

Structure and mechanical properties of Al-Si metal matrix composite with additions of SiC particles

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Abstract

Metal matrix composites have microstructures that depend on the base metal, on the characteristics and on the properties of the dispersed particles, on the manufacturing modalities, on the heat treatment. The mechanical properties are strongly function of the composite structure. In this work, the microstructure of a metal matrix composite of Al-Si containing 10 and 20 volume percent SiC particles has been extensively studied, analyzing some mechanical properties of this composite, essentially using traction tests. The influence of the different SiC per cent and of the different microstructures both on the elastic deformation and on the strain hardening of the composite has been analyzed.

Riassunto

I compositi a matrice metallica hanno microstrutture che dipendono dal metallo base, dalle caratteristiche e dalle proprietà delle particelle disperse, dalle modalità di fabbricazione, dal trattamento termico. Le proprietà meccaniche sono a loro volta fortemente dipendenti dalla struttura del composito. In questo lavoro è stata estesamente studiata la struttura di un composito a matrice metallica a base di Al-Si con aggiunta di particelle di SiC in percentuali del 10 e del 20% in volume, analizzando alcune proprietà meccaniche di tale composito essenzialmente attraverso delle prove di trazione. Si è analizzata l'influenza che le diverse percentuali di SiC e le diverse strutture microcristalline hanno sia sulla deformazione elastica che sull'incrudimento del composito.

INTRODUCTION

Composite materials can be applied in economically viable ways only if we are able to predict their service life with a sufficient degree of reliability, but without any excessive conservative safety margin. As these materials are non-homogeneous in their properties, such predictions are extremely difficult, and are possible at all only if we find out all that is

relevant about the materials. This paper describes the structure and mechanical properties of a metal matrix composite, or MMC, based on an eutectic Al-Si alloy and containing either 10 or 20 volume per cent of added SiC particles. These discontinuously strengthened composites are marketed under the designation Duralcan F3D.10S and F3D.20S.

EXPERIMENTAL MATERIALS AND TECHNIQUES

According to the manufacturers [1], the metal matrix of these composites has the following analysis, in per cent by weight:

TABLE 1 - Chemical composition of matrix in per cent by weight (the Al is rest)

	Si	Fe	Cu	Mn	Mg	Ni	Ti	Zn	Others
Min.	9.5	0.8	3.0	0.5	0.3	1.0	0	0	0.03
Max.	10.5	1.2	3.5	0.8	0.5	1.5	0.2	0.03	0.1

The SiC particles were injected during the pressure die casting process; in view of the high Si content of the matrix were apparently not impregnated [2, 3]. The castings were in our

case plates measuring about 12.7 x 140 x 140 mm. Although this is a discontinuously strengthened MMC, it was decided to examine both its structure and its properties in various

locations and directions, as indicated in figure 1:

- microstructure and hardness were investigated both on the plate surfaces and on the surfaces of perpendicular sections, i.e. in the directions denoted 1 and 2 in figure 1a.
- rectangular-section tensile test specimens (figure 1b) and notch toughness specimens (figure 1c) were tested in both the lengthwise and transversal direction; latter specimens had their notch located so that the fracture surface ran across the plate.

The Brinell hardness was ascertained, with impression of 2.5 mm diameter at 1875 N, both on as cast surfaces (direction 1 in figure 1) and on machined notch toughness specimens (direction 1' in figure 1c); as well as on tensile specimens (direction 2 in figure 1b), and on these specimens after they had ruptured, in that part of the gauge length where uniform plastic deformation may have hardened the material (direction 2' in figure 1). The specimens proved difficult to machine. Troubles began when the cast plates were to be sawn apart: the very hard SiC particles quickly blunted the saws. A more economical way was to cut the plates with a water jet, but the sections were not truly perpendicular, not even at a slow jet feed rate of only 17 mm/min. The specimens were finished by milling, hand grinding and polishing of the relatively soft matrix. Similar difficulties were encountered in the preparation of sections for microstructural studies and WDX phase analyses: diamond grinding paste had to be used. In tensile tests, flat specimens of 3 x 12.7 mm cross section and 35 mm gauge length

