rather encouraging as it is somewhat higher than observed for 350 maraging steels aged to peak hardness.[14] However, both the base and base + Ni + Al steels exhibit nonductile fracture modes when tempered at 550 °C. Studies of the thermal and mechanical stability of the retained austenite as a function of tempering temperature for both steels suggest the possibility that these fracture modes are associated with the decomposition of retained austenite and a decrease in the mechanical stability of the retained austenite.

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A Method of Assessing the Reactivity between SiC and Molten Al

D. J. LLOYD and I. JIN

A metal matrix composite with considerable potential as an engineering material is Al reinforced with SiC. All the methods of making these composites involve contact between molten Al and SiC reinforcement at some stage of their fabrication. As a result, it is important to assess the stability of SiC when exposed to molten Al and to develop methods of measuring the extent of the reaction. In this paper we want to consider these two aspects of Al-SiC composites.

SiC is thermodynamically unstable in molten Al[1,2] and reacts to form aluminum carbide according to the reaction:

$$4Al + 3SiC \rightleftarrows Al_4C_3 + 3Si \qquad [1]$$

The extent of the reaction has been followed by measuring the intensity of the aluminum carbide and silicon X-ray peaks from the composite.[3] The reaction can also be followed by chemical analysis of the composite and, in the case of fiber composites, by measuring the thickness of the fiber-matrix interaction layer.[4]

There is an alternative method for those alloys where the influence of Si on the phase diagram is known, and this is determining changes in the liquidus temperature. As the reaction occurs according to Eq. [1], the Si content of the alloy increases and the liquidus temperature decreases in most alloys of interest.

In this paper the effect of silicon carbide reaction on the liquidus of an AA6061-20 vol pct SiC particulate reinforced composite is considered. The initial composite was remelted and held for one hour at 675, 800, and 900 °C and then resolidified. The liquidus temperatures of the remelted material were then measured from differential scanning calorimetry (DSC) traces, using a 20 deg per minute heating rate, under argon flowing at 60 ml/min and a specimen size of 15 to 20 mgms.

The changes in the Al-Si liquidus temperature with remelt temperature are shown in Figure 1, and as expected the liquidus temperature decreases with increasing remelt temperature. The data points are the mean of five samples with a scatter of ±2 °C.

These temperatures can then be plotted on the liquidus surface of the Al-Mg-Si phase diagram, as in Figure 2, taking the matrix composition as Al-1 wt pct Mg-0.8 wt pct Si, and the changes in the Si content of the composites obtained.

Using these silicon concentrations the extent of silicon carbide transformation can be obtained as follows:

$$\text{wt pct SiC} = V_c\rho_c * 100/[V_m\rho_m + V_c\rho_c] \qquad [2]$$

where V_c and V_m are the volume fraction of silicon carbide and

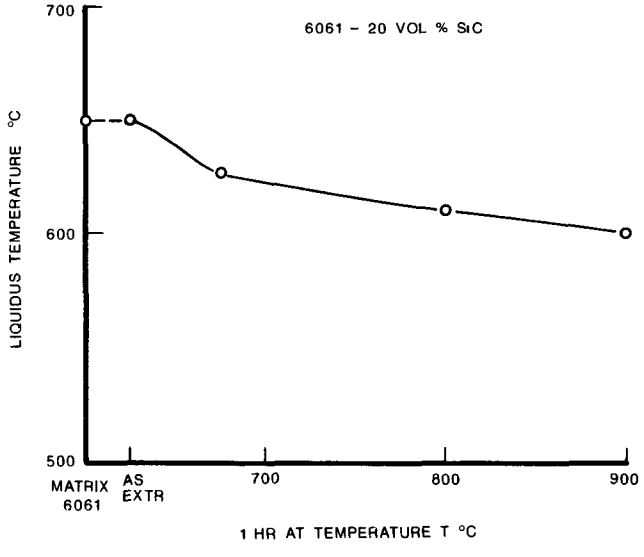

Fig. 1 — Change in the liquidus with remelting temperature.

