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The Structure and Properties of Grain Boundaries in B2 Ordered Alloys

A Progress Report

by

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ABSTRACT

This report covers the period from April 1st 1989 to March 30th 1990. During this period further extrusions of fine-grained alloys of FeAl and NiAl with different aluminum contents were produced. Thus far, researchers have

1) completed room temperature mechanical testing which shows that the yield strength, σ_y, of FeAl shows a minimum near Fe-45Al while that for NiAl increases with increasing (nickel-rich) deviation from stoichiometry;
2) showed that the grain boundary strengthening parameter. k, in the Hall-Petch relationship, $\sigma_y = \sigma_0 + k\, d^{-1/2}$ (where σ_0 is the lattice resistance and d the grain size), in FeAl shows a minimum at Fe-45Al whilst in NiAl it decreases with increasing deviation from stoichiometry;
3) studied grain boundary chemistry in near-stoichiometric NiAl and FeAl using scanning Auger electron spectroscopy at Oak Ridge National Laboratory, with the aid of Mr. R.A. Padgett, Jr;
4) studied grain boundary chemistry using EDS and Z-contrast imaging on field emission gun high resolution scanning transmission electron microscopes;
5) continued in-situ straining experiments in the transmission electron microscope to examine dislocation/grain boundary interactions;
6) examined dislocation structures in lightly-deformed NiAl and FeAl and, based on observations in the latter, proposed a mechanism for transgranular fracture;
7) completed work which examines the effect of cooling rate following annealing on the strength and hardness of FeAl.

INTRODUCTION

A study, funded by the U.S. Department of Energy (grant no. DE-FG02-87ER45311) was initiated on September 1st 1987 to examine the structure and properties of grain boundaries in the B2 ordered alloys FeAl and NiAl. The project is aimed at increasing the understanding of grain boundaries in ordered alloys as a whole. These two compounds were chosen as model materials largely because they exist over a wide range of compositions and can thus be expected to show significant variations in structure and properties over their composition range.

The effect of alloy composition on the structure/chemistry and properties of grain boundaries is being examined:

1) by measuring yield strength, σ_y, as a function of grain size, d, and, thus, determining the constant k in the Hall-Petch relationship, $\sigma_y = \sigma_0 + k\, d^{-1/2}$ (where σ_0 is the lattice resistance),

2) by measuring grain boundary composition using, a scanning Auger electron microprobe (SAM) and high resolution field emission gun (FEG) scanning transmission electron microscopes (STEM) at the Oak Ridge National Laboratory and at Bethlehem Steel;

3) by examining dislocations in the grain boundaries of lightly-strained compression samples using a transmission electron microscope (TEM);

4) by examining dislocation/grain boundary interactions during in-situ straining in a TEM;

5) by measuring grain growth kinetics.

The progress in each of the above areas is reviewed overleaf. Dislocations within the grains of lightly-strained FeAl and NiAl have also been examined using TEM. In addition, work on the effect of cooling rate on the strength of FeAl and NiAl is described. Subsidiary research on the B2-structured alloys PdIn, NiAl+V, Mn or Cr and Fe_3Al+Cr is also described.

AUGER ELECTRON SPECTROSCOPY

Auger electron spectroscopy (AES) studies of NiAl and FeAl were continued using the PHI660 SAM at the High Temperature Materials Laboratory (at ORNL) with the assistance of Mr. R.A. Padgett, Jr. It was found possible to fracture intergranularly, at least partially, all NiAl and FeAl compositions at 77K if the grain size was large (>50µm). The data shown below are for Fe-34Al, the most Fe-rich composition and, thus, the composition thought most likely to show compositional differences at the grain boundary. The results show that there are no clear compositional differences between transgranular areas and intergranular facets. Note that there is quite a wide range of values, even for transgranular features where no compositional differences might be expected. These variations may be a consequence of different angles of the fracture surfaces with respect to the analyzer, as recently suggested by Rothman and Menyhard (Scripta Metall. **23** (1989) 1169).

Table 1.

Compositions measured from the fracture surfaces of *in-situ* fractured Fe-34Al.
The values on the same line are adjacent intergranular and transgranular areas.

Intergranular		Transgranular	
At.% Fe	At.%Al	At.% Fe	At.% Al
81.1	18.9	68.9	31.1
67.8	32.2	65.0	35.0
68.6	31.4	73.4	26.6
66.0	34.0	67.8	32.2
66.5	33.5	68.2	31.8
72.2	27.8	68.0	32.0
62.9	37.1	64.0	36.0
61.7	38.3	73.1	26.9

Although AES could not find any compositional differences between intergranular facets and transgranular facets, two unexpected phenomena were observed.

First, it was found that oxygen adsorption onto *in-situ* fractured surface was quite rapid and that the oxygen content rose to high levels (up to 20%), see Figure 1. This adsorbed oxygen changed the measured Al/Fe ratio.

Second, changing the area of scan for a constant beam current and, hence, increasing the current density, on a sample with considerable adsorbed oxygen led to a change in the measured oxygen concentration, see Figure 2.

Increasing the electron probe current density appears to have no effect until a critical current density is reached. Above this point increasing the current density decreases the measured oxygen concentration. At high current densities the oxygen concentration again appears to be independent of current density, suggesting that an equilibrium has been reached between oxygen being adsorbed and oxygen being driven off the surface.