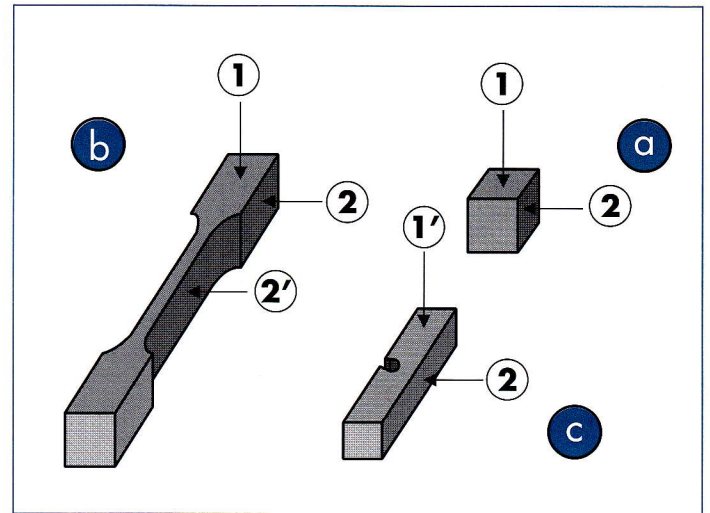


Figure 1: Sampling points for microstructure investigations (a), tensile bars (b) and notch toughness (c) specimens cut from plates of 12.7 x 140 x 140 mm. Locations and directions are marked 1 (as cast surface), 1' (approximately 1.5 mm beneath the surface), 2 (middle of section surface on a ruptured specimen).

were ruptured at room temperature at a rate of 0.2 cm/min. For the notch toughness tests, also run at room temperature, specimens of 10 x 10 x 55 mm were provided with a keyhole notch 3 mm in size. Fracture surfaces were subjected to scanning electron microscopy with EDX analyses.

RESULTS AND DISCUSSION

3.1 Hardness

The results of Brinell hardness testing are summarized in Table 2. The data in table 2a indicate that there is next no difference between the hardness levels at the surface (1 in figure 1) and in the subsurface layer (1') in both composites. The same applies to figures for the middle of the section surface (2) and for the middle of ruptured tensile or notch toughness specimens (2'). Some hardness differences are evident between the two materials, i.e. in $(\Delta H_B)_{m=c}$ data in table 2b. This suggests a non-uniform distribution of SiC particles, with greater particle concentration entailing higher hardness values. The differences between surface and core hardness levels appear to be caused by the plate manufacturing technique, which led to different particle concentration at the surface and in the interior, because these differences are very similar in both the examined materials: in the 1 and 1' directions, the surfaces of both these composites are harder by 6 or 7 BHN than the cores.

3.2 Microstructure

The microstructure was examined in the two mutually perpendicular directions, marked 1 and 2 in figure 1, in which the hardness was investigated. Polishing the sections without etching revealed a strongly non-uniform SiC distribution at and close to the surface of both composites; this is evident in figure 2, obtained on F3D.10S material. In the middles of the specimens this non-uniformity was only slight, as is shown in figure 3 on electrolytically etched F3D.20S material. Such optical micrographs, gained on normally prepared sections, did not permit any analysis of the microstructure or of the proportion of area f , covered by SiC particles, and nor did the microscopic observations. They could reveal only the uniformity of the SiC distribution or, as in figure 3, the grain size. Only meticulous hand finishing of the specimens with diamond paste allowed analyses. Figures 4 and 5 are examples of pictures gained in WDX analyses of F3D.10S and F3D.20S materials respectively, figures 4a and 5a by sec-

TABLE 2 - Brinell hardness HB: (a) at the locations and in the directions shown in Fig. 1, in both materials. In part (b) of the Table $(\Delta H_B)_{m=c}$ marks the differences between the hardness observed in location 1+1' and 2+2'. In both parts of the Table, $(\Delta H_B)_{d=c}$ denotes the differences detected between the two materials at the same location in the same direction.

a)	F3D.10S		F3D.20S		
Location and direction	H _B ± s	n	H _B ± s	n	(ΔH _B) _*
1	134 ± 7	18	145 ± 3	18	11
1'	133 ± 6	12	142 ± 3	12	9
2	126 ± 4	27	138 ± 3	27	12
2'	129 ± 3	15	136 ± 3	15	7

b)					
1 + 1'	134 ± 6	30	144 ± 3	30	10
2 + 2'	128 ± 5	42	137 ± 3	42	9
(ΔH _B) _{m=c}	6		7		

