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Creep behaviour of iron–aluminium–chromium intermetallics strengthened by nano-sized oxide particles



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ABSTRACT

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1. Introduction

Iron aluminides have been extensively examined over the past decades, especially with a view to their possible application as structural materials at high temperature for aggressively oxidising/ corrosive environments [1-5]. One drawback which has significantly retarded their application is the rapid loss of strength and creep strength at temperatures above 500–600 °C [2,4], and this has led to extensive programmes of alloy development and microstructure control whereby fine and stable precipitates or dispersoids can be introduced into the intermetallic matrix. Mechanical alloying has been widely examined, both as a way of producing intermetallic matrices [6–10] as well as a way to introduce stable oxide particles, leading to the so-called Oxide Dispersion Strengthened (ODS) materials [4,6-11]. Previous studies of ODS iron aluminides have indeed shown excellent creep behaviour, especially at higher temperatures [4,6,7,11]. In a similar way, ODS Fe-Cr alloys have received much interest for structural applications at high temperatures, for example in the nuclear industry, where new nanostructured alloys based on Fe–Cr with small additions of Y₂O₃ as well as elements such as Ti [12.13] have been developed. These new materials show outstanding high temperature strength [14,15] and creep resistance [16-19], and are stable at high temperatures as well as under irradiation [12,20,21].

Recent studies [22] of ODS Fe–Al–Cr alloys have shown that nanosized Al–Y–O oxide particles form in the intermetallic matrix

The creep behaviour of some oxide-dispersion-hardened Fe–Al–Cr iron aluminides has been examined at 700–750 °C and compared with similar dispersion-hardened steels and commercial Fe–Cr alloys. Very fine oxide particles form during material processing, which are generally fcc Y₂O₃ phase with a thin disk morphology, and partially coherent with the ordered FeAl matrix. The materials show excellent creep strength at the tested temperatures which may be related to the oxide particles being barriers opposing dislocation motion. Precise details of dislocation–particle interactions are not clear, but it is suggested that the fine particles are cut relatively easily by the dislocations.

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during processing, most of which have a simple fcc structure, albeit with some evolution towards more complex monoclinic structure. This study showed very good flow strength in conventional tension and compression testing up to high temperatures, with this strength interpreted in terms of contributions from the ordered matrix, the fine grain size, and the many fine particles. It was also shown that the elastic moduli changed significantly with alloy composition and with temperature, and this played an important role in modifying the strength.

The present work examines the creep behaviour of the same ODS Fe–Al–Cr materials at temperatures of 700–750 °C, comparing their behaviour with that of a similar ODS binary Fe–Al alloy. The very limited amount of material available has meant, however, that creep testing has been carried out over a limited range of stress and temperature and analysis of creep rates for the determination of controlling mechanisms has not been possible. The observed creep behaviour is related to the morphologies of the particles present and to the dislocation microstructures found within the crept materials.

2. Experimental details

Details of alloys and their preparation have been given before [22], and the creep testing procedure detailed elsewhere [4]; only a brief reminder is given here. Three ternary Fe–Al–Cr alloys with Al content in the range 30–40% and Cr content in the range 5–10% (atomic per cent is used throughout to indicate metallic composition) were prepared by Plansee AG using similar industrial pilot-scale milling conditions to earlier studies [3,4,11] on the binary Fe–40%AI

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а

100nm

b

alloy. The alloys are referred to throughout as FAC-1, FAC-2, FAC-3 and FAC-4, with nominal compositions of Fe–40Al, Fe–39Al–10Cr, Fe–33Al–8Cr and Fe–34Al–7Cr, all confirmed as correct within several tenths of 1% by chemical analysis. The alloys contain Y_2O_3 additions of 1% (nominal yttria addition is indicated as volume per cent) in all cases, apart from alloy FAC-4, which had a reduced Y_2O_3 content of 0.3%. Alloys were consolidated by Hot Isostatic Pressing at 1100 °C and extruded to round bar at the same temperature.

