strain of ~10%. This means that in the bulk, zirconia is unable to sustain the transformation without catastrophic failure of a fabricated body. To overcome this situation stabilisers may be added to retain the high temperature cubic phase. Driven by improved mechanical property requirements, development of partially stabilised zirconia systems, consisting of cubic and monoclinic phases, became established. However properties were still not optimised or understood and the real breakthrough came with the work of Garvie et al [13].

Claussen [17] has categorised three main groups of ZTC. These groups were subdivided into six different microstructural systems, which comprise the general class of ZTC. Commercial development of the ZTC systems has lead to the emergency of three favoured systems. These are

designated partially stabilised zirconia (PSZ), tetragonal zirconia polycrystals (TZP) and dispersed zirconia ceramics (DZC). The three systems vary in the manner by which the tetragonal phase is dispersed within a fabricated body.

The prime objective of fabrication is to produce TTC materials where the t-ZrO_2 phase is metastable so that it will transform to m-ZrO_2 at/or near room temperature with the aid of an applied stress. The temperature at which the martensitic t to m transformation starts is known as the M_s temperature. Thus chemical alloying and thermal treatment must produce an M_s for the t-phase, such that spontaneous transformation does not occur on cooling to room temperature from the fabrication temperature. A comprehensive review of the thermal treatment for zirconia based ceramics can be found elsewhere [15].

Three pronounced mechanisms have been identified which will contribute to the critical stress intensity factor, K_{Ic}, in zirconia alloys [6,15]. These are:

  a)   transformation toughening, ΔK_{cT},
  b)   transformation induced microcrack toughening, ΔK_{cM}.
  c)   crack deflection toughening, ΔK_{cD}, described in 3.1.

Other mechanisms may include crack bridging and crack trapping. The degree of contribution from each mechanism will depend on the size, morphology, dispersion and volume fraction of transforming zirconia particles. However in this section we shall only be concerned with ΔK_{cT}.

Theories developed to describe the transformation toughening increment are generally defined on the basis of energetic or mechanics arguments. These arguments assume isotropic materials with a diminishing transformation zone extending from the crack tip and surface [18,19]. Both energetic and mechanistic approaches, although incomplete, predict comparable relationships for a transforming t-particle, and in the steady state the situation is given in equation (4), so that ΔK_e in equation (1) becomes ΔK_{cT}:

$$\Delta K_{cT} = \eta \ E^* \ e^T \ V_f \ \sqrt{h} / (1-\upsilon) \qquad (4)$$

where ΔK_{cT} is the transformation toughening increment, η a factor depending on the zone shape at the crack tip and is typically 0.22 to 0.38 depending up on the hydrostatic or shear nature of the zone, E^* the effective modulus of the material, e^T the dilational strain, V_f the volume fraction of the transforming particles, h the width of the transformation zone and υ the Poisson's ratio.

McMeeking [18] has pointed out that E^* plays a very important role in determining the effectiveness of the dilational strain produced by the zirconia phase on the matrix. For example for a zirconia matrix E is ~210GPa and υ = 0.3 while for an alumina matrix E is ~380GPa and υ = 0.2; this means that the influence of the dilational strain in the higher modulus alumina matrix (a ZTC material) is only one third as effective as in PSZ or TZP materials, ie. the back stress imposed on the transformation by the high E effectively lowers the transformation efficiency, reducing h.

Typical examples of the three common forms of stress-induced transformation toughened microstructures are shown in Fig. 5. In magnesia-partially stabilised zirconia (Mg-PSZ) the toughening is due to the transformation of lenticular precipitates, Fig 5a, in yttria and ceria-zirconia polycrystals (Y-TZP and Ce-TZP) the transformation occurs in within grains of tetragonal zirconia, Fig. 5b, while in zirconia toughened alumina (ZTA) the transformation occurs in dispersed zirconia particles.

For Mg-PSZ material the optimum toughness is more readily and reliably achieved by sub-eutectoid ageing at 1100°C rather than pro- or eutectoid heat treatments at 1400°C or above [20]. Such 1100°C aged materials, at peak values, display crack growth stability during notch fracture tests, indicative of R-curve behaviour, while the 1420°C aged material has very limited R-curve behaviour and behaves more like a conventional brittle ceramic.