Secondary electron emission, figures 4b and 5b by back scattered electron technique. Secondary electron emission yielded at least a partial picture of the eutectic structure, but did not enable us to distinguish SiC particles from other phases, es-

pecially from large matrix particles. Electron back scattering produced the opposite effect, as is obvious from a comparison of figure 4a with figure 4b or figure 5a with figure 5b. Both these pictures and optical micrographs allow us to dis-

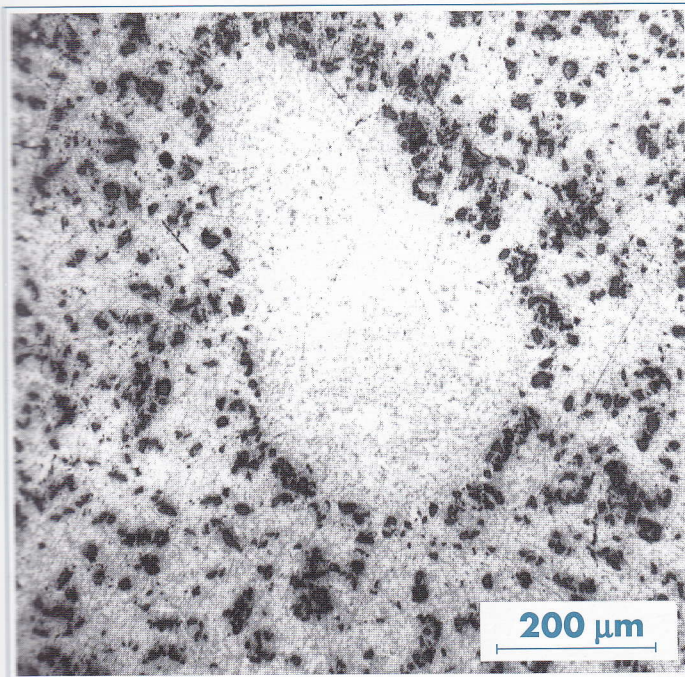


Figure 2: Non uniform distribution of SiC particles in direction 1 (see Fig. 1a), perpendicular to the cast plate surface, in F3D.10S material. Conventionally polished, no etching

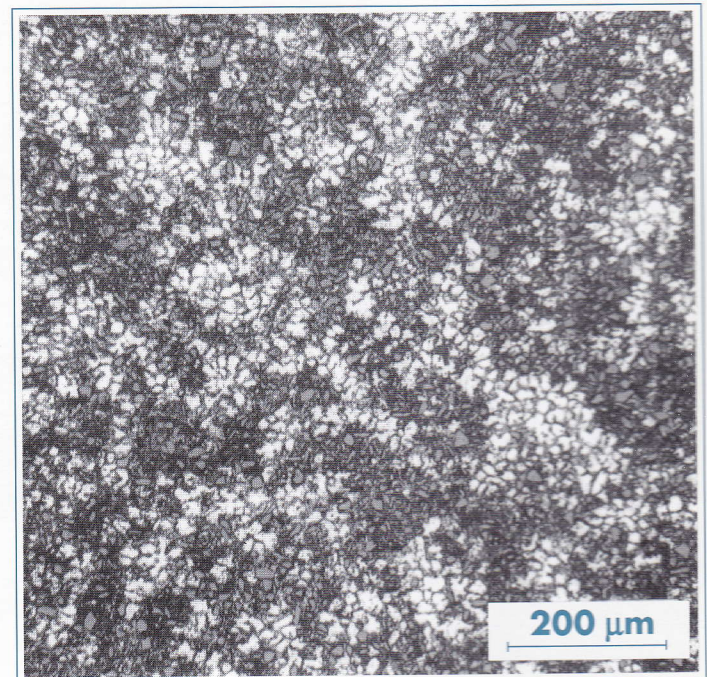


Figure 3: SiC distribution and microstructure of F3D.20S composite in direction 2 (see Fig. 1a), perpendicular to the machined casting surface. Polished conventionally and etched by 1 to 2 seconds in 1000 ml methyl alcohol + 50 ml $HC10_4$ + 15 ml HNO_3

