presence of a small quantity (100 mg) of powdered beta-alumina ($Na_2O . 11Al_2O_3$) which slowly evolves sodium oxide at high temperature [13]. While the volatile alkali metal oxides, or the metals themselves, are likely contaminants, it is more difficult at this stage to be certain about which of the other impurities in the alumina could be involved. Steele and Williams [12] argue that both CaO and MgO would be appreciably volatile and, with an assumed partial pressure of oxygen of 10^{-18} atm, lead to vapour pressures of the metals in the region of 10^{-3} atm at 1300°C. Work is now in progress to determine the extent of pick up of metals by silicon under these conditions.

It is clear that under "high purity" conditions the alpha silicon nitride forming reaction tends to be kinetically highly favoured. It is also apparent that in some circumstances very small amounts of impurity can be more important in determining the alpha/beta silicon nitride ratio than the overall volume of oxygen in the system. The formation of the isostructural, and chemically similar beta germanium nitride in the nitridation of germanium powder [2, 14] also appears to be promoted by oxide impurities at certain concentrations.

The results of this work may help to explain the apparently favoured production of beta silicon nitride in the nitridation of silicon powder compacts at temperatures above the melting point of silicon [15]. The significant factor may not be a direct effect of temperature on reaction rate, or the appearance of liquid, but an increased rate of evolution of catalytic oxide impurities from furnace refractories. A further conclusion is that true beta silicon nitride will be very difficult to prepare in a state where it can be guaranteed that impurities are not a factor contributing to the stability of the phase, and until the role of trace impurities is better understood existing data on phase stability in the higher temperature regions may need to be treated with caution.

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On spiral eutectic growth

Spiral eutectic structures are relatively uncommon. They have been reported in the Al–Th [1] and in the Zn–Mg systems [2-4]. Recently, we reported some observations on spiral growth of the αAl–Mg_2Si eutectic [5] and discussed this growth in relation to instability behaviour of the Mg_2Si phase. Primary crystals of the latter, in the system investigated, grow as

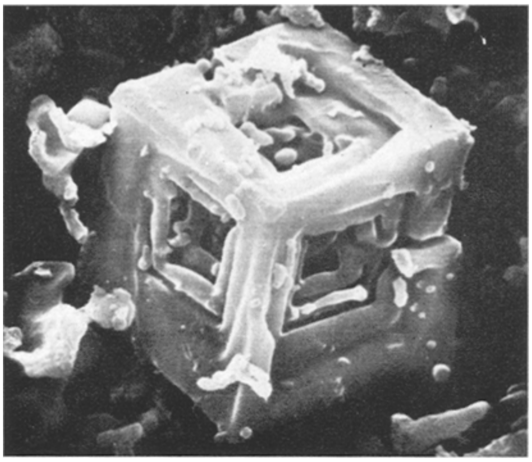

Figure 1 Hopper crystal of Mg_2Si extracted from Al—Mg—Si alloy (× 3500).

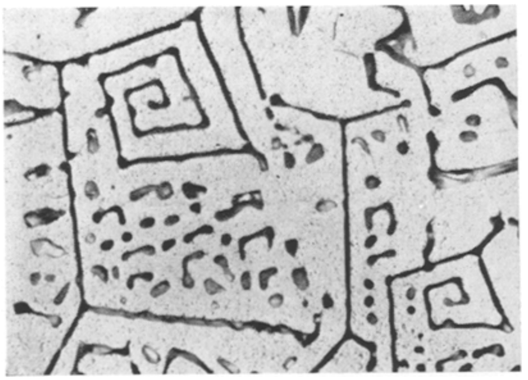

Figure 2 Spiral eutectic of αAl—Mg_2Si in Al—Mg—Si alloy (× 550).

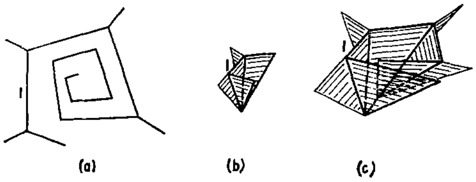

Figure 3 Growth of first spiral turn by edge branching of Mg_2Si plates. (a) Outline of Fig. 2. (b) Branching of plate 1 at either edge. (c) Branching of plates in sequence to form first turn.

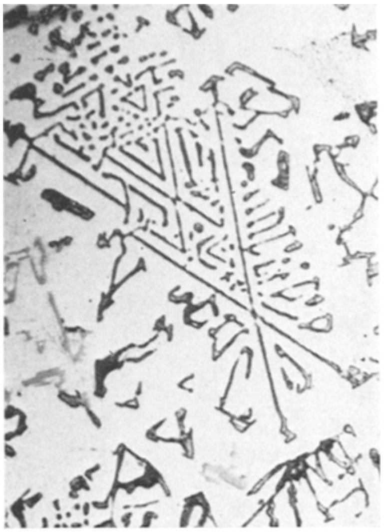

Figure 4 αAl—Mg_2Si eutectic in longitudinal section (× 450).