In transformation toughening materials both the strength and fracture toughness decrease with increasing temperature, due to the increased
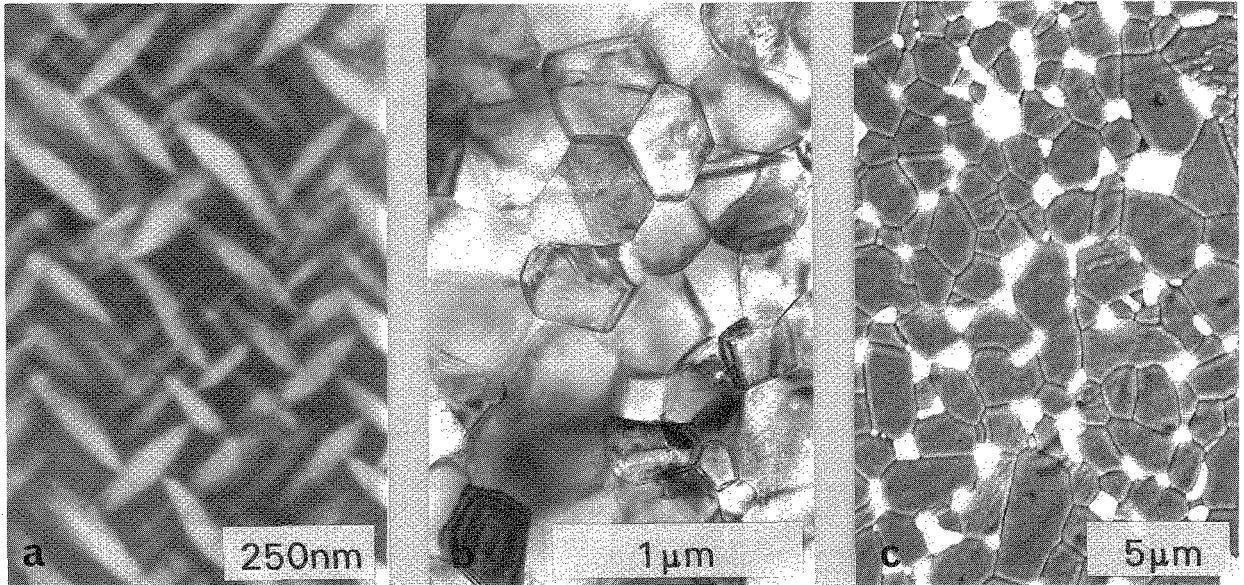
Figure 5. Microstructures of transforming phases in zirconia toughened ceramics, (a) tetragonal precipitates in MgO-partially stabilised zirconia, (b) tetragonal grains in Y_2O_3-tetragonal zirconia polycrystals and (c) intergranular zirconia inclusions in zirconia toughened alumina.

stability of the tetragonal phase, ie. a reduction in the driving force for the tetragonal to monoclinic transformation [20]. This increase in t-phase stability also implies both the volume fraction of transforming precipitates, V_f, and the transformation zone size, h, both decrease rapidly with temperature. Figure 6 illustrates the rate of K decrease for a series of different zirconia based materials.

More recently Jensen et al [21] have attempted to address the rapid decrease in high temperature toughness of Mg-PSZ by utilising the significant parameters of the various other shielding mechanism to "design" the optimal microstructure which will maximise toughening at high temperatures. These workers have modelled a CeO_2-MgO-ZrO_2 alloy in an effort to retain the transformation toughening capability at room temperature and exploit tip deflection, crack branching and sliding mechanisms at high temperatures. Preliminary efforts have concentrated on particle aspect ratio which has been increased from about 4:1 to about 9:1 through an increase in particle length (diameter)

from 0.2μm to over 0.5μm while retaining the t-form. The typical t-precipitate morphology is shown in Fig. 7.

