SILICON CARBIDE FIBRE-REINFORCED GLASS-CERAMIC MATRIX COMPOSITES : A HIGH TEMPERATURE MATERIAL FOR HIGH PERFORMANCE APPLICATION

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ABSTRACT

This paper presents the latest results concerning the development of a SiC-Nicalon/lithium aluminosilicate glass-ceramic matrix composite within the framework of a joint research-programme conducted by ONERA and SAINT-GOBAIN RECHERCHE. After a short justification for this development, it is shown how to obtain a non brittle failure through a precise control of both the composition of glass-ceramic matrix and the processing route. The mechanical properties of U.D. and 2D composites are discussed on the basis of a linear rupture-mechanic approach.

1. INTRODUCTION

The constant need for increasing the paying load in aeronautical civil aircrafts as well as improving performance and efficiency in missiles and jet engines has resulted in a continuous decrease of the structure weight. This demand for new structural design has driven the development of low density high performance materials able to operate at high temperatures. The main objectives that must be paid attention as far as the development of new materials is concerned are the specific properties due to the prime importance of weight saving on the one hand and cost effectiveness related to the component manufacture and life duration on the other hand.

For high temperature components to be used either for the propulsion system such as the modern gas turbine of jet engines, or the structural thermal insulation of re-entry vehicules, Ceramic Matrix Composites (CMC) appear as very promising materials due to a unique set of combined properties :

- a high temperature capability up to 2000°C
- high specific strength and stiffness
- a specific gravity lower than 2.5 g/cm³
- a good environmental resistance resulting from the nature of the matrix
- a damage tolerant mechanical behaviour contrary to monolithic ceramics

This particular behaviour of CMC is highlighted in figure 1 which compares the rupture stress and the rupture energy for three different kinds of ceramic materials [1].

For a single crystal, the rupture stress may be high, but with a very low energy of rupture related to a failure by cleavage. In a polycrystalline material, the rupture stress is lower due to a cohesive bond within the grain boundaries lower than that of the perfect crystal lattice, whereas the
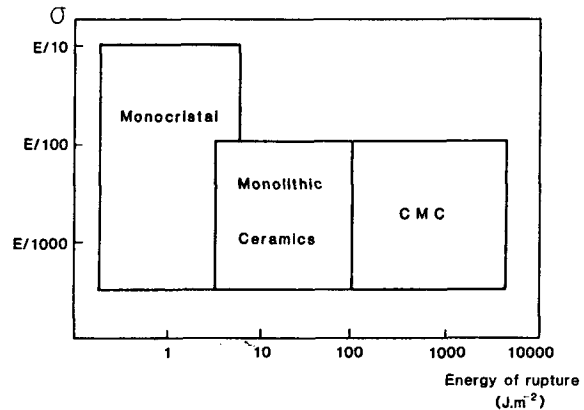

Figure 1 : Rupture stress and rupture energy for different kinds of ceramic materials.

energy of rupture is higher because of the much larger path followed by the crack along the grain boundaries. In composite material, reinforced by long fibres, the large increase in the rupture energy results mainly from the phenomena occurring at the fibre-matrix interface.

The evolution of the specific strength as a function of temperature for some ceramic matrix composites and the main materials used for airframe structure and engine components is given in figure 2. The potential temperature domain of CMC extends from 400°C to more than 2000°C with carbon-carbon composites. Among all the potential CMC candidates, glass-ceramic composites appear as promising materials because of their ability to be fabricated at low cost due to the good processability of the glass matrix at not a too high temperature.
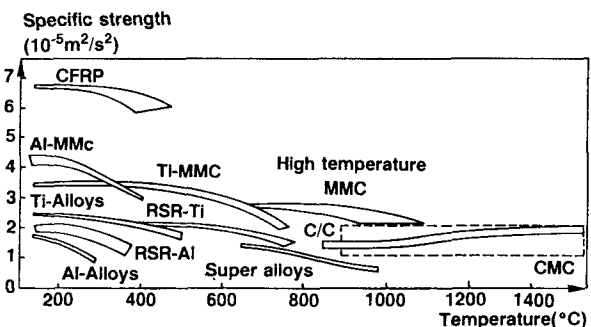

Figure 2 : Compared specific strength versus temperature for CMC and the materials most widely used in aeronautical industries.

The potential of glass-ceramic composites led ONERA and SAINT-GOBAIN RECHERCHE to initiate a research-programme aimed at the development of a composite with a lithium aluminosilicate (LAS) matrix reinforced by SiC nicalon fibres. The research programme was directed towards two main directions, a comprehensive study of the mechanical behaviour of the composite on the one hand, and optimisation of both the matrix composition and the processing route on the other hand.