cern various phases only by their morphology, for instance by their Chinese script or rounded appearance, but do not permit their color coding or discrimination by color intensity. This means that the prevalent types of image analyzers are not readily capable of rendering data for quantifying the proportions of volume occupied by individual phases. LECO 2001 instrument analysis indicated that both the proportion of area covered by SiC and the SiC particle size differ between the two composite types. In F3D.10S the proportion

of area varied from 7 to 22 %, the arithmetic mean of particle areas was about $70 \mu\text{m}^2$ at a length to width ratio of 1.85: 1. In F3D.20S the corresponding figures were 13 to 23% and roughly $100 \mu\text{m}^2$ at a length to width ratio of 1.7:1. Comparison of figures 4b and 5b demonstrates that in the latter picture the SiC particles are much coarser. The particle morphologies seen in figure 4 and 5 governed the choice of phases subjected to wave dispersion X-ray, or WMY, analyses. WMX analyses of never fewer than four particles pro-

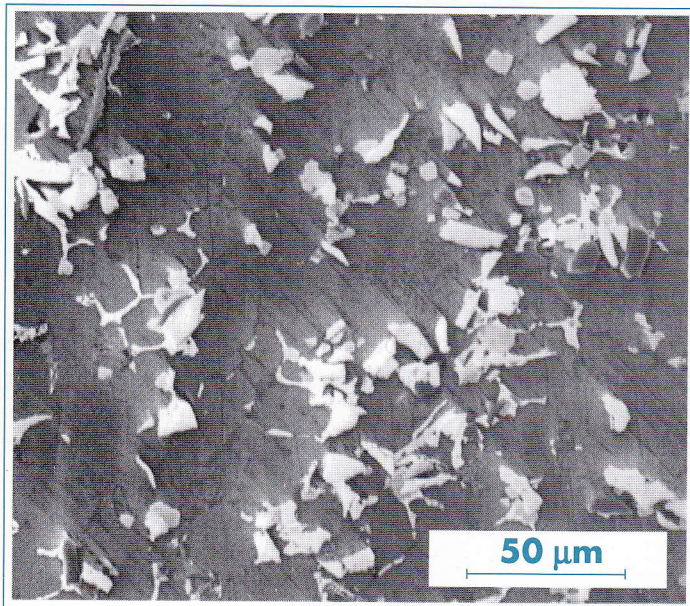


Figure 4a: Secondary electron image of F3D.10S in direction 2. A part from particles with a Chinese script morphology, phases are indistinguishable from SiC particles. Note trace of eutectic component

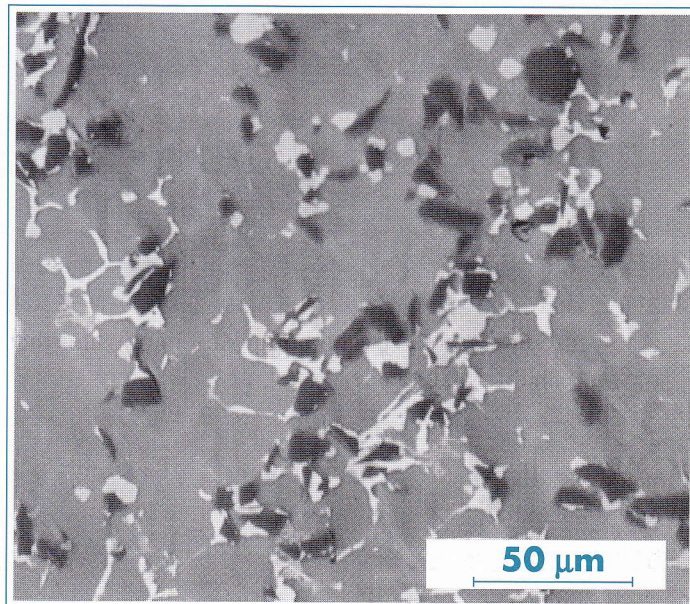


Figure 4b: Back scattered electron image of the same material as in Fig. 4a, in direction 2. The SiC particles are dark, the various phases light, the eutectic component is not visible at all

