CREEP IN STOICHIOMETRIC BETA-NiAl

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Modes of Deformation

The slip modes and mechanical properties of β -NiAl have been studied by several workers (1-5) but less attention has been devoted to its high temperature creep behavior. From the practical standpoint the latter is of prime interest since the behavior of an alloy during prolonged service at elevated temperatures cannot be predicted with certitude from short duration tests. A summary of the creep properties of β -NiAl is given by Strutt & Dodd (6), and some more recent results are summarized in this paper.

The orientation dependence of the creep strength of β -NiAl single crystals has been shown (6). Fig. 1 shows creep curves for single crystal compression specimens with [001] and [101] axes crept at 850°C, the higher creep strength of the [001] crystals is clearly evident. The [101] crystals deformed on the (001)[100] and (100)[001] glide systems, and the macroscopic distortion of the specimens was uniform without evidence of kinking. In contrast it was found by Hocking, Strutt & Dodd (7) that in compression creep tests on crystals within the standard stereographic triangle slip occurred on only one of the cube systems, and the specimens were kinked severely. Single slip was also noted in tensile tests by Wasilewski, Butler & Hanlon (1), and Kanne, Strutt & Dodd (8) suggested the operative cube system to be that with the greatest resolved shear stress on the {110}<100> cross slip system. Consequently [101] crystals would be expected to deform by uniform duplex slip since (011)[100] & (0 $\bar{1}$ 1)[100] the cross slip systems for (001)[100], and (110)[001] & (1 $\bar{1}$ 0)[001] the cross slip systems (100)[001] are equally stressed. The [001] orientation crystals deformed uniformly (without kinking) on the (011)[011] & (0 $\bar{1}$ 1)[011] glide systems; macroscopic slip on the equally favored (101)[$\bar{1}$ 01] & ($\bar{1}$ 01)[101] was not observed. Thin film specimens were made parallel to the dodecahedral planes and groups of [011] edge dislocations were observed, see Fig. 3e. Electron microscope evidence for {110}<110> glide in NiAl creep specimens was also reported by Strutt, Kanne, & Bevk (9).

An intriguing effect found by Yang & Dodd (10) in polycrystalline NiAl was a precipitous increase in creep strength when the temperature increased from 795°C to 800°C; this is seen by examining Fig. 2. Above 800°C the creep strength then decreases with increasing temperature. A similar effect has also been found in AuZn single crystals presumably it arises from structural lattice defects. Considering Fig. 2, the progressively diminishing value of n in the creep relation $\dot{\gamma} = A\tau^n \exp - \Delta H/kT$ with increasing temperature follows the same trend as found by Vandervoort, Mukherjee & Dorn (11).

Electron Microscopy

Thin-film specimens sectioned from deformed crystals parallel to {100}, {101}, & {111} were examined in the Philips 300 electron microscope using the goniometer stage; with {111} foils it was possible to employ 12 major Bragg reflections. As well as determining b uniquely by the $\bar{g} \cdot b = 0$ criterion, with consideration of the $b \times u$ edge

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FIG. 1

Creep curves for β -NiAl single crystal compression specimens with [101] & [001] axes.

FIG. 2

Creep data for polycrystalline β -NiAl compression specimens.

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FIG. 3

(a) A tilt boundary in (001) with edge dislocations along [100] and of Burgers vector $a_0[010]$, Mag.x15,000. (b) A network formed by reaction of $a_0[010]$ and $a_0[001]$ screw dislocations to form $a_0[011]$ screw dislocation segments, Mag.x30,000. (c) A twist boundary in (100) with $a_0[001]$, and $a_0[010]$ screw dislocations, Mag.x15,000. (d) A group of $a_0[001]$ edge dislocations, Mag.x15,000. (e) Edge dislocations of the (011)[011] glide system, Mag.x15,000. In all cases $\vec{g} = [0\bar{1}1]$.

dislocation displacement, image profiles were computed. The Howie & Whelan equations were integrated using the I.B.M. 360 double precision, adjustable integration step Runge-Kutta sub-routine. The displacement field around the general dislocation in an anisotropic media was computed from the equations of Eshelby, Read & Shockley (12). For the $\langle 111 \rangle$ screw dislocation exact agreement was found with the displacement field calculated by Hirth & Gehlan (13). It should be noted that the effect of elastic anisotropy does not invalidate the $\bar{g} \cdot \vec{b} = 0$ criterion for dislocations lying along an evenfold symmetry axis; e.g. $\langle 100 \rangle$ & $\langle 101 \rangle$ screw dislocations as determined in Figs. 3b & 3c. Examples of some of the electron microscope results are described briefly.

(1). Fig. 3a shows a tilt boundary in which the dislocations lie in (001) and along [100] and have $a_0[010]$ Burgers vector. A twist boundary in (100) with $a_0[001]$ and $a_0[010]$ screw dislocations is shown in Fig. 3c.