2. MECHANICAL BEHAVIOUR OF CERAMIC MATRIX COMPOSITES

2.1. Brittle and dissipative failure of Unidirectional composites : the ACK model

2.1.1. Multiple and single fracture

The explanation of the rather unexpected tensile behaviour of CMC in which the association of two brittle phases results in a non brittle material was first given by Kelly [2] in the case of unidirectional composites. The observed "ductile" behaviour – it will be more accurate to speak of a dissipative rupture- was shown to result from the multicracking of the matrix throughout the whole section of the specimen, the fibres, whose elongation to rupture is higher than that of the matrix, remaining unbroken within each crack plane. However this non brittle behaviour can only occur provided that the two following requirements are fulfilled, as shown by Aveston Cooper and Kelly [3] :

- The fibre volume fraction V_f must exceed a critical value $V_f{}^c$ determined by the relationship $\varepsilon_{fu} \geq \varepsilon_{mu} (1+a)$ where ε_{fu} and ε_{mu} are the elongation to rupture of the fibres and the matrix respectively, $a = E_m V_m / E_f V_f$ being the ratio of the elastic moduli of the matrix and fibres weighted by their respective matrix and fibre volume fractions. This relationship simply expresses that the load beared by the matrix previously to its failure can be sustained by the fibres once the matrix has failed.

- The matrix crack has to grow within the composite without breaking the fibres : for this to occur, debonding of the fibres from the matrix in front of the matrix crack must take place so as to limit the stress concentrations at the crack tip in the immediate vicinity of the fibres. This implies a relatively weak fibre-matrix cohesion. Therefore, if the fibre-matrix bond is weak enough, multi-fissuration of the matrix can take place ; on the opposite, in the case of a strong fibre-matrix bond, the first matrix crack will grow straight forward, ignoring the presence of the fibres, and a brittle rupture will then be observed.

Behind the crack tip and within the crack plane, the load is therefore fully sustained by the fibres alone ; reloading of the matrix on both sides of the crack is then due to load transfer from the fibres to the matrix owing to the interfacial shear which occurs at a constant friction shear stress τ_1 over the debonded length l_c such as :

$$l_c = \frac{V_m}{V_f} \frac{\sigma_{mu}\ r}{2\ \tau_1} \qquad (1)$$

The evolution of the longitudinal stresses within the fibre and the matrix are reported on figure 3. Within the limit of the deterministic A.C.K. approach where the stress to rupture of the matrix is supposed to be unique, the length of the matrix blocks at crack saturation of the composite will therefore lie between l_c and $2\ l_c$.
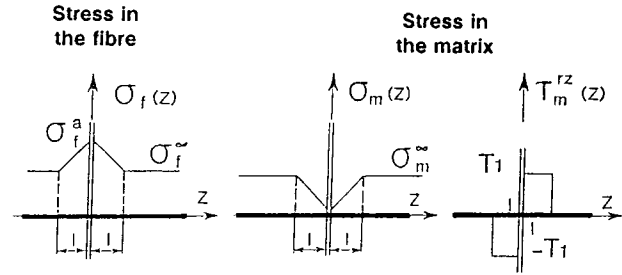
$\sigma_f^{\sim}$: Stress in the fibre away from the matrix crack

$\sigma_m^{\sim}$: Stress in the matrix away from the matrix crack

σ_f^{a} : Stress in the fibre in front of the matrix crack

Figure 3 : Stress distribution in the fibre and the matrix of a tensile loaded ceramic matrix composite in front of a matrix crack.

The tensile curve of a U.D. composite will therefore comprise the three following domains (fig.4) :

- a first fully elastic domain which ends with the first matrix crack (extending throughout the whole cross-section of the specimen),

- a second domain corresponding to the multi-fissuration of the matrix, occurring at the constant rupture stress of the matrix, thus yielding a plateau region,

- finally a third linear stage with an apparent modulus of $E_f V_f$ during which the incremental load is beared solely by the fibres.