When significant transformation occurs in zirconia based materials, PSZ and TZP, they display a limited amount of "ductility" through non-linear stress strain behaviour. For materials tested in tension this phenomenon leads to "co-operative" transformation zones [22] around indents and Luders band type features on the surface of flexural bend bars [23]. While the transformation can be readily controlled to induce transformation ductility, it has been shown that in general there is an inverse relationship between strength and toughness. Swain [24] has elegantly illustrated this fact for various zirconia based alloys. The strength versus toughness plot shown in Fig. 8 illustrates the dramatic decrease in the strength as the toughness increases above about 8 MPa √m. It is proposed that for materials with K_{lc} below about 8 MPa √m, the strength is flaw size limited while above this value is be transformation limited. In practical terms this means the transformation stress is lower than the
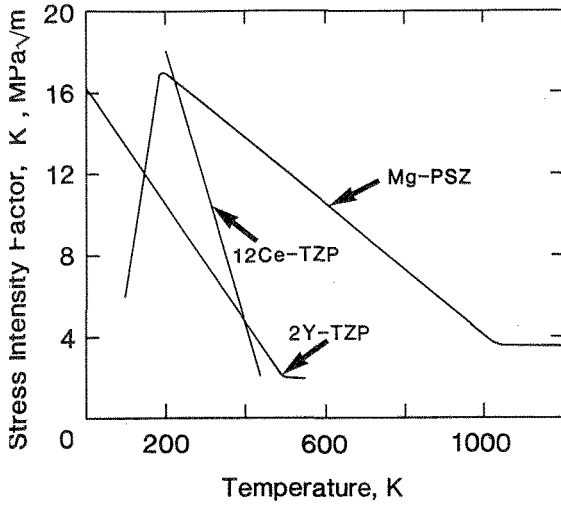
Figure 6. Temperature dependence of fracture toughness for Y_2O_3- and CeO_2-tetragonal zirconia polycrystals (Y- and CeO_2-TZP) and MgO-partially stabilised zirconia (Mg-PSZ).

fracture stress so that transformation occurs preferentially and carries on until no more t-phase is available for transformation in which event the material becomes flaw size limited again. A study determining the transformation stress has been presented elsewhere [25]

Designing microstructures for optimum toughness and R-curve behaviour has been addressed in the groups led by Marshall [26-28] and Nicholson [29-30]. These groups have both been involved in the development of transformation toughening laminates. The aim of the work is to maintain V_f, influence $\dot{E}$ to make it somewhat higher while dramatically increasing h, see equation (4). The recent work of Marshall and Ratto [28], reproduced in Fig 9(a-b), shows the dramatic increase in h when the crack tip has been constrained, within the $CeO_2/Al_2O_3-Al_2O_3$ layers of the laminate, Fig. 9a. As the applied stress is increased and the crack finally bursts through to the other side a massive zone is present and the crack is constrained within the transformed zone, Fig. 9b. This sample, of
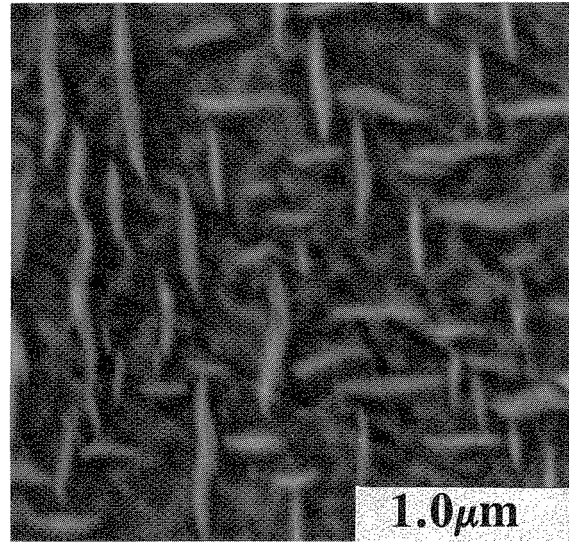

Figure 7. Large aspect ratio precipitates in a $MgO-CeO_2-ZrO_2$ alloy designed to use a number of crack tip shielding mechanisms for both high and low temperature toughening.

eleven layers, displayed a final toughness of 18MPa√m and an R-curve extending well over 1mm. Marshall and Ratto [28] have predicted K_R values well in excess of 30MPa√m for materials with sufficient layers.
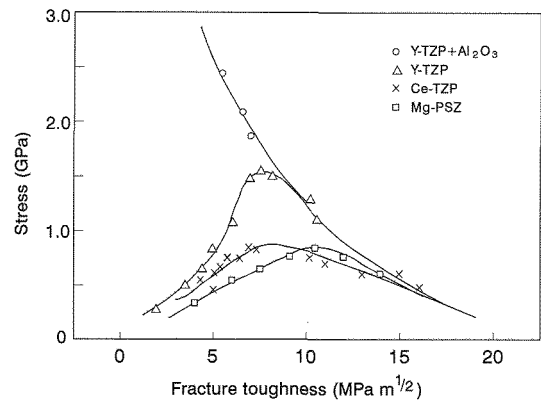