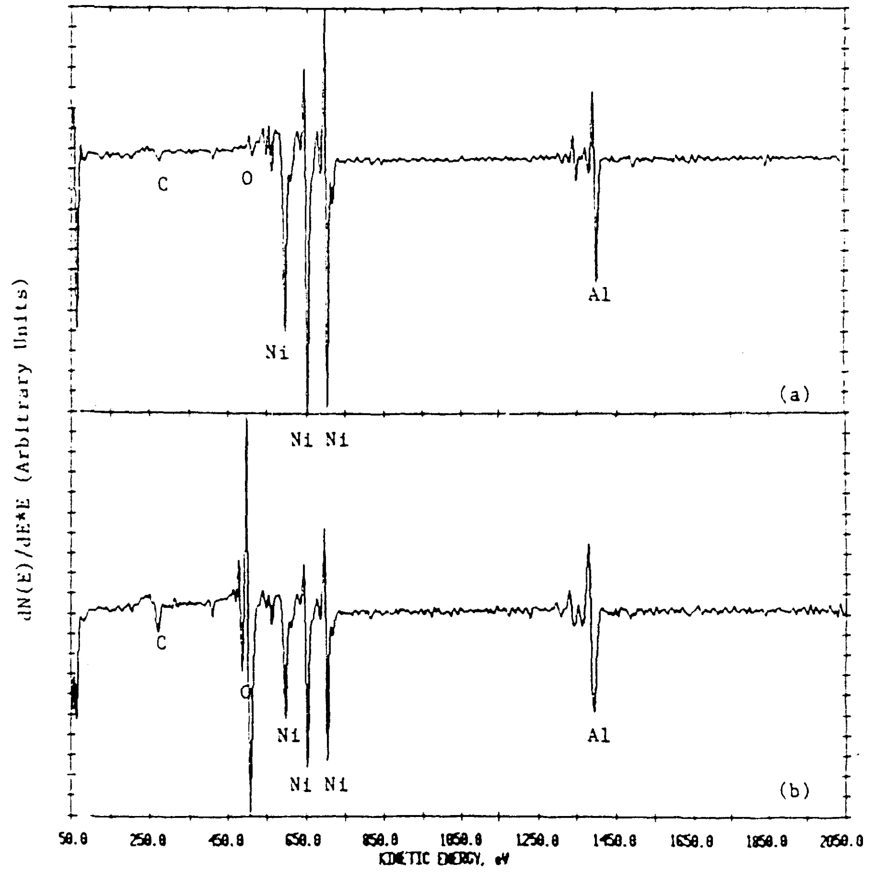

Figure 1. Auger electron spectra from the fracture surface of unalloyed FeAl; a) immediately after fracture and b) after leaving the sample overnight.

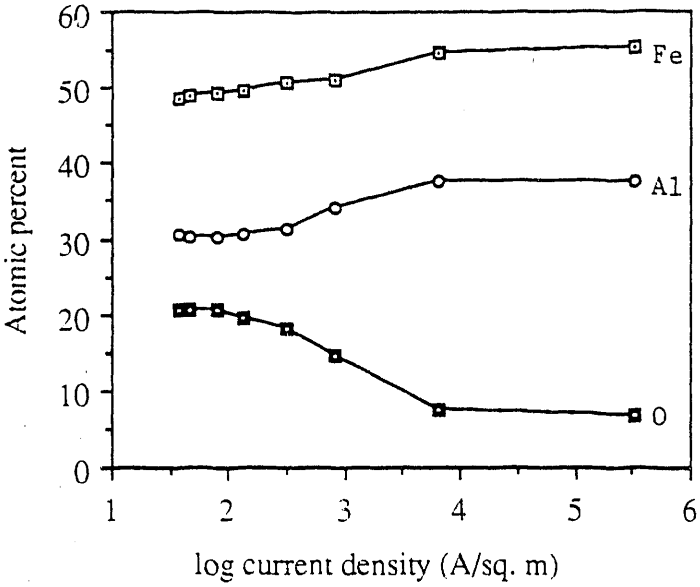

Figure 2. Calculated Al, Fe and O concentrations from an oxygen-'saturated' transgranular facet of Fe-34Al as a function of current density in the electron probe.

Drs. George and Liu at ORNL have also observed the above effects on both NiAl and FeAl and a joint paper is being prepared on this subject. The effects of oxygen pick-up and its variation with beam current density have important implications for the determination of surface chemistries because the oxygen can shield other elements, in particular light elements such as boron.

OTHER GRAIN BOUNDARY CHEMISTRY/STRUCTURE EXAMINATIONS

In addition to AES, several other methods were employed to examine grain boundary chemistry. Energy dispersive x-ray microanalysis was used on a high resolution FEG STEM, a Vacuum Generators HB501, to examine the grain boundaries of Fe-40Al at Bethlehem Steel, with the assistance of Dr. J.R. Michael. No difference in composition between the grain boundary and matrix was observed. Z-contrast imaging was also used to examine the composition of the grain boundaries of Fe-40Al using a HB501 at ORNL, with the assistance of Dr. S.J. Pennycook. Again, no compositional differences were observed between grain boundary and matrix.

Pavan Nagpal, Ph.D. candidate, spent a month at ORNL working with Drs. M.K. Miller and J.A. Horton and (partially supported by the ORAU-SHaRE program) attempting to examine grain boundaries in NiAl using field ion microscopy (FIM). This was unsuccessful since no grain boundaries were found in the FIM tips, probably because the grain size of the material was too large at ~10µm.