The rupture of the composite occurs at the stress $\sigma_{fu} V_f$.
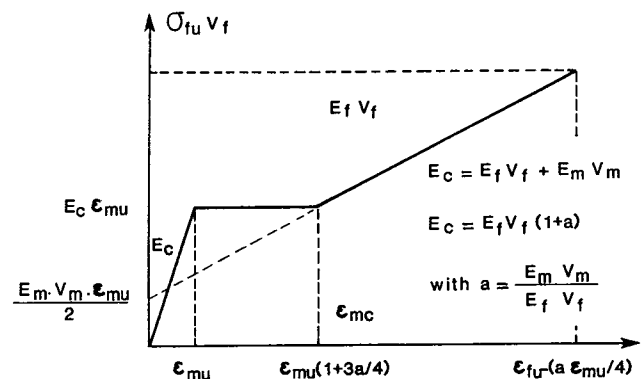

Figure 4 : Theoritical stress-strain curve according of the A.C.K. linear mechanic approach.

2.1.2. Energetic approach of the matrix fissuration

In the former simple linear mechanic approach, the elongation to rupture of the matrix within the composite has been supposed to be unique and equal to that of the monilithic matrix which has been shown not to be the case for a number of composites with a brittle matrix [4]. To account for this phenomenon of an elongation to rupture of the matrix within the composite larger than that of the matrix taken alone, ACK [3] have proposed a rupture criterion of the matrix within the composite based on an energetic approach which only considers the initial stage of the uncracked composite and the final stage where the matrix crack run throughout the whole cross-section of the specimen. The first matrix crack running throughout the whole specimen cross-section appears when the energy released by the composite due the matrix cracking ΔU becomes larger than the energy necessary to make the matrix crack $2 \gamma_m V_m + \gamma_i$, where $2 \gamma_m V_m$ is the energy necessary for the crack to grow and γ_i that required to insure the debonding of the fibres from the matrix over the length l_c. Considering in the calculus of their energy balance a simple solid friction of the fibres within their matrix sheaths along the debonded length and neglecting the energy γ_i necessary for this debonding, ACK demonstrated that the elongation to rupture of the matrix within the composite ε^*_{mu} was given by the following relationship :

$$\varepsilon^*_{mu} = \left[\frac{12 \ \tau_i \ \gamma_m \ E_f \ V_f^2}{E_c \ E^2_m \ r \ V_m} \right]^{1/3} \qquad (2)$$

which account for the reported dependance of the matrix elongation to rupture within the composite upon both the radius and volume of fibres fraction as reported by Cooper and Sillwood [4].

From this analysis ACK concluded that if the elongation to rupture of the matrix within the composite ε^*_{mu} is larger than that of the matrix taken alone ε_{mu}, then first cracking of the composite will occur at ε^*_{mu}. On the opposite, if ε^*_{mu} remains smaller than ε_{mu}, cracking of the composite then occurs at ε_{mu} ; ACK were obliged to formulate that assumption due to the fact that the relationship giving ε^*_{mu} does not allow to meet the matrix value ε_{mu} as V_m tends towards zero.

However, the ACK analysis which presents similarities with rupture mechanics does not take into account the growth stage of the matrix crack within the composite. Some attempts made in that area using the conventional homogeneisation techniques [5,6] currently used by the mechancians to derive constitutive laws for crack growth also fail in taking into account the local effects of the fibres which act as stiffeners. However, these local effects may lead to either a stable or an unstable growth of the matrix crack depending upon the conditions (volume fraction of fibres, critical energy release rate of the matrix GI_m [7]).

2.2. Modelisation of the matrix crack growth in U.D. composites

As the local approach developped by PERES [7] to confirm or infirm the validity of the previously was mentioned models too intricate to refer to an analytical calculation, a numerical simulation has been used. The composite is schematisized by a matrix plate on which are bonded the aligned fibres (fig.5). The load transfer between the fibres and the matrix occurs as in ACK model due to interfacial shear. The crack growth is studied using the energy released rate at the crack tip $G_I(a)$ according to the following relationship :

$$G_I(a) = \lim_{\Delta a \ --> \ 0} \int_0^{\Delta a} (R.\Delta u) \ dx \qquad (3)$$

where R represents the crack closing forces and Δu the displacement field resulting from a growth of the crack of an incremental step Δa.
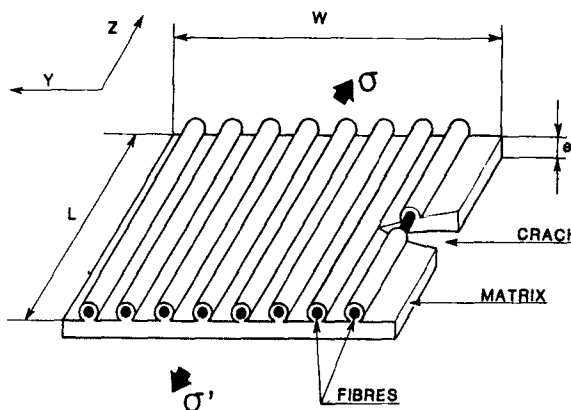

Figure 5 : Schematic drawing of the unidirectional fibre single layer composite used for the crack growth modelisation.