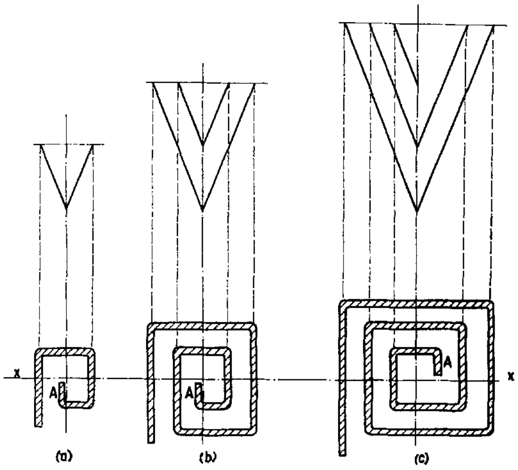

Figure 5 Cross-sections of spiral during growth. (a) Cross-section corresponding to first turn as in Fig. 3b. (b) Development of second turn. (c) Section corresponding to Fig. 4.

hopper forms. The object of this letter is to present some additional geometrical features of these spirals as observed in the experiments, and describe briefly the mode of growth.

The alloy studied had the composition 80% Al, 10% Mg, 10% Si, and was melted in a vacuum furnace, the vacuum being broken to introduce the magnesium. Solidification was carried out in the crucible, cooling rates of 50°C min^{-1} being recorded. In some experiments, 0.01% Na was added. Spiral eutectic growth of the αAl-Mg_2Si eutectic was observed in all cases.

Fig. 1 shows an extracted hopper growth Mg_2Si crystal while Fig. 2 shows spiral structures in the αAl-Mg_2Si eutectic observed in metallographic cross-section [5]. Fig. 3a gives the outline of the spiral observed in Fig. 2 and Fig. 3b and c, give the corresponding branching mode of

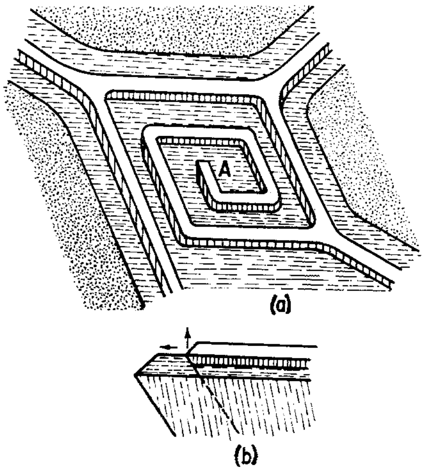

Figure 6 (a) Mg_2Si phase grows ahead of the eutectic interface enabling edgewise growth at A. (b) Leading edge of Mg_2Si grows at angle with vertical.

the eutectic leading to the first turn of the spiral. These features are as observed by metallography and by scanning electron microscopy of extracted crystals.

The Mg_2Si phase in the eutectic grows as (100) and (110) oriented plates. Starting at the Mg_2Si plate marked 1 in Fig. 3b, branching occurs at either edge, as the αAl phase grows over the Mg_2Si plate. The plates then branch again in turn, to give the first spiral loop with attached appendages shown in Fig. 3c.

Fig. 4 shows a longitudinal metallographic section of a spiral eutectic formation. This microstructure is arrived at by the stages of growth shown in Fig. 5 a, b and c. The cross-section of Fig. 5a corresponds to the first loop shown in Fig. 3c.

The mode of edge branching observed in the eutectic, which leads to the first spiral loop, is characteristic of the mode of growth of the primary crystal in hopper form. For the latter type of growth, crystal edges are unstable relative to crystal faces, and this is expressed in the branching behaviour of the eutectic Mg_2Si plate. Once the first loop has formed, growth proceeds as a spiral through the stages shown in Fig. 5, by a combination of lengthwise growth with a lateral component at A. The outer branches of the first loop shown in Fig. 3c have only limited growth and become redundant. Fresh turns add on at the centre by solute attachment to the leading edge A, providing the lateral growth component to wind the spiral. Lengthwise growth of the eutectic, is by the usual short range diffusion between phases.

The Mg_2Si plates could not spiral, if this phase did not grow ahead of the eutectic interface as is noted in eutectics of this irregular character [6]. The situation is shown in Fig. 6 the leading edge of the spiral is required to grow at an angle to the vertical.

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Comments on "Dependence of room temperature fracture strength on strain-rate in sapphire"

In a recent paper, Pollock and Hurley [1] have studied the strain-rate dependence of the fracture of sapphire filaments at room temperature. They show, very convincingly, that a strain-rate dependence of the strength is obtained, even when an environment is excluded from the region adjacent to the fracture initiating flaws.

They interpret the strain-rate dependence as a manifestation of dislocation-assisted slow crack growth. We believe that this interpretation is speculative and propose that the observed behaviour is due to another phenomenon,