To conclude, using three techniques (AES, and EDS and Z-contrast imaging on a STEM), there is no clear evidence that grain boundaries in NiAl and FeAl have compositions different from the matrix.

It is worth noting that TEM examination of the grain boundaries of lightly-deformed FeAl could not determine any obvious differences in the structure of extrinsic dislocations. In other words, the APB-coupled dislocations could not be resolved in the grain boundary, indicating little reduction in order there.

IN-SITU STRAINING EXPERIMENTS

In-situ straining experiments in the TEM have been hampered by two problems. First, the brittleness of the materials concerned has caused severe difficulties in specimen preparation and, thus far, only the most ductile compositions (Fe-34Al and Fe-40Al) have been made into *in-situ* tensile samples. Second, the gearbox on the *in-situ* straining stage broke and this took time for JEOL to repair. *In-situ* straining experiments are now underway again.

COOLING RATE EFFECTS AFTER ANNEALING

The study of the effect of cooling rate after annealing on the mechanical properties of FeAl was continued and widened to include an examination of cooling rate effects on NiAl.

Five millimeter thick discs were annealed in air for 30 mins. at 1273K for FeAl and 1573K for NiAl and then either air-cooled, furnace-cooled (at 50K/hr or 100K/hr) or water-quenched. After the anneals, specimens were dislocation-free and had large grain sizes (~ 200µm for NiAl and ~400µm for FeAl). Subsequently, some of the furnace-cooled FeAl samples were annealed at

673K for either 118hr or 240hr and the air-cooled NiAl samples were annealed at 673K for 45hr and then 923K for 170hr. (Annealing at 673K leads to the formation of a DO_3 structure in Fe-34Al.)

For each alloy and heat-treatment the Vickers microhardness was measured (300g load) from the centers of grains on samples polished through 600 grit SiC paper. The results are shown in Figures 3 and 4 where each microhardness value is the average of ten measurements, the error bars having been omitted for clarity. For FeAl, data is shown for 118hr at 673K only since data after 240hr anneals were identical. Similarly, the microhardness values after annealing NiAl at 673K were identical to those after annealing at 923K so that only the 673K values are shown.

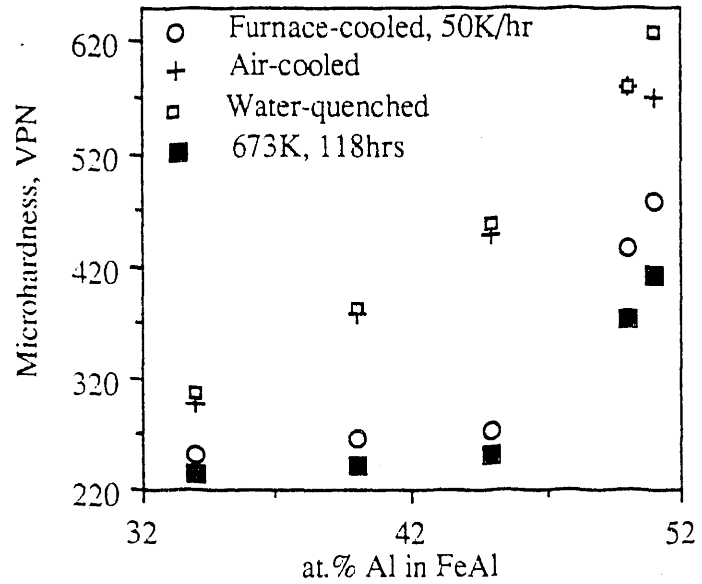

Figure 3. Microhardness as a function of composition for FeAl after various heat treatments following annealing for 30 mins. at 1273K.

Figure 3 shows the microhardness versus aluminum concentration for various cooling rates in FeAl. The microhardness clearly depends strongly on the heat-treatment. Several points are worth noting. First, on annealing even the slowly cooled (50K/hr) samples at 673K the hardness decreases further, indicating that even a slow cool cannot completely eliminate retained thermal vacancies. This effect is most pronounced for Fe-50Al and Fe-51Al. Second, except for Fe-51Al which has constitutional vacancies on the iron sublattice, the air-cooled samples have similar hardness values to the water-quenched samples. For example, the hardness of Fe-45Al after either water quenching or air cooling is 80% greater than after a low temperature anneal at 673K. Third, in the iron-rich alloys, the difference between furnace-cooled and air-cooled increases with increasing aluminum content, suggesting vacancy migration is more difficult nearer stoichiometry. Fourth, Fe-50Al and Fe-51Al can retain large concentrations of vacancies even after slow cooling at 50K/hr. This is consistent with the observation that vacancies, as measured by positron annihilation, are retained in the equiatomic alloy even after slow cooling (D. Weber et al., J. de Physique C7, **38** (1977) 332).