With a static loading at a constant stress and using an energy balance similar to that used by ACK, but considered here for each step of the crack growth, PERES derived the energy released during the crack growth ΔU as being half the value of the sum of the work done by the external forces and that resulting from the friction at the fibre-matrix interface during the glide of the fibres in their matrix sheaths. The variation of this work calculated for each step of the crack growth Δa is plotted as a function of the crack length for both a fully rigid fibre-matrix bond and a weak fibre-matrix bond on fig. (6a and 6b) respectively. The energy released at the crack tip is a pseudo-periodic function of the crack length with a series of minima, the lower each minimum, the higher the fibre-matrix cohesion. For a fully rigid bond, it appears that in front of each fibre, the crack can no longer grow whatever the matrix critical energy released rate GI_m^c.

The minimum value of the energy released at the crack tip increases during the crack growth so much the more as the interfacial shear strength decreases. The total energy released during the crack growth is well depicted by the area subtended by the pseudo-periodic curve. Therefore a weak fibre-matrix shear strength will lead to a high value of the energy released during the crack growth, and then to a non brittle dissipative failure.

However, these conclusions derived in the case of a simple single layer ideal composite will be modified in the case of a real unidirectional composite due to average effects as pointed out by ANQUEZ

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et al [8]. These average effects come from either the third dimension which exists in real unidirectional composites or from statistical scattering of the interfacial shear stress and the local fibre volume fraction which both have a strong influence on the value of the minimum of the released energy. Nevertheless, these fluctuations will mainly prevail during the first stages of the crack growth i.e for the over crossing of the very first fibres. During further propagation of the crack, the released energy will therefore lie between the lower and the upper limits of the ΔU curve of figure 6. Moreover, it is worth noting that this mean value of the energy necessary for the crack to cross the whole section of the composite is similar to that of the ACK model.
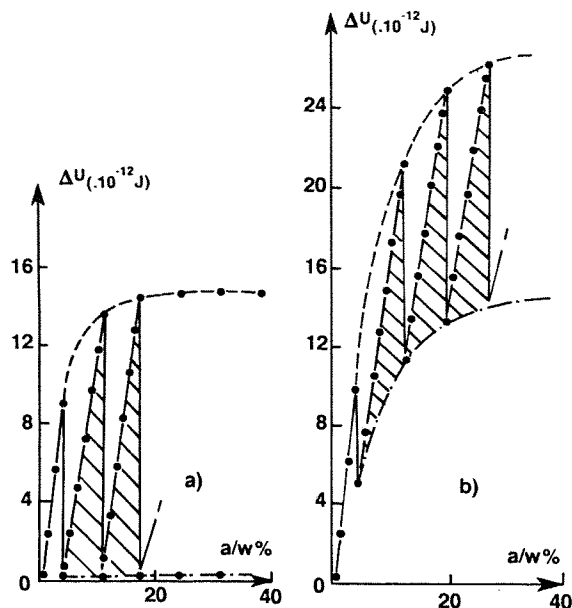

Figure 6 : Energy released at the crack tip during the crack growth for each propagation step.
a) with no fibre-matrix displacement at the interface
b) with decohesion and glide at the fibre matrix interface ($\tau = 10$ MPa)

In conclusion of this analysis of the mechanical behaviour of the multifissuration of unidirectional composites it may be recalled that :

- The ACK model allows a good interpretation of the matrix multifissuration despite its incorrectness relative to rupture mechanics.

- The more accurate PERES model which takes into account the propagation of the crack has clearly evidenced the role of stiffener played by the fibres in a way similar to what is observed in a stiffened metallic panel.

3. The SiC nicalon - LAS composite

   3.1. Chemistry, Microstructure and Processing of the SiC-LAS composite

The composite used for this study is obtained by hot pressing of prepreg tapes made from Nicalon SiC fibres and a LAS (Lithium Oxyde, Alumina, Silica) glass choosen for both its low density of 2.5 and its ability to work at high temperatures as pointed out by PREWO et al [9,10,11]. The development of this

class of composite has necessitated in a first step an optimisation of the matrix composition to meet the requirements of a sufficiently low compaction temperature compatible with the limited thermal stability of the SiC nicalon fibre [12].