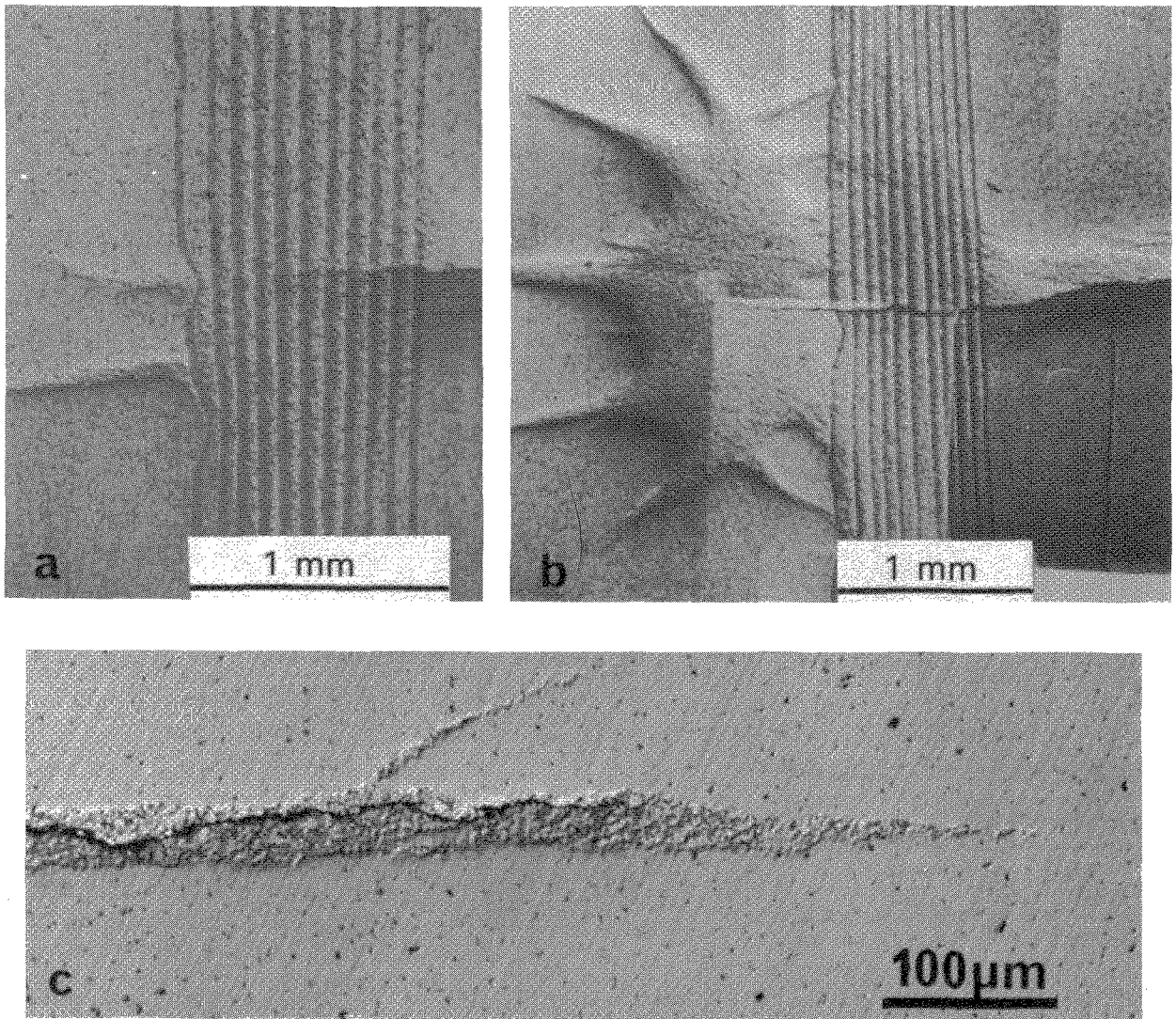Figure 8. Strength-toughness plot for a number of zirconia toughened ceramic systems, from [Swain].