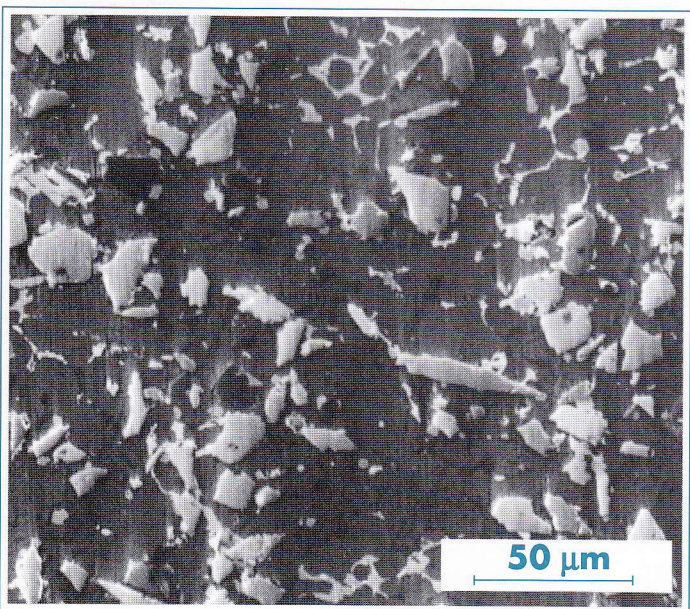


Figure 5a: Secondary electron image of composite F3dd.20S in direction 2; compare with Figs 4a and 5b

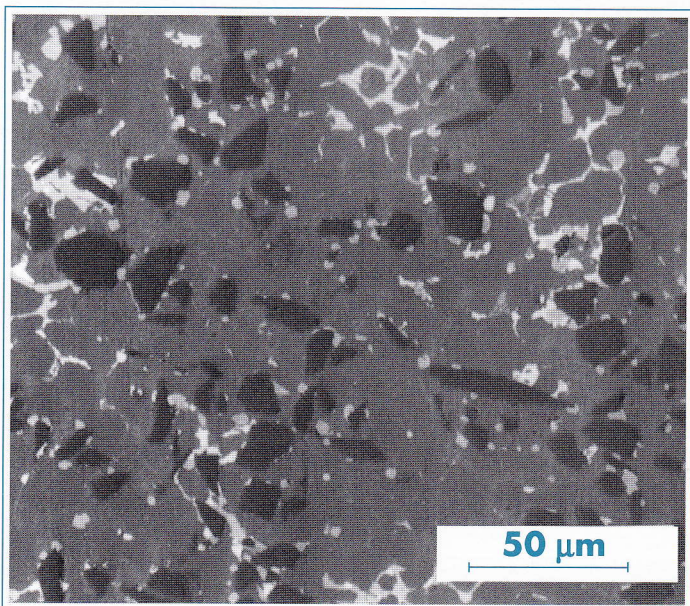


Figure 5b: Back scattered electron image of the same material as in Fig. 5a, in direction 2

duced the average composition and deviation data listed in Table 3. Analyses of images such as those presented in figures 4 to 5, and of the composition in Table 3 of some particles, and comparison of these findings with data in the literature [4], all seem to point to the following conclusions:

- The matrix microstructure is very fine-grained, as is obvious from Table 4, because SiC particles serve as preferential nucleation sites, so that the nucleation rate is high. SiC grains were often found to bear $(Al, Cu, Ni)_{15}(Fe, Mn)_3Si_2$ particles. It has been stated [5] that SiC particles of size d provide sites for the nucleation of matrix grains of size D , the two size being related as follows:

$$D = d [(1 - f_s)/f_s]^{1/3} \quad (1)$$

where f_s is the proportion of volume taken up by SiC particles. Table 4 suggests good agreement with experimental findings, but the differences between the grain sizes of the two composites is too slight to have any substantial effect, e.g. in its Hall-Petch contribution to hardening of aluminum alloys in the matrix of metal matrix composites.

- The particles resembling Chinese script apparently arose at lower temperature [4]; their composition indicates that they are an Al_3NiCu phase.
- WDX analyses of the grey background revealed practi-

cally nothing but aluminum: the dendrites of the matrix. Nevertheless, apart from Al dendrites the matrix also holds the above phases and the complex eutectic alloy seen in figure 4a and 5a, whose other phases are too fine to be exactly analyzed by the methods employed in this work. For instance, attempts to analyze a particle in figure 4a, of about 3 μm diameter, suggests that this might be an Al_7Cu_4Ni particle.