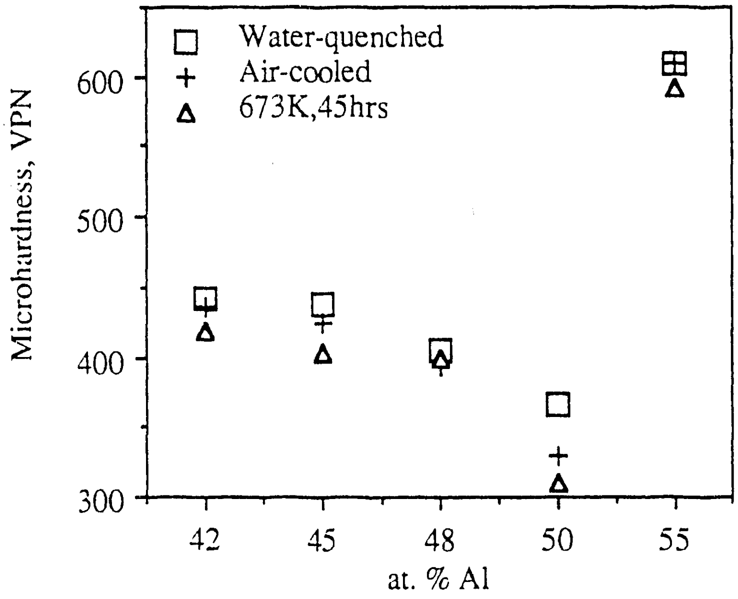

Figure 4. Microhardness as a function of composition for NiAl after various heat treatments following annealing for 30 mins. at 1473K.

Figure 4 shows microhardness as a function of aluminum concentration for various cooling rates in NiAl. Although equiatomic NiAl can retain a large concentration of vacancies (A. Ball and R.E. Smallman, Acta Metall. **16** (1968) 233), the difference in microhardness between water-quenched samples and samples given a low temperature anneal is much less than in FeAl. Stoichiometric NiAl showed the largest increase in hardness due to retained vacancies from water-quenching, however, even in this composition the hardness increase due to air cooling was small compared to FeAl

Thus, it has been shown that differences in cooling rate can have dramatic effects on the hardness of FeAl but much lesser effects on the hardness of NiAl. The present study shows that even after slowly (furnace) cooling FeAl, retained vacancies can increase the room temperature hardness and presumably, yield strength. The implication of these results is that previous reports of the low temperature strength of FeAl need to be treated with caution, since different cooling rates can produce changes in strength of up to 80% which may be more than the grain size strengthening effects in this material.

YIELD STRENGTH MEASUREMENTS

It is clear that to measure room temperature properties that are not influenced by thermal vacancy retention FeAl has to be annealed at low temperature, ~673K, for around 100hrs. This anneal was performed prior to mechanical testing of all FeAl except Fe-34Al which changes to the DO_3 structure after 673K. The measurement of yield stress as a function of grain size, d, was completed: figs 5 and 6 show plots of yield stress versus $d^{-1/2}$ for FeAl and NiAl. Note plots of yield stress versus $d^{-0.8}$, the relationship used by Schulson et al. for Ni_3Al and Ni_3Si (Acta Metall., **34** (1986) 1395; Acta Metall **38** (1990) 207), did not result in straight lines.

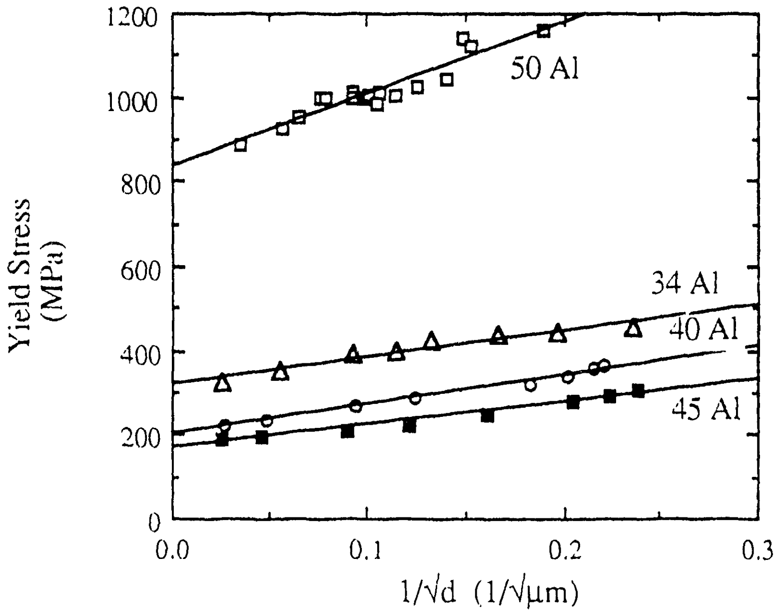

Figure 5. σ_y vs. $d^{-1/2}$ for FeAl.

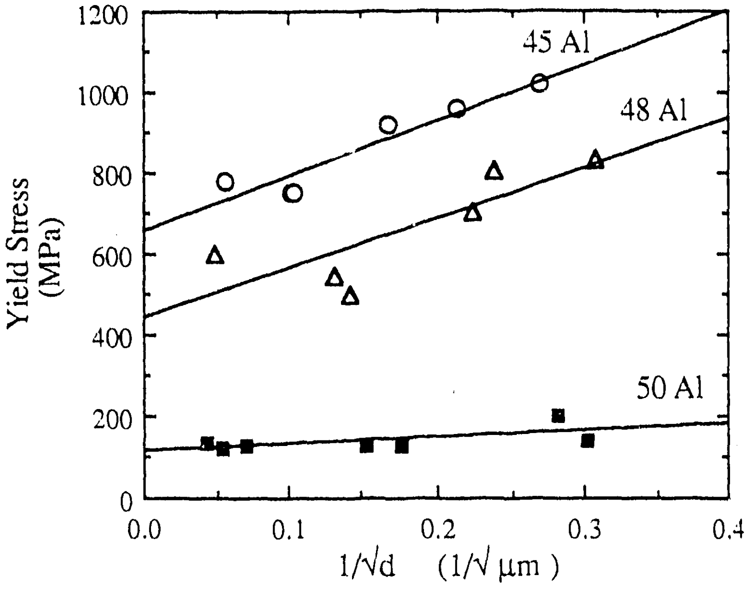

Figure 6. σ_y vs. $d^{-1/2}$ for NiAl.