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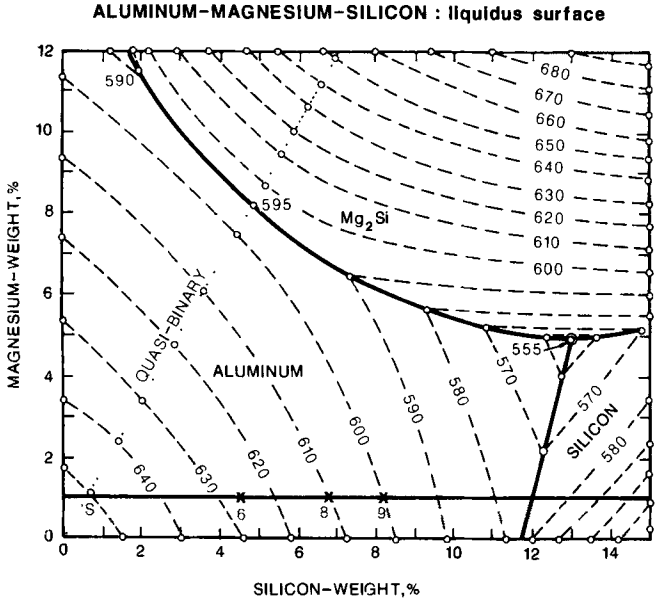


Fig. 2—Liquidus surface of AA6061 and the compositions of the remelted alloys: S—as-extruded, 6—after 1 h at 675 °C, 8—after 1 h at 800 °C, and 9—after 1 h at 900 °C.

matrix, respectively, and ρ_c and ρ_m are the densities of silicon carbide (3.217 g/cc) and matrix (2.7 g/cc), respectively.

For the 20 vol pct SiC in the present composite, the wt pct of SiC is 23.

The wt pct of silicon available from this carbide is:

$$\text{wt pct Si} = 23 * [\text{Atomic wt Si/Atomic wt SiC}]$$
$$= 16.1 \text{ pct} \quad [3]$$

If the liquidus temperature corresponds to Y wt pct Si, the fraction of silicon carbide conversion, X, is obtained from

$$X = [Y - 0.8]/16.1 \quad [4]$$

where the 0.8 allows for the initial Si content (in wt pct) in the alloy. Figure 3 shows the changes in silicon content and the degree of silicon carbide conversion with increasing remelt temperature.

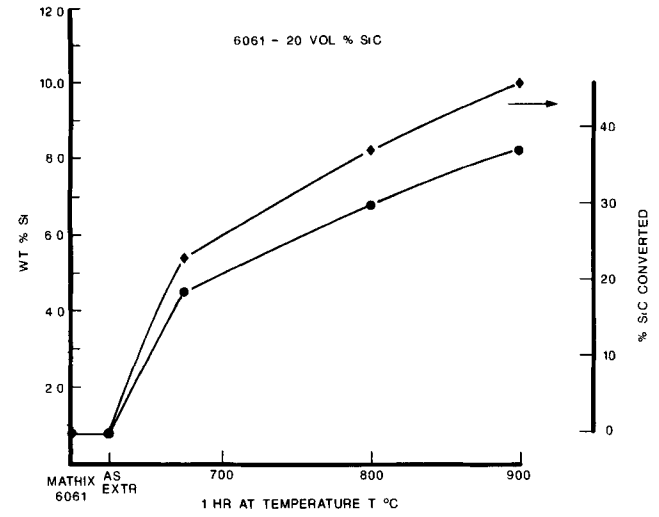


Fig. 3—Changes in the Si content of the composite and the percentage of SiC converted after 1 h at different temperatures.