   3.1.1. Optimisation of the matrix composition

One of the main interests of the LAS matrix consists in its ability to allow a precise tayloring of the crystalline microstructure, based on the quartz allotropic varieties, through the ceramed heat-treatment temperature. It is therefore possible to obtain :

- for a low ceramed heat-treatment performed in the range 700°C-900°C, a high quartz solid-solution usually known as β-eucryptite ; this phase presents a fairly low thermal expansion coefficient leading to the possibility of designing a large variety of thermal shock resistant materials mainly used for both domestic and optical applications [13] ;

- for ceramed heat-treatments performed at a higher temperature within the range 1000°C-1200°C a keatite solid-solution best known as β-spodumène whose thermal stability is better fitted for heavily loaded high temperature components : as an example, the so called LAS 4 composition (LiO_2, Al_2O_3 $4SiO_2$) exhibits a high melting point of 1430°C.

Starting from a glass-ceramic matrix with compositions located between LAS 4 (LiO_2, Al_2O_3, $4SiO_2$) and LAS 6 (LiO_2, Al_2O_3, $6SiO_2$) systems, the optimisation of the matrix composition was mainly directed towards the search for a glass composition combining the two following characteristics :

- a glass-transition temperature sufficiently low to allow processing of the composite at a temperature compatible with the thermal stability of the fibres,

- an ability of the matrix to form a diffusion barrier at the fibre-matrix interface so as to taylor the interfacial shear strength within the range necessary to obtain a non brittle failure of the composite. This was achieved though controlled additions of minor elements acting either as fluxing agents (MgO, Nb_2O_5) or as nucleation agents (ZrO_2, P_2O_5).

The glass is melted in a platinum crucible starting from a mixture of pure oxides and carbonates in a gas furnace at a temperature of about 1650°C. The glass powder, 10 µm in diameter, is subsequently obtained by grinding.

   3.1.2. Processing and microstructure of the composite

The composite is processed from prepreg tapes obtained by deeping the previously unsized fibre preforms in a slip casting containing the glass powder and the binders. Hot pressing of the composite within a graphite mould is performed under a reducing atmosphere of nitrogen or a mixture of nitrogen and hydrogen at a temperature ranging between 1300°C and 1400°C, where a viscosity of 5000 poises can be reached, thus allowing a low moulding pressure of 2 to 10 MPa to be used. These pressing conditions insure a very low porosity within the composite. The compaction stage is subsequently followed by a two step ceramed heat treatment performed at 750°C and 1100°C respectively to control the grain size by nucleation and growth.

The characteristics of the fibre-matrix interface, which are of prime importance for the achievement of a non brittle composite, have been paid particular attention. Processing of the composite has therefore been performed using both unsized SiC fibres and previously coated fibres with either a niobium oxide (Nb2O5) PVD layer or a carbon CVD layer, so as to check the best route to built the interfacial diffusion barrier. It has been further shown that the best results concerning the diffusion barrier were obtained by processing the composite with uncoated fibres. In that case a continuous niobium carbide layer was shown to exist all around the fibres (fig.7) ; this continuous layer limits the chemical reactions between the reactive species present in the matrix such as lithium or aluminium, and the fibres thus insuring the required weak fibre-matrix bond, as already reported by BRENNAN and co-workers [14,15].
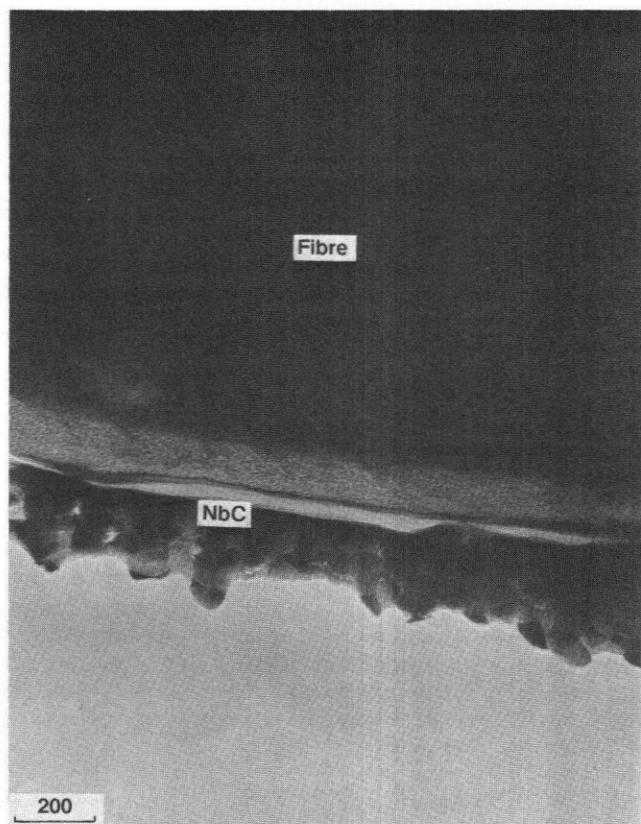

Figure 7 : Niobium carbide layer located around the SiC fibres. T.E.M. micrograph.