The variation of k as a function of composition for FeAl and NiAl is shown in Figure 7.

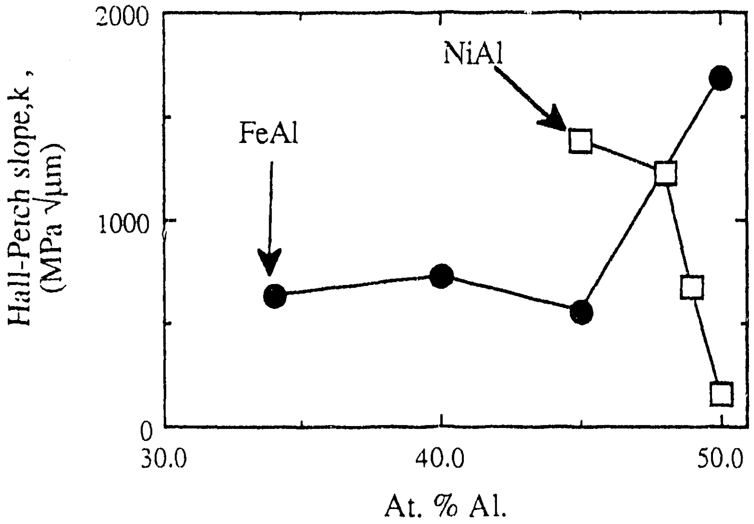

Figure 7. The Hall-Petch slope, k, as a function of aluminum concentration for FeAl and NiAl. The data for Ni-49Al is from E.M. Schulson and D.R. Barker, Scripta Metall. **17** (1983) 519.

For FeAl, which deforms at room temperature by the motion of APB-coupled a/2 <111> dislocations, k decreases abruptly on deviating from the stoichiometric composition but changes little thereafter. (The rapid change of a property for small deviations from stoichiometry but smaller changes in that property for further deviation from stoichiometry is a feature of many properties of B2 compounds.) Optical metallographic examination of polished surfaces of large-grained FeAl after deformation show wavy slip at the stoichiometric composition (where APB-coupled partials are very closely spaced due to the high APB energy) but planar slip away from stoichiometry (where the APB energy is lower and hence APB-coupled partials are more widely spaced and more difficult to re-associate and cross-slip). Slip was found by TEM to be by <111> dislocations.

For NiAl, which was shown to deform by the slip of perfect a<100> dislocations, k is very small at the stoichiometric composition and increases with increasing nickel-rich deviation from stoichiometry, see Figure 7. Interestingly, the magnitude of this increase in k is similar to the decrease in k for FeAl over the same composition range. Also, the ductilities of FeAl and NiAl correlate with the change in k, *viz.*, stoichiometric FeAl (where k is large) fractures before macroscopic yielding in tension but the other compositions (in Figure 7) can show tensile ductility, whereas NiAl is slightly ductile in tension when it exhibits a low value of k (stoichiometric composition) but it is brittle when it exhibits larger k values (Ni-rich compositions). It is worth noting that the ductility of stoichiometric NiAl is not simply due to the low yield strength (100-200 MPa) compared to off-stoichiometric compositions (250-1000 MPa) since a related alloy Ni-30Al-20Fe which also deforms by <100> slip shows tensile ductility even though the yield stress is ~800 MPa (S. Guha. P.R. Munroe and I. Baker, Scripta Metall. **23** (1989) 897).

The grain boundaries must be becoming increasingly constitutionally disordered with increasing deviation from the stoichiometric composition for both FeAl and NiAl but k decreases for the former (initially) but increases for the latter. Thus, there is no clear correlation between constitutional disorder and k.

The variation of lattice resistance with at.% Al for FeAl and NiAl is shown in Figure 8.

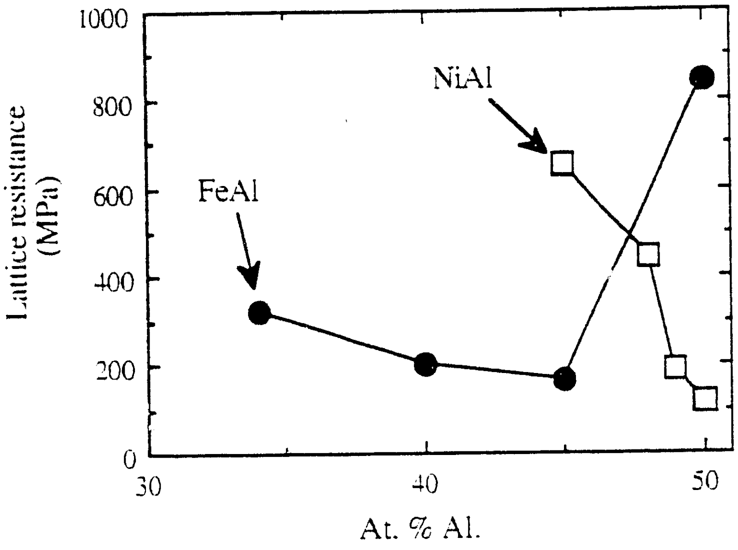

Figure 8. Lattice resistance as a function of aluminum concentration for FeAl and NiAl.