(2). Arrays of edge dislocations were frequently seen within the sub-grains but screw dislocations were only seen in sub-boundaries or in networks. Fig. 3e shows edge dislocation of the (101)[101] glide system and Fig. 3d edge dislocations of the (001)[100] glide system.

(3). A network formed by the reaction of $a_0[010]$, and $a_0[001]$ screw dislocations to produce $a_0[011]$ screw dislocation segments is shown in Fig. 3b; the segments that are invisible with the operative $\bar{g}$ vector have a $a_0[011]$ Burgers vector. Another example of the reaction $a_0[010] + a_0[001] \rightarrow a_0[011]$ is evident in the lower part of the boundary in Fig. 3c, where the $a_0[011]$ segments are invisible. This reaction and that shown in Fig. 4 have been observed in extruded NiAl by Lloyd & Loretto (14). For screw dislocations it is highly favorable since if elastic anisotropy is considered there is an 81% reduction.

(4). The reaction of three $a_0\langle 100 \rangle$ dislocations to produce a $a_0\langle 111 \rangle$ dislocation was observed several times in thin foil specimens from a crystal crept at 675°C whose axis was near [213]. The resolved shear stress on the (001)[100], (001)[010], and (100)[010] systems was 1.35, 0.96, and 0.60 kg. mm.⁻². Four micrographs used in the Burgers vector analysis are shown in Fig. 4. Altogether 12 two-beam conditions were employed to uniquely determine the Burgers vectors of all the dislocation segments; $\bar{s}$ was slightly positive in all cases. The dislocation segments AB & BC with Burgers vectors $a_0[010]$ and $a_0[100]$ react to form BD with Burgers vector $a_0[110]$. Segment DE of Burgers vector $a_0[001]$ reacts with BD to produce DF of Burgers vector $a_0[111]$. Segments FH & FG have Burgers vectors of $a_0[100]$ and $a_0[110]$. The straight segment DF lies along $[1\bar{2}1]$ which is parallel to the foil surface whose normal is [111]. Fig. 4b shows DF in good contrast with $\bar{g} = [002]$, and Figs. 4a, c and d show very faint to weak contrast with $\bar{g} = [10\bar{1}]$, $[110]$, and $[011]$. The strongest contrast is for $\bar{g} = [10\bar{1}]$, where $\bar{g}$ is parallel to $\vec{b} \times \vec{u}$ and the corresponding image profile is shown in Fig. 5b, which predicts a double image. The image profiles for the $\bar{g} = [10\bar{1}]$, & $[110]$ reflections are shown in Fig. 5a and are identical and predict a rather faint double image. In the actual electron micrographs, Figs. 4c & d, the image is seen to be extremely weak. It would thus appear that elastic anisotropy has a negligible effect on the diffraction contrast. However Fig. 6 shows two profiles in which appreciable contrast arises entirely from the effect of elastic anisotropy for the [111] edge dislocation. In Fig. 6a the foil normal is [111], and in Fig. 6b the foil normal is $[101]$; the same $\bar{g}$ -vector is used in each case, namely $[1\bar{2}1]$.

In examining the reaction $a_0[100] + a_0[011] \rightarrow a_0[111]$ where the dislocations lie along $[1\bar{2}1]$ the self energies of the $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ dislocations are 8.20×10^{-4} , 4.58×10^{-4} , and 29.30×10^{-4} ergs. cm.⁻². If the $a_0\langle 111 \rangle$ dislocation dissociates into two $a_0/2\langle 111 \rangle$ dislocations separated by an A.P.B., the total energy of this arrangement is $E^* = 2E_1 + w_f$, where E_1 is the self energy of each $a_0/2\langle 111 \rangle$ dislocation, E_{12} the interaction energy, and w_f the energy of the fault. By a simple calculation similar

FIG. 4

A network in which $a_0\langle 100 \rangle$ dislocations have reacted together to form dislocation segments with $a_0\langle 110 \rangle$, & $a_0\langle 111 \rangle$ Burgers vectors. Mag.x50,000.

FIG. 5

Image profiles for the $a_0\langle 111 \rangle$ edge dislocation such as DF in Fig. 4.

FIG. 6

Image profiles for the $a_0\langle 111 \rangle$ edge dislocation, for detail see text.

to that of Ball and Smallman (4) E^* is found to be 22.88×10^{-4} ergs. cm.⁻², thus $E_{\langle 100 \rangle} + E_{\langle 110 \rangle} \approx E^*$, and there is no energy reduction. It would be of interest to determine the favorability of the reaction at high temperatures, however only the temperature variation of C_{11} is known (15). Even though the reaction does not involve an energy reduction, it may occur since it involves a diminution of the total length of dislocation in a network.

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