3.2. Mechanical behaviour of unidirectional composites

The mechanical behaviour and more precisely the brittle or dissipative failure of SiC-LAS glass-ceramic composites is strongly dependant upon the interfacial matrix shear strength as shown by ACK ; moreover from a practical user point of view, it is of prime importance to have an accurate knowledge of the onset of the matrix multifissuration and of the crack growth, because of the deleterious effects of the environment on the cracked material and the design limitations that may result for real components.

Two techniques were shown to be particularly valuable to reach the interface mechanical characteristics on the one hand and to get some insight on the damaging by fissuration on the other :

- the instrumented indentation test very well fitted to CMC, which allows a precise measurement of the interfacial friction shear stress,

- the microtensile test coupled with a micrographic observation of the specimen surface by using an optical microscope, also called microscope In-Situ tensile testing, which allows a direct observation of the matrix crack nucleation and growth.

3.2.1. Instrumented micro-indentation

In the micro-indentation test first used by MARSHALL [16] a diamond Vickers indenter is used to push on the fibre till debonding and glide of the fibre in its matrix sheath occur (fig.8)

However this method does not allow an accurate measurement of the interfacial friction shear stress mainly due to scattering of the fibre hardness from one fibre to the other.
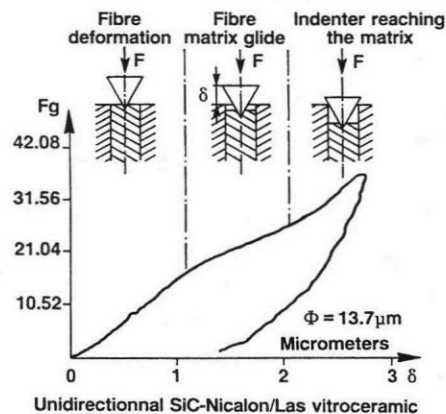

Figure 8 : Instrumented Vickers indentation test : load versus displacement curve.

By a simultaneous measurement of both the applied load and the displacement, the instrumented micro-indentation technique allows to overcome the former inaccuracy and to confirm the validity of the model used for the exploitation of the rough measurements results. The load displacement curve of a micro-indentation test figure 8 usually exhibits three stages during loading :

- the first stage which corresponds to the indentation of the fibre ends with debonding of the fibre from the matrix,

- the second stage which corresponds to the glide of the debonded length of the fibre in its matrix sheath ends when the indenter reaches the matrix,

- the third one corresponds to the indentation of the matrix.

Determination of the interfacial friction shear stress is done using only the second stage part of the load displacement curve using a simple shear lag analysis.

With a constant interfacial friction shear stress τ_1, the displacement u by glide of a fibre with a radius r_f within its matrix sheath expresses simply as a parabolic function of the applied load F as shown by PERES [7]

$$u\ (F) = k_2\ F^2 \qquad (4)$$

with $k_2 = \dfrac{\Pi^2}{4}\ E_f\ r_f^3\ \tau_1$

This technique allows an accurate discrimation between SiC-LAS composites processed by different routes so as to modify the nature of the interphase present at the fibre-matrix interface as stated previously (§ 3.1.2.). Table (1) clearly shows that this interface tayloring allows to vary the interfacial friction shear stress by a factor ranging between 2 and 3.

Composites	τ_1 (MPa)	Flexural strength	Feature of the rupture
SiC/LAS	11	1020	dissipative (pull out)
SiC+C/LAS	20	600	rough rupture surface
SiC+Nb$_2$O$_5$/LAS	25	580	rough rupture surface

Table 1 : Mechanical properties of 1D SiC/LAS composites in relation with the fibre-matrix interfacial friction shear-stress τ_1.

The same table higlights the correlation between the friction shear stress, the flexural rupture strength and the kind of rupture. It can easily be seen that a high rupture stress and a dissipative rupture can only be achieved provided that the friction shear stress remains sufficiently low. For SiC-LAS composites only materials processed with uncoated fibres exhibit good mechanical properties.

