MECHANISM OF STRUCTURE FORMATION IN SPLAT-COOLED Al-Fe ALLOYS

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Introduction

An earlier paper (1) described a microstructural transition (fig.1) in cross-sections of Al-Fe alloy splats. Surface regions, designated 'zone A', showed a limited etching response and about twice the hardness of central regions darkened by etching, and designated 'zone B'. Electron microscopy of thinned samples and x-ray diffraction showed (a) zone A to contain a fine-grained second phase network (spacing $\sim 0.03\,\mu m$) within αAl grains, and supersaturated regions, (b) zone B to have a coarser second phase network (spacing $\sim 0.03\,\mu m$), and (c) concurrent development of primary intermetallic (notably including $FeAl_6$) growth.

This paper presents more results, including replica studies, as a basis for discussion of possible transformation mechanisms operative. A particular aim, in the light of ancillary studies (2-4), was to assess the role of primary intermetallic growth with respect to the zone A to B transition in these splats.

Results

Fig.1 shows a carbon replica of a zone A to B transition. The linear array normal to the zone B boundary is notable. A

similar array (with a smaller spacing) was observed (fig.2a) adjacent to regions apparently free of substructure in thinned regions of splat otherwise showing typical (1) zone A network structure. Also, radial growth with this morphology from primary intermetallic growth centres was evident where these were present (fig.2b). Replicas of zone B (figs.3a and b) as well as zone A regions showed variations in frequency of identifiable primary intermetallic particles. Dendritic branching is evident in the αAl background structure in fig.3a.

Discussion

Two main points emerge from this study:
(a) An 'elongated dendritic' (5), or cellular, growth morphology is evident where the growth direction makes a small enough angle with the plane of observation (figs.1, 2a and b):
(b) The volume fraction of identifiable primary intermetallic phase is always well below the maximum total intermetallic possible, and as low as 1% or less in some zone A areas, so that most of the intermetallic forms intercellularly or interdendritically, accompanying or following αAl formation.

There would seem to be three basic mechanisms which might form the polycrystalline second phase network distribution in zone A. Spinoidal decomposition (6) in the liquid or solid phases, however, lacks thermodynamic (7) or constitutional (8) support, and a second transformation would be needed to give the phases observed (the second phase has been identified (4) as FeAl, basically a bcc structure). Furthermore, in spite of indications (9) that relatively coarse cellular structures develop on ageing a dilute Al-Fe alloy, the finely polycrystalline structure of our second phase would not be expected if cellular precipitation (10) were responsible. Also, a cell spacing smaller than typical of zone A would be expected from the diffusion coefficient of Fe in solid Al (11) and the available time for diffusion (2).

This consideration is no limitation for solidification, however, which thus accounts satisfactorily for the microstructure. The absence of dendritic branching and evidence such as that of figs.1, 2a and 2b suggest that the αAl in zone A has the rod-like growth morphology which displaces normal plate-like cellular-dendritic growth at high freezing rates (12). Thermodynamic (13) and kinetic (14) considerations predict the distribution coefficient $k \rightarrow 1$ at sufficiently high undercooling or freezing velocity. The magnitude of the lattice parameter change observed (1) suggests $k = 1$ might be approached in some regions (e.g. fig.2a). The marked second phase pattern elsewhere in zone A implies a k-value somewhat less than unity. Theory (14) predicts a transition range of freezing velocity in which k increases from its equilibrium value (0.03 for Al-Fe) to unity and that the freezing velocity $\sim$ 40mm/s (2) characteristic of zone A could be in this range. It is easily shown from the Scheil equation, as modified (15) for solute back-diffusion, that supersaturation even to ten times equilibrium (i.e. to $k = 0.3$) still allows enough microsegregation to form interdendritic intermetallic.

Possible mechanisms for the displacement of zone A by zone B might involve interface stability, competititive growth or nucleation concepts (1). Except possibly for regions showing no second phase (fig.2a), αAl in zone A freezes by non-planar growth. This is consistent with the conditional stability criterion (16) for the possible temperature gradients (1) and growth rates (2) applicable and with the absolute stability criterion (16) provided $k<0.75$. Perturbation theory has not yet been developed for suppression of eutectic growth in favour of growth of either component phase. Our preliminary steady state experiments however indicate that both coupled eutectic and primary $FeAl_3$ growth are unstable with respect to αAl growth at applicable growth rates (2). The limited primary intermetallic growth observed in splats plays a role in acting as centres for αAl growth but there was no clear evidence for any more direct role in giving rise to instability of zone

A relative to zone B.

The key problem is then to account for the difference in cell spacing between zones A and B. Heat transfer considerations (1) and observations on Al-$CuAl_2$ eutectic splats (2) indicate only small changes in cooling rate and freezing velocity with position in the splat thickness. Marked changes would be expected, however, if a process akin to a columnar to equiaxed crystal transition occurred. The earlier suggestion (1) that zone A was columnar has been confirmed by heavy etching, and electron microscopy indicates an αAl grain size much larger in zone B than in zone A, as expected if zone B were equiaxed. In ingots, however, no discontinuity in dendrite cell size occurs at the columnar/equiaxed boundary (17) because dendrite coarsening (15) can occur in both regions. This coarsening cannot occur in zone A since cellular growth appears throughout.

The mean hardness difference between zones A and B can be accounted for by modifying, for a continuous second phase, Hansen's model (18) of hardening by a network of discrete hard particles. If the Orowan stress required to force piled-up dislocations between particles is replaced in his analysis by the yield stress of a continuous second phase, hardness values in good agreement with those reported (1) for zone A and B can be derived.

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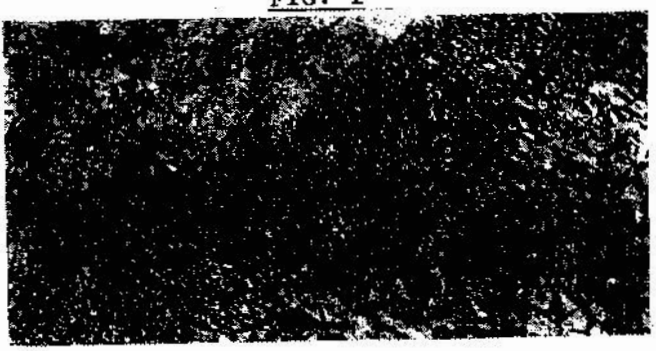
FIG. 1



Two stage plastic/carbon replica of etched normal section of an Al 4^{a}/o (8^{w}/o) Fe splat, showing zone A to B transition.

FIG. 2a

FIG. 2b

Electron micrograph of thinned splat (a) Al 2^{a}/o (4^{w}/o) Fe showing cellular growth and areas apparently free of second phase network (b) Al 4^{a}/o (8^{w}/o) Fe showing primary intermetallic acting as growth centre for αAl.

FIG. 3a

FIG. 3b

Single stage carbon replicas of an etched section of Al 4^{a}/o (8^{w}/o) Fe splat showing αAl dendritic branching and variation in amount of identifiable primary intermetallic phase.

DISCUSSION

A. Tonejc: Did you obtain samples containing only zone A, because on your first slide you showed two distinct zones (A and B) ?

H. Jones: We obtained samples containing zone A and zone B and studied both by replica and transmision electron microscopy. Zone A had a cell size 0.03 μm, while zone B had one of 0,3 μm for Al 4 at% Fe produced by the gun technique.

D. Kunstelj: We have performed some experiments on Al-Fe system and we observed globular precipitates at the cell boundaries. How do your observations fit this fact ?

H. Jones: H. Jacobs will speak about it.