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PROGRESS REPORT NO. 5

For Nickel-Aluminum Alloys
January 1969 - January 1970

MASTER

During the present reporting period the work on CoAl and beta-brass was completed and presented for publication (see publication list and also enclosed pre-prints for details). The NiAl research is nearing completion and is discussed below.

1. Orientation-Dependence of Creep Deformation in Stoichiometric NiAl Single Crystals (J. Bevk)

Single crystal tensile specimens were cut from coarse-grain polycrystalline ingots to the tensile axis orientations shown below, and creep tests have been completed.

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An anticipated, the combined data for the various orientations does not conform to any of the well-known creep equations on account of the inconstancy of the slip systems. For instance, the major deformation made of orientation (a) is $(011)[0\bar{1}1]$ although there is strong evidence for a $[111]$ Burgers vector also. The typical structure after creep is shown in Figure 1 and this contains what appear to be superlattice dislocations as part of the network since a single reflection was operative. Further careful electron microscopy is now in progress using a newly-installed EM 802 microscope with dark-field facility. This will permit

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FIGURE 1

Composite Micrograph of Typical Creep Structure in NiAl Crystal of Orientation (a). Burgers vectors of $\langle 110 \rangle$, $\langle 100 \rangle$, and possibly $\langle 111 \rangle$ type.

FIGURE 2

Orthogonal Dislocation Sets in Creep Specimen
of Orientation (d). $b = [100]$

unambiguous Burgers vector determinations which has been very difficult up to this time. It will be noted that the Schmidt factor on the cube system for orientation (a) is quite small. Nevertheless, this is not a particularly "hard" orientation due to easier operation of the $[011](0\bar{1}1)$ system than previously supposed.

Orientations (b), (c), (d) and (e) were cut to give a constant Schmidt factor on the cube system but a variable one on the cross slip system. However, both the creep and electron microscope data indicate that these orientations behave quite similarly. This also is contrary to earlier opinions that the intrinsic drag stress is markedly higher on the dodecahedral than on the cube plane. A typical structure is shown in Figure 2.

Finally, orientation (f) was cut to give duplex slip. Electron microscopy of these specimens has not yet been completed.

2. Creep Mechanisms in Vacancy-Defective NiAl (W. Yang)

As noted in the previous progress report it was necessary to study aluminum-rich NiAl using polycrystalline compression samples. A new improved compression creep machine has been constructed for this work, incorporating more sensitive strain measurement arrangements, and a large number of creep curves have been determined and analyzed. Some of the resulting data is given in Figures 3-6 and summarized below in tabular manner.

Alloy a/o Al	ΔH , Apparent Activation Energy	n, Stress Exponent
50	<ol style="list-style-type: none"> 1. ΔH is stress-sensitive in the range 800-1000°C. 2. In the above temperature range ΔH varies from 62 k. cal. mole⁻¹ at 3000 psi to 42 k. cal. mole⁻¹ at 6000 psi. 3. From 588°C to 795°C ΔH is constant at 60 k. cal. mole⁻¹. 	<ol style="list-style-type: none"> 1. n decreases as the temperature increases, viz from 10.3 at 800°C to 4.9 at 1000°C. 2. n is approx. constant and equal to 10 between 588 and 795°C.

FIGURE 4

Shear Stress/Shear Strain Relationships at Various Temperatures for Ni/52 at.% Al.

FIGURE 5

Shear Stress/Shear Strain Relationships at Various Temperatures for Ni/54 at.% Al.

Alloy a/o Al	ΔH , Apparent Activation Energy	n, Stress Exponent
52	<ol style="list-style-type: none"> 1. ΔH is again stress-sensitive, varying from 49 k. cal. mole⁻¹ at 1500 psi to 55 k. cal. mole⁻¹ at 6000 psi. 2. The stress-sensitivity is less and opposite in sense to that of the stoichiometric alloy. 	<ol style="list-style-type: none"> 1. n increases slightly as the temperature increases, viz from 3.1 at 850°C to 3.9 at 1000°C. 2. n is constant and equal to 3.1 between 750 and 850°C.
54	ΔH is stress-independent and equal to 67k. cal. mole ⁻¹ .	n is constant at 3.4

In addition to the tabulated data, various additional observations can be listed.

(a) Aluminum-rich NiAl creeps only above ~750°C. At lower temperatures the alloys crack, so it is tentatively concluded that diffusion is a necessary part of the creep mechanism in these alloys unless a strengthening effect at lower temperatures inhibits slip. Such strengthening could take the form of grain boundary hardening by oxygen and for nitrogen, but this is not favored since it should also occur in stoichiometric NiAl. Alternatively, the structural vacancies may act as potent hardeners at low temperatures, particularly if they condense and associate with the excess aluminum atoms to form stacking-fault types of defects as suggested by Ball. Earlier work (Item 1 in publication list) has suggested that low-temperature creep in stoichiometric NiAl is controlled by thermally-activated cross-slip. This suggests that cross-slip may be inhibited in aluminum-rich NiAl, but the comparatively small amount of electron microscopy completed so far does not permit definite conclusions on such points.

(b) Figure 6 shows the stresses necessary to achieve a given steady-state creep rate in alloys of different composition. It is seen that at temperatures $>850^{\circ}\text{C}$ stoichiometric NiAl is stronger than vacancy-defective NiAl, the difference being more marked the higher the temperature. This correlates well with the higher diffusivities in the non-stoichiometric alloy. At 750°C the strengths are reversed - as is clear from the observations in (a) above. However, an additional point not covered in (a) is that stoichiometric NiAl behaves anomalously in the sense that it has less creep resistance at 750°C than at 800 or even at 850°C . This point has been well verified, but must clearly be considered in conjunction with the stress and temperature-sensitivity of ΔH and n for stoichiometric NiAl.

3. Operative Slip Systems and Creep in NiGa (D. Ku)

It was noted in the previous progress report that size factor considerations suggest the possibility of a $[111]$ Burgers vector in NiGa. Also, this probability is enhanced by the lower melting temperature of this compound compared to NiAl, suggesting less strong ordering and a lower anti-phase domain energy.

During the summer of 1969 a high-temperature (1800°C) graphite-element furnace was acquired for growing single crystals of high-melting-point alloys. This has been used to grow large cylindrical single crystals of NiGa, some of which have been electromachined into tensile creep specimens, and some cut into smaller compression specimens. The former are now being creep tested with the principal aim of determining operative slip systems, while the latter will be used for activation energy measurements, etc.

4. Operative Slip Systems and Creep in FeAl, etc.

Coarse grained ingots of FeAl have been prepared for sectioning into single crystals, but the research has not proceeded beyond this point owing to lack of manpower (the three students supported by AEC

are working on 1-3 above). However, the AEC-sponsored research has benefited during the past year through the part-time assistance of a post-doctoral fellow supported by the Graduate School and the Engineering Experiment Station of the University of Wisconsin. Briefly, large single crystals of AuZn and AuMg have been prepared and shaped into both tensile and compression creep specimens. Most of the compression tests have been completed, and the tensile tests will shortly be run to assist in determining the slip systems. It will be recollected that the previous progress report listed these CsCl-type compounds as most interesting from a size factor standpoint. Consequently this work is very valuable to an overall understanding of slip in CsCl phases which is the basic aim of the research.