Figure 9. Crack propagating in a Ce-TZP/Al$_2$O$_3$-Al$_2$O$_3$ laminate composite, (a) crack trapped in laminate layers, dashed line indicates size of shielding zone, (b) crack trapped by transformation zone (from [Marshall]). Compare (b) and (c) which shows size of transformation zone about a propagated crack in 12 mol.% CeO$_2$-TZP alloy.

Recent developments in zirconia toughened composites have also been concerned with improving the formability or mechanical shaping properties of composites. Shobu et al [31] have incorporated TiB$_2$ into Y-TZP to form, by hot isostatic pressing, very strong and moderately tough, electrically conducting zirconia/non-oxide composites, an example of ZTC. For example a 50:50 TiB$_2$:Y-TZP composite capable of being shaped by electrical discharge machining has

achieved a hardness of ~18GPa, flexural bend strength of ~1.1GPa and a fracture toughness of ~6MPa√m. A variety of carbides and nitrides have also been trialed with varying degrees of success.

In this section we have shown that transformation toughening in ceramics has been the domain of zirconia. While a general approach has been used, we have purposely not dealt with the complicating effects of dynamic fatigue [32] and corrosion, ref. eg. [14b], suffice it to say that both effects have been well studied, moderately understood and current microstructural manipulation are attempting to overcome the problems, albeit at the expense of some of the mechanical property attributes.

3.2.2 Microcrack toughening

Zone shielding by microcrack toughening was proposed nearly twenty years ago [32-34], but very few systems have been identified where the toughening increment can be positively attributed to this mechanism. Microcracks may be induced as a result of local tension, the result of thermal expansion mismatch (anisotropy) and/or transformation strains, with the resultant effect that an applied stress may initiate microfracture in the highly strained regions. If the strain build up is excessive, as for example in large grained non-cubic materials, spontaneous fracture may occur on cooling or thermal cycling [35]

There have been two approaches to determine the contribution to toughness derived from microcracking. One approach has been to treat the toughness increment in analogous fashion to transformation toughening and to take account of the change in elastic modulus [36]. The second approach, based on an energy dissipation mechanism, concerns a process zone which contains regions within which the microcracks can link.

The favoured approach has been the modification analogous to transformation toughening; in this case the incremental toughening contribution, ΔK_{mc}, can be written [37]:

$$\Delta K_{mc} = 2.67 \, \sigma_R \, V \, \sqrt{h} \qquad (5)$$

where $\sigma_R \approx (2/9) \, E^* \, \varepsilon^T/(1-\nu)$

and V is the volume fraction of spherical particles associated with a residual tensile stress, σ_R, h is the zone width, E^* the Young's modulus, ε^T the volume strain and ν the Poisson's ration.

Efforts to positively associate a toughening increment with microcracking have been very limited. For engineering ceramics, based on ZTA, Rühle and co workers [38] have performed very careful microstructural and microcrack studies. These workers found that the total increases in toughness for the ZTA system was only about 1.5 MPa √m, starting with a matrix of about 6 MPa √m.

Microcrack toughening has also been employed by Garvie at al [39] and Lutz et al [40] by dispersing relatively low volumes, <20 vol.%, of large grain zirconia in different matrices, for example zircon, alumina and yttria-tetragonal zirconia polycrystals. These workers found that large pressure zones developed on cooling, predominantly as a result of zirconia transformation, giving rise to large microcrack zones. Such materials exhibit non-linear stress-strain curves as resulting very pronounced R-curve behaviour and large wake zones, as shown in Fig. 10 from [40].

The material developed by Garvie and co-workers may find useful application as a high density, thermal shock resistant refractory material in the steel industry.