The conclusion to be drawn from all this are that the presence of SiC particles substantially refines the microstructure of the matrix; that the latter contains numerous intermetallic phases which are usually both very hard and brittle and that the difficulties encountered in processing these materials could be due to these intermetallic phases as well as to the SiC particles. Neither morphological nor chemical analyses confirmed the presence of a coarse acicular Al_5FeSi phase, which is very brittle in itself and by its shape effect further embrittles the matrix. Our attempts at approximate chemical analyses of the matrix, e.g. in direction 1 (figure 1) on F3D.10S material, showed an excess of silicon as against the specified composition. This surplus of Si (see Table 4) and superior hardness in direction 1 are probably due to a greater SiC particle concentration in the surface layers than in the core regions of the plates.

TABLE 3 - Results of WDX analyses of both composites: contents at % $\pm$ at %

La b	morph.	Al	Si	Mg	Fe	Mn	Ti	Ni	Cu	Likely phases	MMC	Dir -
1,2	Chinese script	61.80 ± 0.724	1.03 ± 0.826	0.04 $\pm$ 0.013	0.85 $\pm$ 0.081	0.04 $\pm$ 0.014	0.01 $\pm$ 0.005	19.43 $\pm$ ± 0.20	16.8 $\pm$ 0.239	Al_3NiCu	20S	2
1,4	Large rounded	71.41 ± 0.542	11.30 ± 0.37	0.02 $\pm$ 0.007	8.70 $\pm$ 0.064	6.84 $\pm$ 0.26	0.05 $\pm$ 0.066	0.80 $\pm$ 0.148	0.92 $\pm$ 0.226	$(Al, Cu, Ni)_{15}$ $(Fe, Mn)_3Si_2$	10S 20S	1 2
1,3, 4	grey back- ground	97.72 ± 0.385	1.47 ± 0.09	0.16 0.035	0.02 $\pm$ 0.013	0.03 $\pm$ 0.007	0.01 $\pm$ 0.013	0.06 $\pm$ 0.055	0.52 $\pm$ 0.234	Al + (Si)	10S 20S	1 2
2,4	"Matrix"	66.46 ± 0.17	30.08 ± 0.17	0.02 $\pm$ 0.01	0.51 $\pm$ 0.04	0.30 $\pm$ 0.03	0.14 $\pm$ 0.02	0.70 $\pm$ 0.04	1.78 $\pm$ 0.07		10S	1
1	SiC particles	0.30 $\pm$ 0.074	67.84 ± 0.71	0.01 $\pm$ 0.011	0.04 $\pm$ 0.013	0.02 $\pm$ 0.019	0.00 $\pm$ 0.007	0.06 $\pm$ 0.026	0.14 $\pm$ 0.08	SiC	10S 20S	2 1

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TABLE 4 - Matrix grain size D estimated from experimental findings (e.g. in direction 2 in Fig. 1) and as calculated by equation (1)

MMC	Experimental D [μm]	Calculated D [μm]
F3D.10S	15.71	19.62
F3D.20S	13.63	17.85

3.3 Mechanical properties

It has been reported previously that a fiber-reinforced matrix is not liable to fail by the initiation and propagation of a single dominant crack [6]. It may fail only by a sequence of events starting when a "critical" elastic stress level is attained and disrupts the fibers, continuing with the separation of particles from the matrix and, when this is complete, ending in failure of the matrix itself. In the composites investigated in this work, the separation of SiC particles from the Al-Si alloy matrix has no effect on the failure mechanism; however, its indirect influence cannot be altogether ruled out, because failure is generally a complex rather than a simple process. Moreover, we must bear in mind that the SiC particles in the examined materials are relatively small, measuring about 70 μm^2 in the F3D.10S and 100 μm^2 in F3D.20S MMC, and these sizes are in the category where both micro and macromechanisms play a role in the hardening and failure of MMC materials. The macromechanisms might be affected both by the proportion of volume occupied by SiC particles and, more particularly, by their distribution, which could cause internal splitting or delamination defects. The micromechanisms of work hardening and failure might be affected by the size of the SiC particles and of the matrix grains [5, 7], as has been outlined previously, and in the case of failure mechanisms by the presence of brittle phases (see Table 3), in other words by the heat treatment [8, 9]. Other authors [10] suggest that deformation or work hardening may affect the outcome by increasing the dislocation density: the many dislocations generated in the vicinity of undeformable SiC particles contribute to hardening. If that is indeed the case, then hardening should be proportional to the square root of the volume proportion of these particles [8]. Ashby [11] states that the dislocation density ρ [M^{-2}] around rigid particles is proportional to the local gradient of slip deformation, so that