Similar conclusions can be drawn for other CMC's. Therefore, for each CMC a critical value of the interfacial friction shear stress below which a dissipative failure is observed can be derived, as for example 150 MPa for the SiC-SiC composite as compared to 10 MPa for the SiC-LAS composite.

As a conclusion instrumented microindentation appears as a very efficient test method to determine the onset of the fibre decohesion and glide thus allowing an accurate measurement of the interfacial friction shear stress. Moreover this method may in the future develop as a Non Destructive Testing mean allowing the control of the components interfacial characteristics necessary to achieve a dissipative failure.

3.2.2. Tensile behaviour of U.D. composites and microscope in-situ tensile testing

Tensile stress-strain curve

The typical stress-strain curve of a U.D. SiC-LAS composite (fig.9) exhibits three domains as the theoritical curve derived from the deterministic ACK model ; the only difference between the two curves arises from stage 2 where the multifissuration of the matrix within the SiC-LAS composite does not occur at a constant stress. This discrepancy between the two curves might arise :

- either from the statistical flaws distribution within the matrix leading to a variety of crack nucleation stresses ;
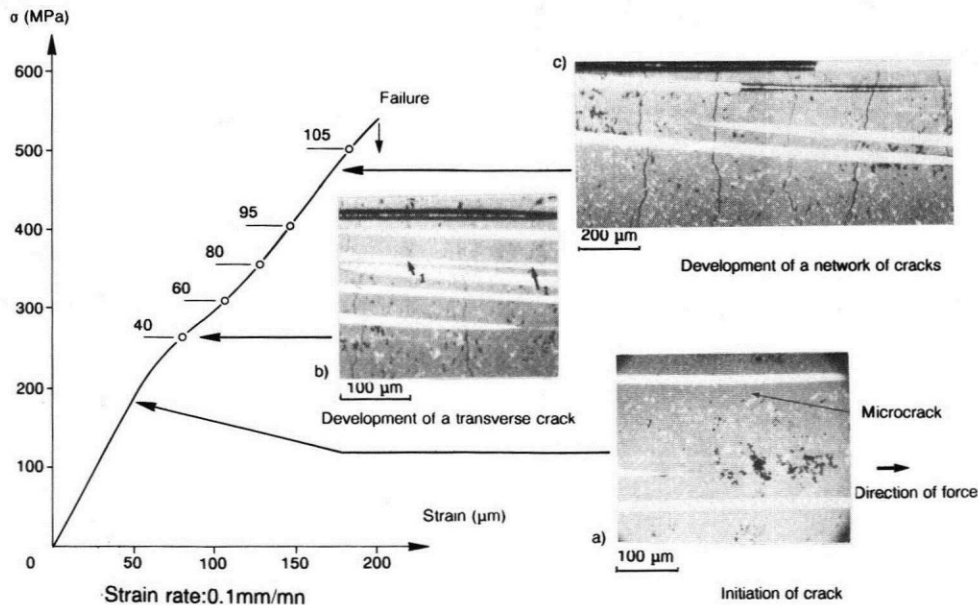

Figure 9 : Tensile stress strain curve for a unidirectional SiC-Nicalon/LAS composite (Vf 37%) exhibiting the different damaging stages and the number of matrix macrocracks.

− or, as stated by PERES, from the fact that the limiting step of the matrix macrocracking is not the nucleation stage but the propagation one.

Microscope in-situ tensile-testing

Recording of the composite damage evolution with the loading reveals the following steps :

− first the nucleation of small microcracks, most of them appearing in the immediate vicinity of the fibres at the end of the elastic domain (fig.9a mark A) ;

− secondly the development of transverse macrocrack related to a loss of stiffness of the composite (fig.9b, marks A, B) ; it is worth noting that no fibre failure was observed during this stage ;

− thirdly the multifissuration of the matrix (fig.9c).

During this stage, which ends at the saturation of the matrix cracking (mark C) a keen observation of the cracks reveals that their growth might be either stable or unstable depending upon the location of the crack relative to the surrounding fibres. A rapid growth rate is observed for cracks located away from fibres whereas a slow growth rate is always related with either the vicinity of a fibre − thus evidencing the crack closing effects arising from the stiffener role of the fibres − or from a state of lower local stresses due to the presence of close transverse cracks.