Limited data are also available on the compound AuZn (A. Causey, Ph.D. Thesis, 1968, U. British Columbia) which deforms by <100> slip but may also deform by a stress-induced martensitic transformation in some compositions. (The martensitic transformation could explain why the alloy is ductile even though <100> slip provides only three independent slip systems.) The behavior of k for AuZn is similar to NiAl, *viz.*, k is very small at the stoichiometric composition but increases for off-stoichiometric compositions.

Table 2 summarizes the effect of constitutional disorder on NiAl, FeAl, and AuZn.

Table 2. Effects of constitutional disorder on NiAl, FeAl and AuZn.

		Effect of Increasing disorder on			
Compound	Slip Vector	Planarity of slip	k	σ_0	Ductility
FeAl	<111>	becomes planar	decreases	decreases	increases
NiAl	<100>	no change	increases	increases	decreases
AuZn	<100>	unknown	increases	increases	unknown

A feature of the data for FeAl and NiAl (and to a lesser extent for AuZn) is that an increase in k roughly correlates with an increase in σ_0, see Figures 7 and 8. An increase in σ_0 as k increases has also been noted for copper alloys containing various amounts of aluminum (Y. Higo et al., Met. Sci. **15** (1981) 233).

The slope, k, on the basis of a dislocation pile-up model, can be defined as

$$k = \beta\sigma_n r^{1/2} \tag{1}$$

where σ_n is the shear stress needed to operate a dislocation source in a grain (to which slip is being transmitted) and r is the distance of the source from the grain boundary (through which slip is being transmitted). β is a constant (~1) whose value depends on the details of the model used to derive the equation (1). It seems reasonable that the stress to operate a source is affected similarly to the lattice resistance by solutes, etc., that is, σ_n would be proportional to the lattice resistance, σ_0. In other words

$$k = C\,\sigma_0 r^{1/2} \tag{2}$$

where C is a constant. Thus, we should expect k to correlate with σ_0 for solute strengthening.

The above analysis of k, of course, did not explicitly incorporate any microstructural features such as wavy versus planar slip behavior, or order versus disorder of the grain boundaries. Possibly these features affect k through changes in C or r in equation (2). Although k appears to correlate closely with ductility in NiAl and in FeAl, it is not clear why at the moment.

TEM Examination of Strained FeAl and a Mechanism for Transgranular Fracture

Based on the above work, the fracture behavior of FeAl can be straightforwardly rationalized. Stoichiometric FeAl has grain boundaries which are very resistant to slip (high k). Thus, this composition shows intergranular failure and low ductility, presumably due to the difficulty of slip transmittal across grain boundaries.

At off-stoichiometric compositions transgranular fracture occurs in FeAl[†]. A dislocation mechanism has been identified which may lead to the nucleation of transgranular cracking. This

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[†] k is lower off-stoichiometry and so slip transmittal is possible.

involves the interaction of a pair of gliding APB-coupled a/2 <111> dislocations to produce a pair of near-edge <100> dislocations, a reaction which is energetically favorable.

<100> dislocations are relatively immobile in FeAl at low temperature, therefore, the <100> dislocations could act as a crack nucleus. The mechanism, indicated schematically below, is similar to the Cottrell mechanism for low temperature cleavage fracture in b.c.c. metals. That fracture in FeAl requires considerable (local) plastic flow is evidenced by the fact that selected area channeling patterns (SACP's) could not be obtained on transgranular fracture surfaces of FeAl.