$$\rho = (4\gamma f_s) / (rb) \quad (2)$$

where f_s is the volume fraction taken up by undeformable particles of radius r [m], γ is shear deformation (approx-

mately 0.01) [8], and b is the Burgers vector. If plastic deformation ϵ_t generates new dislocations, thus increasing their density, and if in the domain of true feasible stresses σ_t [MPa], given by $\sigma_t \in [R_e, R_m]$, hardening is described by

$$\sigma_t = K_1 + K_2 + \epsilon_t^{1/2} \quad (3)$$

where K_2 is a constant, and K_1 represents the proportionality limit, then $\rho \propto \epsilon_t$, so that equation (2) further implies that in the course of a tensile test, hardening will be proportional to $f_s^{1/2}$. According to a previous paper [8], equation (3) can be replaced by

$$\sigma_t = K \epsilon_t^n \quad (4)$$

where the hardening coefficient n may be taken proportional to $f_s^{1/2}$ and K [MPa] is a constant. The K_1 , K_2 , K and n values that apply to our work are listed in Table 5 together with the average R_e , R_{p02} , R_m and A_5 values. The n , $\epsilon_t^{1/2}$ and K_2 data, in other words hardening-related data in this Table display so large standard deviations, especially in the composite with less SiC, that these coefficients practically overlap; only the R_m and hardness figures in Table 5 and 2 exhibit any substantial difference between the two composites. Figure 6 is a graphic representation of equations (3) and (4) for constant values quoted in Table 5: figure 6a presents the σ_t versus ϵ_t curves for equation (4), figure 6b the σ_t versus $\epsilon_t^{1/2}$ lines for equation (3). The diagram also lists the average R_e and R_u values for both of the examined composites. The differences in R_e and R_u between the two materials are closer to the differences indicated by equation (4) than to those implied by equation (3). The fact that the differences in R_u values are greater is due to the incidence of premature fracture, i.e. fractures occurring in the elastic deformation domain, or before the maximum stress was attained in the domain of uniform plastic deformation. These premature fractures were generated more easily in the F3D.10S material; as is clear from Table 5 and the lower ϵ_u (or maximum plastic uniform deformation) values. The standard deviations are larger than in the F3D.20S specimens. Both these facts are attributed to a less homogeneous distribution of SiC particles, especially on the surfaces of tensile test specimens (direction 1 in figure 1). This view is supported by the notch toughness values recorded

TABLE 5 - Mechanical properties and hardening coefficients for equations (3)

MMC	R_e MPa	R_{p02} MPa	R_m MPa	A_5 %	K_1 MPa	K_2 MPa	KC J cm ⁻²	ϵ_u	K MPa	n
F3D.10S	93.2± 14.0	14.2± 23	184.0± 16	0.2±0.3	62±12	1734± 279	3	0.0049± 0.0019	890± 160	0.30± 0.04
F3D.20S	104.9± 7.6	151.6± 16	209.0± 9	0.6±0.4	78±11	1688± 161	3	0.0060± 0.0014	775± 76	0.26± 0.02
Mostly broken off by premature fractures										

on the keyhole notch specimens, as listed in Table 5: these values are virtually identical for both materials and both test directions, and are so low as to indicate brittle failure.

The R_{p02} , R_m and A_5 figures ascertained in this work are all lower than those quoted in the maker's handbook [1], where the minima listed for the worst possible heat treatment case are as follows:

F3D.10S: $R_{p02} = 152 \text{ MPa}$ $R_m = 241 \text{ MPa}$ $A_5 = 1.2 \%$
 F3D.20S: 186 MPa 303 MPa 0.8%