3.3 Contact Shielding

As the name implies, contact shielding is associated with the direct physical contact between separating surfaces. This may occur as sliding or wedging due to the presence of fracture surface asperities resulting from a tortuous crack path or other crack bridging features such as fibres, whiskers, metal inclusions or elongated particles. For ceramics, specifically designed ceramics matrix composites containing fibres or whiskers are the currently favoured techniques used to maximise the toughening increment. We shall briefly describe the current systems that have been developed.
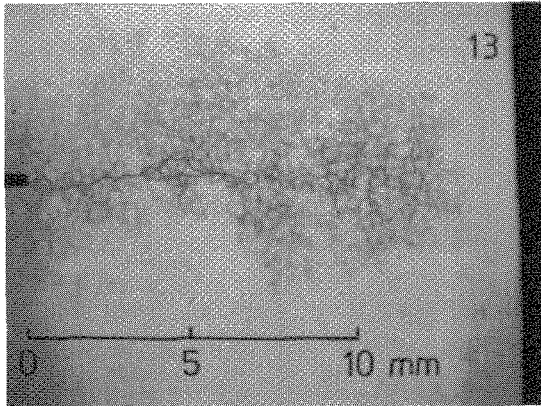
Figure 10. Extensive microcrack network around a crack in a zirconia duplex alloy, (from [40]).

3.3.1 Whisker/fibre Reinforcement

The microstructural design involving the utilisation of both whisker or fibre reinforcement is nearly the same. The difference between the reinforcing agents is predominantly the aspect ratio. For full benefit to be attained from the shielding mechanism the composite must be microstructurally tailored such that the whisker/fibre-matrix interface is sufficiently weak to permit delamination to occur when the crack encounters the interface and for debonding to continue when the crack passes such that crack bridging occurs and sliding commences as the whisker/fibre fracture surfaces separate.

It is by control of the interfacial properties by that optimal toughening can be achieved. Rühle and Evans [7] have described the physical nature of the toughening process basically as: "high fibre/whisker strength and low sliding resistance combine to maximise the frictional dissipation, by inducing sliding over the largest possible fibre surface area". As stated, this is only achievable through the microstructural control of the interface. The bridge zone toughening increment for whiskers has been determined by Becher et al [41] and has a relationship of the form:

$$\Delta K_w = \sigma_f^w \left[(V_f r/3(1-v))(E_c/E_w)(G_m/G_i) \right]^{1/2} \tag{6}$$

where σ_f^w is the whisker strength, V_f the volume fraction of whiskers, E_c and E_w are the elastic moduli of the composite and whisker respectively, and G_m and G_i are the interfacial strain of the matrix and interface respectively.

Experimental evidence for whisker reinforced composites of alumina and mullite matrices has shown that the relationship predicts the toughening component exceedingly well. Ceramic matrix systems which have been developed to exploit the whisker mechanism, in addition to alumina-SiC$_w$ and mullite-SiC$_w$, include silicon nitride-SiC$_w$, silicon nitride-silicon nitride. Fibre systems which have been developed include lithium aluminium silicate glass (LAS)-SiC, alumina-silicon carbide, SiC-SiC, SiC-C and C-C.

In order to overcome some of the processing and high temperature thermodynamic stability problems involved with multiphase composites, efforts have been made to develop single phase systems containing grains with very large aspect ratios such that they behave as whiskers or fibres. In addition to the earlier β-silicon nitride work by Lange [11] further work on these systems also has been reported where large aspect ratio grains were purposely developed [42-44]. Using this approach an improvement of two to three times was achieved in the fracture toughness over similar materials with equiaxed grains.

3.3.2 Crack Bridging

This form of contact shielding has considerable overlap with fibre and whisker reinforcement but in our discussions also includes grain and ductile-metal particle bridging. Crack-bridging may be considered to include any mechanism which will act to reduce the crack tip stress intensity factor by tensile ligament or traction forces operating in the wake of the crack rather than at the frontal process zone. The process has been appreciated for over a decade particularly in large grain single-phase alumina [45-49] and silicon nitride [42-44], multi-phase ceramics [50] and overaged-PSZ (shown in Fig 11), and will include also refractories and cementations composites.

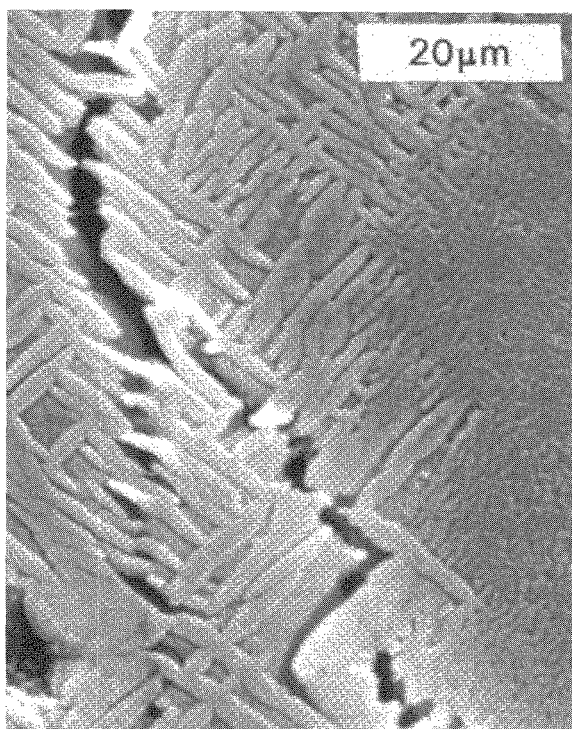It has been possible to demonstrate [51] that