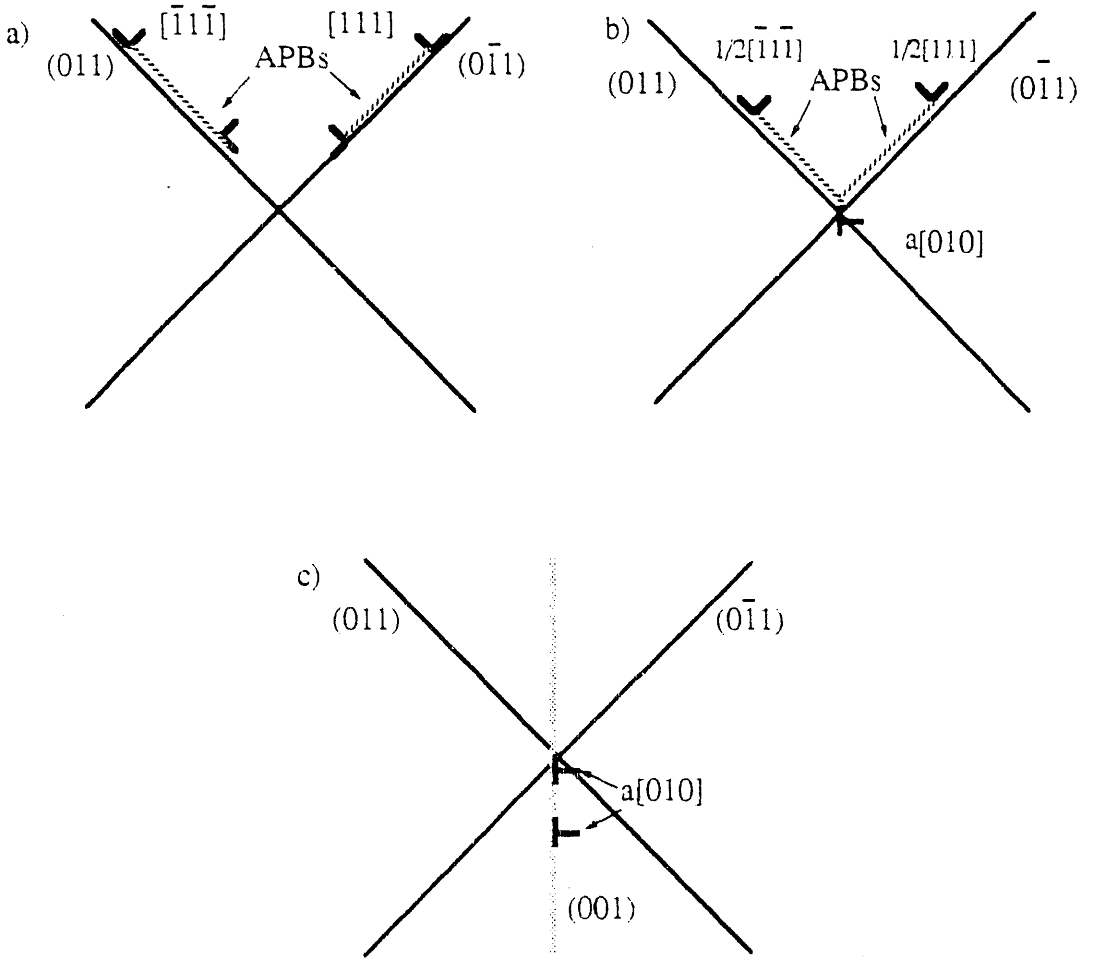

Figure 9. Schematic illustration of the formation of a pair of a<001> dislocations.

a) Two <111> dislocations, each consisting of two a/2<111> dislocations separated by an APB glide on intersecting {011} planes.

b) The leading a/2<111> dislocations interact to form an a<001> dislocation, which is still connected to the trailing a/2<111> partials.

c) The trailing a/2<111> partials interact to form a second a<001> dislocation which repels the a<001> dislocation already present.

The driving force for the formation of <100> dislocations is essentially the removal of the energy associated with the APB (between the APB-coupled a/2 <1 1 1> dislocations). Interestingly, the number of <100> dislocations observed after room temperature deformation decreases with decreasing aluminum concentration which corresponds to a decrease in APB energy (and, hence, the driving force for the reaction) and to an increase tensile ductility. This is the first time, to the author's knowledge, that a possible mechanism for transgranular fracture has been observed in an ordered alloy.

OTHER STUDIES

PdIn

The opportunity arose to study the microstructure, mechanical properties and deformation mode of the B2-structured compound PdIn. The material examined was supplied by Williams Dental Company of Buffalo, NY. PdIn is being considered as a possible alloy for dental prosthetics, largely because of its color which varies with composition from copper-colored to gold-colored.

The alloy, Pd-52at.%In, was examined and tested in the as-cast state. Optical microscopy revealed a single phase structure, with a grain size of about ~100μm, some porosity, presumably related to the as-cast structure, was also noted. Two samples were tested in tension at room temperature. They exhibited yield strengths of 46MPa and 50MPa, fracture strengths of 95MPa and 132MPa, and elongations of 2.0% and 3.6% respectively. Scanning electron microscopy of the fracture surfaces revealed principally intergranular fracture, together with some transgranular cleavage. Voids at triple points were also noted which may have contributed to premature failure.

TEM examination of the dislocation structure from the strained material, revealed the presence of <001>{100} slip only. Since these provide insufficient slip systems to satisfy von Mises criterion and because stress-induced martensitic decomposition in some B2-structured compounds may enhance ductility (A.G. Rozner and R.J. Wasilewski, J. Inst. Metals, **94** (1966)

169; C. Lall, M.H. Loretto and I.R. Harris, Acta Metall., **26** (1978) 1631; D. Hossain, I.R. Harris and G. Barraclough, J. Less-Common Metals, **37** (1974) 35), experiments were performed to induce such a transformation if it occurred. *In-situ* heating and cooling experiments were performed in the TEM, but no evidence of a martensitic transformation was observed. The microstructure showed a 'tweed' structure at all temperatures which was accompanied by streaking and extra reflections in the diffraction patterns, see Figure 10. The origins of the ductile behavior in this compound are unknown, although it is worth noting that other B2 compounds e.g. AuZn, which have not been proved to undergo a martensitic reaction, may exhibit both <001> slip and extensive ductility (A.R. Causey and E. Teightsoonian, Metall. Trans. A, **1A** (1970) 1177)

Figure 11. TEM diffraction pattern from B2 PdIn, note the streaking and extra reflections.

ALCHEMI

ALCHEMI or Atom location by channeling enhanced microanalysis, is a transmission electron microscope technique, developed recently, for the atomic site location of trace impurities in minerals and ternary additions in intermetallic compounds (J.C.H.Spence and J.Tafto, J.Microscopy, **130** (1983) 147; J.Tafto and J.C.H. Spence, Science, **218** (1982) 49; K.M.Krishnan, Mat.Sci.Eng., **B3** (1989) 397). So far, the technique has not been applied to B2-

structured compounds. Studies were performed in two areas: 1) atom site location of ternary additions in near-stoichiometric compounds, and 2) atom site location of ternary additions in non-stoichiometric compounds.