From these observations it can be concluded that the sigmoïdal shape of the tensile curve observed on real composites instead of the plateau region predicted by ACK results from the matrix crack growth and not from pre-existing flaws within the matrix.

3.3. Mechanical behaviour of 2D composites

Real components require generally to be reinforced in more that one direction. In order to get some insight on the performance of composites with a fibre architecture relevant for structural applications a satin weave woven fabrics was used to realise 2D composites by the same route as the U.D. composite.

The flexural strength of 2D composites processed using different fibre-matrix interface control are reported in table (2) as a function of temperature.

Flexural strength (MPa) / Composites	20°C	650°C	900°C
SiC/LAS	300±50	310±50	250±50
SiC + C/LAS	300±30	250±20	320±20
SiC + Nb₂O₅/LAS	260±30	280±30	200±30

Table 2 : Evolution of the flexural strength with temperature for 2D SiC/LAS composites : influence of the fibre-matrix interphase.

As in the case of U.D. composites it appears that the composites processed with uncoated fibres exhibit again the best mechanical properties despite a rather large and unexpected scattering of the rupture stresses at room temperature.

These results obtained by an accurate control of the fibre-matrix interface properties clearly evidence the important potential of glass-ceramic composites for both structural airframe and propulsion applications. The mechanical properties of different kinds of ceramic matrix composites for structural use at high temperatures are compared in table 3 and figure 10. Table 3 highlights the very good properties of the SiC-LAS composite as compared to both those of the CVI processed SiC-SiC composite, today in use

Composites / Properties	SiC/LAS	SiC/SiC	C/SiC
Geometry of 2D woven fabrics	satin weave	plain weave	plain weave
Total volume fraction of fibre	40 %	40 %	45 %
Apparent specific gravity (g/cm³)	2,5	2,5	2,1
Open porosity	2 %	10 %	10 %
Ultimate bending strength (MPa) (typical values) − 20°C	300	300	500
− 900°C	250	400	700

Table 3 : Structural and mechanical resistance of SiC-LAS composite compared to the two main CMC's in use today.

on a number of solid an liquid propellant rocket and launchers, and the equally CVI processed C–SiC composite. For an equivalent load bearing capacity, the advantages of glass–ceramic composites arise mainly from their simple processing route and the good oxidation resistance of the matrix up to 1200°C.

SPECIFIC STRENGTH ($\times 10^5$ m^2 s^{-2})

—•— C/G.C.   —+— C/SiC*
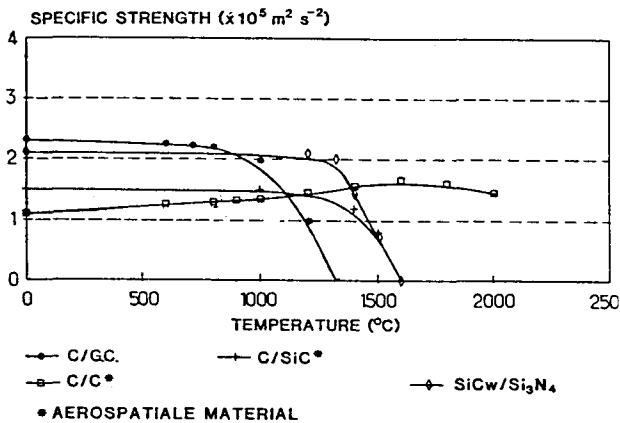—□— C/C*   —◆— SiCw/Si$_3$N$_4$
• AEROSPATIALE MATERIAL

Figure 10 : Specific strength versus temperature for main ceramic matrix composites in use today.

Conclusions :

Glass–ceramic composites reinforced by nicalon SiC fibres appear today as structural materials of interest for the replacement of metallic alloys up to 650°C with a future development for use at much higher temperatures.

The good level of mechanical properties together with the non brittle behaviour results from both the refinement of the matrix composition and the processing route that allow to control the fibre–matrix interface characteristics.

The simulation of the damage of U.D. composite by matrix cracking together with the In–Situ microtensile experiments observations clearly demonstrate that the multifissuration of the matrix is controlled by crack growth and not by pre–existing flaws.

Constitutive laws allowing to predict the evolution of the composite mechanical properties as a function of the matrix damage will then be easier to establish. This will open the way to an extended use of these composites beyond the elastic linear stage, i.e. for more severe loadings.

Finally, from a practical point of view the flexibility and short duration of the processing cycle using either hot pressing or injection moulding will permit the manufacture of complex shapes with a real cost effectiveness if compared to other processing routes such as Chemical Vapour Infiltration of fibre preforms.

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