Figure 11. Crack bridging in overaged Mg-PSZ.

crack bridging may develop when residual stresses of a compressive nature, present due to the thermal expansion anisotropy, cause the crack to circumvent large grain clusters. When this occurs the bridging process may contribute to the steady state toughness through the relationship [52]:

$$\Delta K_{cb} = 2.5 \ V \ E \ \Delta\alpha \ \Delta T \ \sqrt{R} \tag{7}$$

where V is the volume fraction of bridging particles of radius R, $\Delta\alpha$ is the thermal expansion mismatch and ΔT the temperature difference between the fracture temperature and temperature for the onset of creep.

The concept of toughening through inclusion of ductile phases has been exploited for over seventy years with the development of WC-Co alloys. However, recently the process and theoretical investigation has been extend to include cermet composites, low volume ductile phases, particles for crack trapping [53] and crack

bridging. The theory has also been extended to incorporate process zones [54-57]. Thus the metal incorporated ceramic systems may be defined by three types of microstructure:

i) isolated ductile particles, eg. ductile fibres [58], plates [59] or particles [60],

ii) interpenetrating networks, eg. Al in Al_2O_3 as in Lanxide materials [61] and B_4C-Al cermets [62]

iii) continuous ductile phase composites as in cemented carbides [63].

4. FUTURE DEVELOPMENTS AND CONCLUSIONS

The recognition of different crack tip shielding processes has provided a means by which the toughening systems for ceramics may be analysed or modelled. This has also provided the means by which microstructural design of ceramic systems may be investigated for operation in various temperature regimes, eg [21], and different stress states, such as fatigue [3]. The demand for materials to operate over extended temperatures ragemes, with high strength-toughness performance will ensure that ceramics, with their capacity for high strength, modulus and improving toughness, will continue to be subject of intense research effort.