The results show that following changes in the liquidus temperature of the composite is a viable method of following the interaction of silicon carbide with molten aluminum. The degree of accuracy depends on several factors. The homogeneity of the reacted composite will affect the liquidus temperature, particularly when the DSC method, with its small sample size, is used to determine the liquidus. The liquidus will vary with the heating rate used to obtain the DSC trace. For example, using a heating rate of 10 deg per minute instead of 20, reduced the liquidus by 2 deg. This is the extent of scatter from sample to sample. Depending on the matrix alloy, the DSC trace can show other interactions in addition to the liquidus. With the Al-Mg-Si, AA6061 matrix alloy, depending on the composition and degree of homogenization, the following invariant reactions may be observed:[5]

Quasi-binary eutectic Al + Mg_2Si at 595 °C

Ternary eutectic Al + Mg_2Si + Si at 555 °C

Binary eutectic Al + Si at 578 °C

But commercial alloys also contain iron (and chromium), which exhibit other invariant reactions to form Fe_2SiAl_8, $FeSiAl_5$, and $FeMg_3Si_6Al_8$.

The measured temperatures on the DSC trace corresponding to these reactions will again depend on heating rate, and Figure 4 shows two incipient melting reactions prior to general melting in the initial composite material. These incipient melting reactions occur at temperatures close to the invariant reactions:

Al + Si at 577 °C

Al + $FeSiAl_5$ + Si at 575 °C

Al + Si + $FeMg_3Si_6Al_8$ at 567 °C

Al + $FeSiAl_5$ at 612 °C

However, the Mg_2Si and Si reactions cannot be ruled out because the DSC method will tend to overestimate the reaction temperature.

The extent of the invariant reactions involving Si and Fe-Si intermetallics will increase with increasing Al_4C_3 formation, and could be used to monitor the reaction between Al and SiC. However, changes in the liquidus temperature are easier to measure. The DSC method for determining the

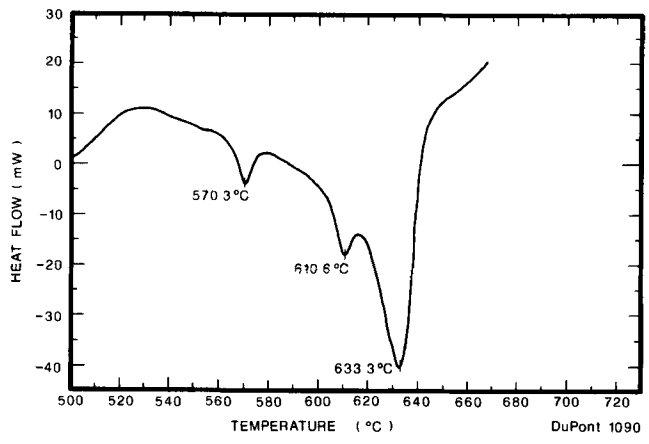
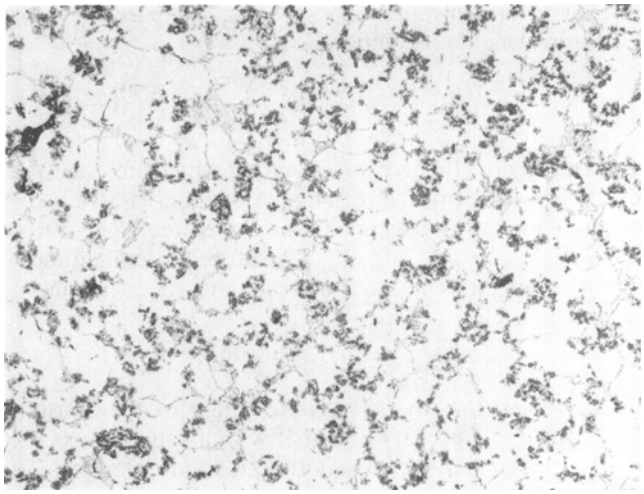
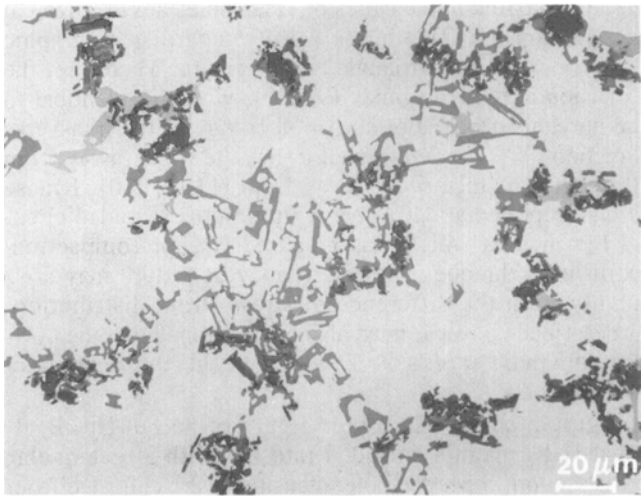