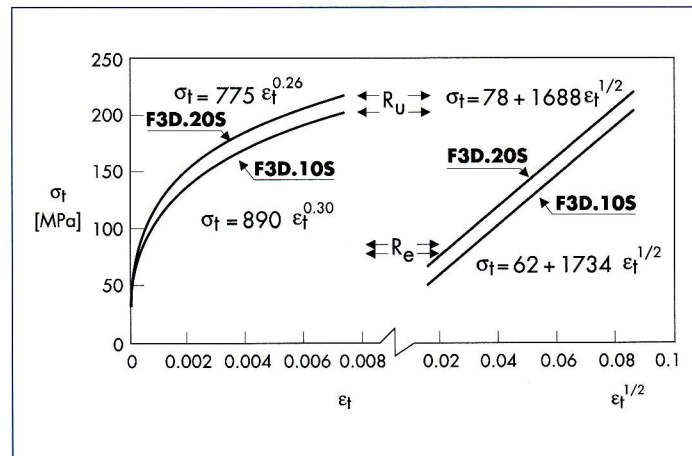


Figure 6: Diagram of equations (3) and (4), and comparison of the R_e (proportionality limit) and R_u (maximum true stress) values for the two examined composites: (a) equation; (b) equation (3).

CONCLUSION

A study of the basic structural and mechanical properties of two commercial metal matrix composites which have an Al-Si alloy matrix discontinuously strengthened with (SiC)_p particles (see Table 2, as marketed by Alcan as Duralcan F3D.10S (with 10% of SiC added) and F3D.20S (with a 20% SiC addition), has led to the following conclusion:

1. The very hard SiC particles with relatively low tensile strengths precluded the preparation of specimens by conventional techniques or with conventional tooling, at least in an economical manner. The tensile test specimens had to be hand polished after their machining, to rectify defects caused by the impression or detachment of SiC particles into or from the matrix. Despite this precaution, premature fractures were common.
2. The R_{p02} values established in this work were lower than expected, an indication of the quality of these materials. The maker's handbook claims that the R_{p02} difference between the two materials is 34 MPa, but the values ascertained in this work differ by only 7.4 MPa. This may be due to the homogeneity of distribution of the SiC particles and/or to the proportion of volume they occupy, a view also supported by the finding that in the F3D.10S

grade the SiC concentration exceeded the maker's specification.

3. The F3D.20S or higher -SiC composite displayed better R_e , R_{p02} , R_m and A_5 values as well as finer-grained matrix than the .10S material. The finer grain is ascribed to the more frequent heterogeneous nucleation processes that take place at the more numerous SiC particles while the matrix solidifies.
4. The Al-Si alloy matrix was found to contain, a part from Al dendrites and Al-Si eutectic material, some $Al_{15}(Fe, Mn)_3Si_2$ and Al_3NiCu phases. Analyses of further but smaller particles such as Al_7Cu_4Ni produced only approximate results.
5. Strain hardening in these materials was found to conform to the expression

$$\sigma_t = K_1 + K_2 \epsilon_t^{1/2}$$

The K_1 constants for the two composites differ, but the K_2 values do not; it follows that the greater volume proportion of SiC contributes to hardening mainly by its effect on the proportionality limit, which broadens the elastic deformation domain.

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CAPTION TO FIGURES

- Figure 1: Sampling points for microstructure investigations (a), tensile bars (b) and notch toughness (c) specimens cut from plates of 12.7 x 140 x 140 mm. Locations and directions are marked 1 (as cast surface), 1' (approximately 1.5 mm beneath the surface), 2 (middle of section surface on a ruptured specimen).
- Figure 2: Non-uniform distribution of SiC particles in direction 1 (see Fig. 1a), perpendicular to the cast plate surface, in F3D.10S material. Conventionally polished, no etching.
- Figure 3: SiC distribution and microstructure of F3D.20S composite in direction 2 (see Fig. 1a), perpendicular to the machined casting surface. Polished conventionally and etched by 1 to 2 seconds in 1 000 ml methyl alcohol + 50 ml HC10_4 + 15 ml HNO_3 .
- Figure 4a: Secondary electron image of F3D.10S in direction 2. A part from particles with a Chinese script morphology, phases are indistinguishable from SiC particles. Note trace of eutectic component.
- Figure 4b: Back scattered electron image of the same material as in Fig. 6a, in direction 2. The SiC particles are dark, the various phases light, the eutectic component is not visible at all.
- Figure 5a: Secondary electron image of composite F3D.20S in direction 2; compare with Figs 6a and 7b.
- Figure 5b: Back scattered electron image of the same material as in Fig. 7a, in direction 2.
- Figure 6: Diagram of equations (3) and (4), and comparison of the (proportionality limit) and R_u (maximum real stress) values for the two examined composites: (a) equation (4); (b) equation (3).