To assist in the selection of engineering materials for various applications, Ashby [64] has developed maps or diagrammatic representations showing the relationship between a wide range of physical material properties. As an indication of where engineering ceramics are approximately placed with respect to other engineering materials, Ashby's diagram has been adapted to show all the engineering groups in relation to their strength-toughness properties. It can be seen from Fig. 12 that the recent developments have extended the toughness of ceramics well into the engineering metals range with an overlap in the properties coming through the cermet family. An interesting feature of the plot shows how the flaw tolerance of some strong ceramics is now about 0.1mm (100μm) as typified by optimally heat
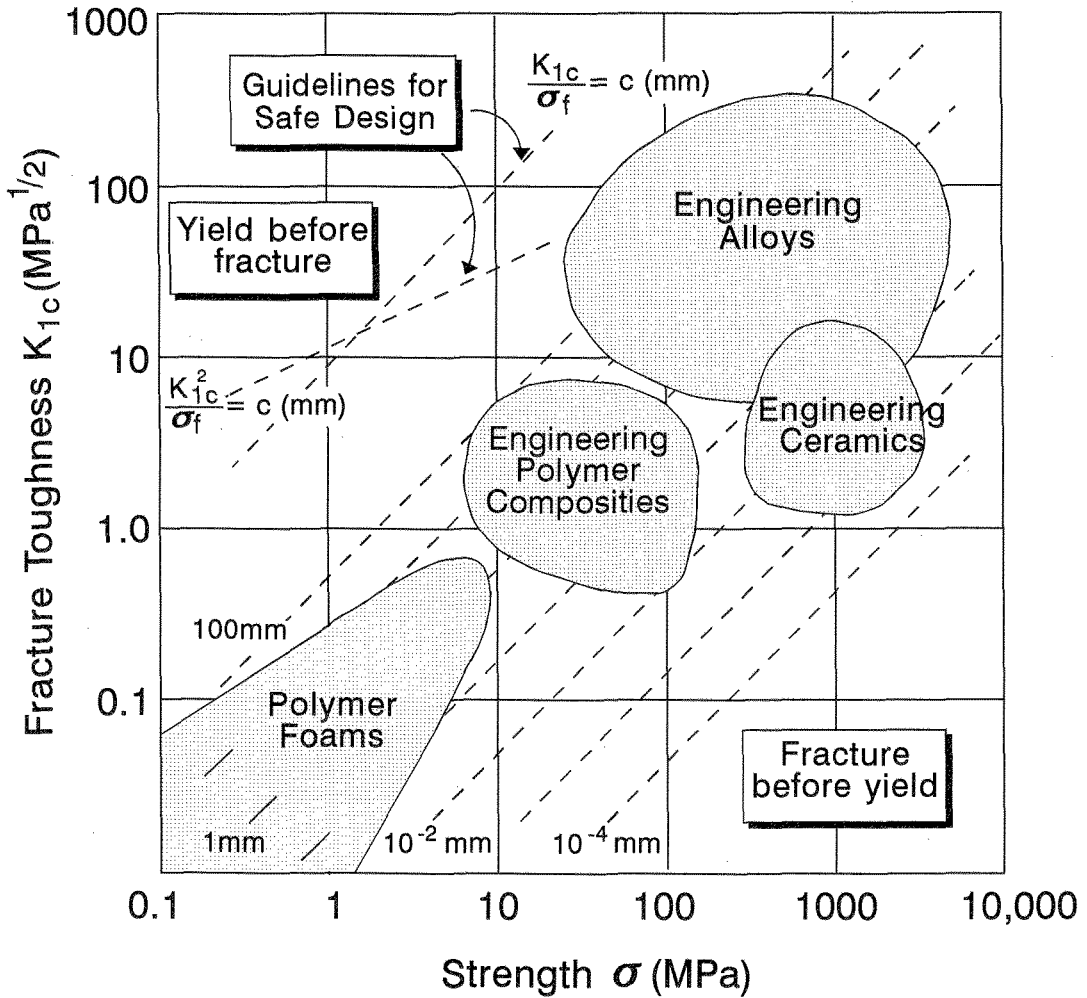
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Figure 12. Schematic illustration of the strength-toughness relationship for various classes of engineering materials. Ceramics and cermets are now beginning to approach the toughness of some metallic systems while retaining their other physical property benefits. (Strength in flexure for ceramics, tension all other systems). (Diagram adapted from Ashby [64]).

treated magnesia-partially stabilised zirconia alloys [25].

The most significant toughness improvements will come through the design of composite ceramic systems incorporating a number of toughening strategies. The most recent example

of this is represented by the work of Marshall at al [26-28] on composite laminates with toughness predicted to exceed 30 MPa √m. Broadening of the approach would see the design of ceramics following a similar path as that for metals, where the toughening mechanisms could be expected to

operate over several orders of magnitude of microstructural scale from nanometres to millimetres, depending upon the applications.

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Richard H.J. Hannink is Manager of the Ceramics and Refractories program at the CSIRO Division of Materials Science and Technology, Melbourne, Australia. He received a Ph.D. degree in Physics from Cambridge University, England in 1973, for studies involving aspects of deformation in cubic transition metal carbides. In 1973 he joined the Commonwealth Scientific and Industrial Organisation (CSIRO) in Australia where he become co-discoverer of transformation toughening in zirconia based ceramics. In addition to work in zirconia based materials, his current research interests include projects aimed at improving the strength, toughness and wear resistance of oxide and non-oxide ceramics, cermets and composites, and the development of wear resistant and high temperature materials, through the correlation of microstructures with mechanical properties and the *in-situ* performance of such materials. Other topics of work include the effects of liquid phase sintering aids and interfacial interaction mechanisms in oxides, non-oxides, cermets and composites. Hannink has authored or co-authored one book, a number of patents and over 150 papers, conference contributions and technical reports. He is a member of the Australasian and American Ceramic Societies, Materials Research Society (U.S.A.), Academy of Ceramics (Italy), an Honorary member of the Materials Research Society-Japan and a member of the editorial board of *Materials Forum*.