Fig. 4—DSC trace after 1 h at 675 °C.



(*a*)



(*b*)

Fig. 5—Microstructure of the composite after 1 h at 800 °C, showing increased Si and Al_4C_3.

liquidus is a very convenient method, and should be sufficiently accurate for composites with normal levels of reinforcement, *e.g.*, greater than 10 vol pct.

The accuracy of the method can be assessed in principle, by quantitative metallography, measuring the volume fraction of silicon phase. The aluminum carbide content can also be determined, but the silicon phase is much easier to resolve, as seen from Figure 5. The silicon content can also be measured by chemical analysis, after passing the molten composite through a ceramic filter to remove the particulate, but this can be done only for low temperatures and short reaction times because the fluidity of the melt decreases rapidly with increasing Al_4C_3 formation. A few experiments of the latter type have been carried out and gave results in good agreement with the DSC data, but improved chemical analytical techniques need to be developed to investigate a broader temperature range.

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Texture Induced Magnetic Anisotropy in Fe-Nd-B Magnet Prepared *via* Rapid Solidification and Hot Extrusion Techniques

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There are three basic criteria that have to be met by Fe (or Co)-R (rare earth) magnets to achieve optimum square-loop behavior in the demagnetization regime and hence high energy products: (1) the material must consist of an assembly of fine particles in order to possess high coercivities, H_{ci} and H_c; (2) the particles must exhibit microscopic magnetic anisotropy; *i.e.*, they must have a preferred easy axis of magnetization; and (3) the microscopically anisotropic particles must be capable of being aligned with their easy axes all pointing in the same direction in order to achieve high B_r. Through the rapid solidification technology, the first criterion can be met in the Fe-R-B system; while the intrinsic characteristics of the Fe-R-B alloys automatically fulfill the second criterion. Recently a hot extrusion technique has been reported to be employed to satisfy the third criterion.[1,2]

The present report examines the magnitude and distribution of normal strain under three different die openings, *i.e.*, of circle, square, and rectangle, respectively, and their effects on evolution of texture in the magnet prepared by hot extrusion of rapidly solidified melt spun powders. At the same time, the anisotropic properties found in the extruded Fe-Nd-B based magnet containing HfB_2 as a grain refiner[1] have been characterized.

The proprietary melt spinning technique was employed to produce rapidly quenched narrow (10 μm thick, 100 μm wide) and short amorphous powders (about 60 mesh), using a substrate speed of 44 m/sec. In this process, the molten metal is extracted by a knife-edged rotating wheel and converted into short filaments solidified at a uniform rate of 10^6 °C/sec. Amorphous material is essential to make a promising extruded magnet.

X-ray diffraction analysis was carried out to examine the amorphous state of melt spun powders using Cu-Kα radiation. The melt spun powders (about 60 mesh) were cold compacted into mild steel (AISI 1080) cans using 50 ksi pressure, followed by hot evacuation at 400 °C until the vacuum reached 0.5×10^{-6} torr when the cans were sealed off. A constant extrusion ratio, 12/1, was employed for all die openings. The extrusion of square and circular bar was carried out at 1000 °C while the bar of rectangular cross section was extruded at 1080 °C due to its required high ex-

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