Samples for tensile creep testing were machined from the extruded bars along a direction perpendicular to the extrusion direction with the central part of diameter 3 mm and length 20 mm polished using 1000 grade SiC. Creep testing was carried out at constant stress in a machine equipped with an Andrade lever arm. Test samples were heated in the creep furnace and temperature stabilised within about 1 h before loading. Creep strain and strain rate were monitored continuously by a data acquisition system, with the test was normally interrupted before failure. Thermal stability of the furnace system, as well as strain resolution, meant that strain rates were subject to larger errors. After the chosen creep strain, samples were cooled rapidly under load, opening and removing the split furnace, such that the sample cooled below about 400 °C (where dislocation recovery becomes slow) within 2–3 min.

Microstructures of as-received materials as well as after deformation were examined by transmission electron microscopy (TEM) on thin foils prepared by electropolishing using a 20% nitric acid in methanol mixture at -30 °C and about 16 V. General studies of particles, as well as examinations of dislocation structures, were carried out using a JEOL 2010 instrument using both Bright Field (BF) and Weak Beam (WB) imaging modes. Detailed analyses of particle chemistry and crystal structure were carried out using JEOL 2100 and JEOL 3000F instruments operating in both TEM and scanning transmission electron microscope (STEM) modes, using also high angle annular dark field (HAADF) and high resolution electron microscope (HREM) imaging.

3. Results

3.1. Initial microstructures

The microstructures of as-prepared materials have been discussed before [3,4,11,22] and only a few pertinent details are given here. All materials had uniform, equiaxed grains of size about 1 μ m in the section perpendicular to the extrusion direction, with a (110) fibre texture along the extrusion direction, moderately strong for the binary alloy (FAC-1) and weak for the ternary alloys. Grain interiors had a high density of fine oxide particles, see Fig. 1, and a very low dislocation density. Grain boundaries were also decorated with a small number of coarser particles [3,4,11,22]. The particles were identified [22] as Al–Y oxides, spherical in the binary alloy and disk shaped in the ternary alloys. Disk particles were identified as generally having an fcc Y₂O₃ structure, with some having the monoclinic Al₂Y₄O₉ structure [22], with typical disk thickness about 5 nm and diameter about 25 nm, such that for all materials the equivalent spherical diameter was about 15–20 nm.

The bright field micrograph of Fig. 1a shows fine disk particles in alloy FAC-2, which are seen end-on along the $\langle 100 \rangle$ directions since they lie on the {001} planes of the B2 ordered matrix. One such particle, in alloy FAC-3, is shown in Fig. 1b. This HREM photograph shows the atomic planes in the (001) oriented B2 matrix (outer part of micrograph) and in the (1–10) oriented fcc particle (inner rectangle). The disk interfaces (seen light) are atomically smooth along the two circular interfaces and the disk circumference. There is a clear orientation relationship between matrix and particle, with most atomic planes being continuous



Crystal

Fig. 1. (a) Bright field TEM micrograph showing high density of oxide particle in alloy FAC-2. The B2 matrix has the (001) orientation and disk oxide particles lying on {100} planes are seen end-on. (b) HREM image showing part of one oxide particle (central rectangle) in the B2 ordered matrix (outer part) of alloy FAC-3. Various planes are marked in this image, as well as interface dislocations (D): see text for details.

across the interface. In some cases the planes are parallel, as for the $(110)_{fcc}$ planes and the $(100)_{B2}$ planes indicated, while in other cases there is a small rotation between planes, as for the $(111)_{fcc}$ planes and the $(110)_{B2}$ planes indicated. The lattice parameters of fcc and B2 phases are slightly different, which means that interface dislocations are required to accommodate this misfit - these are indicated as "D" in Fig. 1b. The misfit between the two phases may be determined experimentally from either the difference of lattice parameters of the two phases, estimated from the fast Fourier transform (FFT) patterns of the lattice images of these phases, or from the density of interface dislocations. In view of the imprecision of the FFT patterns, obtained from a small number of lattice planes for the particles, it seems better to use the density of interface dislocations as a measure of interface misfit. Taking simply the reciprocal of interface dislocation spacing, the misfit is deduced as proportional to 0.7-0.8-0.5 nm⁻¹ for alloys FAC-2, FAC-3 and FAC-4. For the binary alloy FAC-1, the misfit will be larger since no real coherence was observed at the particle-matrix interface. As such, particles are considered to be poorly misfitting in alloy FAC-1, well partially-coherent in alloys FAC-2 and FAC-3, Download English Version:

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