1) ALCHEMI studies were performed on Ni-47Al-3V and Ni-47.5Al-1.5Mn and Ni-43.5Al-5.2Cr. Experimental conditions for the acquiring of data in B2 compounds were established. Qualitative examination of the data indicated that in each case the ternary addition occupied the aluminum sublattice. In each alloy the fraction of the ternary element sitting on the aluminum sublattice was calculated to be close to 100%.

2) ALCHEMI studies were performed on B2-structured alloys Ni-30Al-20Fe, and Fe-28Al-6Cr. In both cases the results from ALCHEMI studies were inconclusive. For Ni-30Al-20Fe. data suggested that both nickel and iron atoms were partitioned between both the 'nickel' and 'aluminum' sublattices. (Note this latter compound also showed a tweed structure in TEM images.) Whilst, ALCHEMI studies on Fe-28Al-6Cr suggested that chromium preferentially occupied the aluminum sublattice. However, in both alloys the presence of these anti-site defects convoluted the data. ALCHEMI was also performed on DO_3-structured Fe-28Al-6Cr, but no unambiguous information regarding the atom site location of the chromium atoms could be obtained.

It was concluded that ALCHEMI was a good technique for near-stoichiometric B2 compositions, but that the accuracy of the data decreases as compositions move away from stoichiometry.

PERSONNEL

The following personnel were supported during this budget period:-

Ian Baker - P.I.
Paul R. Munroe - Research Assistant Professor
Pavan Nagpal - Ph.D. candidate

Pavan Nagpal has completed his Ph.D. qualifying exams and his thesis proposal was approved. He expects to complete his Ph.D. in the Fall of 1990. Fuping Liu worked on this project (until December 31st 1989) as part of his first year doctoral work (prior to starting his thesis project).

PAPERS PUBLISHED OR IN PRESS DURING YEAR -3

1. "On Grain Boundary Disorder and the Tensile Ductility of Polycrystalline Ordered Alloys : a Hypothesis", I. Baker and E.M. Schulson, Scripta Metall., **23** (1989) 345-348.
2. "Observation of <111> Slip in NiAl", P.R. Munroe and I. Baker, Scripta Metall. **23** (1989) 495-499.
3. "Observation of Tweed Microstructures in Ni-Al-Fe Alloys", P.R. Munroe and I. Baker, 47th Proc. EMSA (1989), 676-677.
4. "Annealing Studies of B2 FeAl", B. Schmidt, P. Nagpal and I. Baker, Proc. MRS, 133 (1989), 755-760.
5. [†] "Dislocation Structures in Ordered B2 Alloys", I. Baker, 47th Proc. EMSA (1989), 314-315.
6. "The Microstructure of Extruded FeAl", P.R. Munroe and I. Baker, J. Mat Sci., **24** (1990) 4246-4252.
7. [†] "Properties of B2 Ordered Alloys", I. Baker and P.R. Munroe, Proc. TMS Symp. 'High Temperature Aluminides and Intermetallics', Indianapolis, IN, Oct. 1989 (in press).
8. "Effect of Cooling Rate on Hardness of FeAl and NiAl", P. Nagpal and I. Baker, Met Trans. (in press).

PAPERS SUBMITTED FOR PUBLICATION OR IN PREPARATION DURING YEAR -3

1. "Deformation of B2-Structured PdIn", P.R. Munroe, I. Baker and P. Nagpal, submitted to J. Mat. Sci.
2. "Observation of <001> Dislocations and a Mechanism for Transgranular Fracture in B2-structured FeAl", P.R. Munroe and I. Baker, submitted to Acta Metall.
3. "An ALCHEMI Investigation of Ternary Site Occupancy in NiAl", P.R. Munroe and I. Baker, submitted to XIIth Int. Cong. Electro. Micro.(1990).
4. "Determination of Ternary Site Occupancy Far from the Stoichiometric Composition", P.R. Munroe and I. Baker, submitted to XIIth Int. Cong. Electro. Micro.(1990).

[†] Invited review paper.

5. "On The Effect of Oxygen on Auger Electron Spectroscopy of FeAl and NiAl", I. Baker, E. P. George, C.T. Liu and R.A. Padgett. Scripta Metall. (in preparation).

6[†]. "Examination of Grain Boundary Structure/Chemistry in Ordered Alloys", I. Baker, to be presented at the ASM Symposium on Structure/Property Relationships for Interfaces, Detroit, MI, Fall 1990, and to be published in a conference proceedings (in preparation).

7. "The Effect of Grain Size on the Yield Strength and Fracture Mode of FeAl and NiAl", I. Baker, P. Nagpal, F. Liu and P.R. Munroe, Acta Metall. (in preparation)

PRESENTATIONS DURING YEAR -3

1. "The Microstructure of Extruded Fe-51 at.% Al," P.R. Munroe* and I. Baker, Annual TMS Meeting, Las Vegas, NV, February 1989.

2. "Some Unusual Aspects of the Deformation of B2 Alloys," I. Baker, Swiss Materials Workshop on Current Topics on Intermetallics for Structural Applications, Neuchatel, Switzerland, March 1989.

3. "Effect of Grain Size on Yield Strength and Fracture Mode of B2 NiAl and FeAl", P. Nagpal*, F. Liu and I. Baker, TMS Annual meeting, Anaheim, CA, February 1990.

4. "Effect of Cooling Rate on Hardness of FeAl and NiAl", P. Nagpal* and I. Baker, TMS Annual meeting, Anaheim, CA, February 1990.

PAPERS TO BE PRESENTED

1[†]. "Heat Treatment and Processing of Nickel Aluminides", I. Baker, to be presented at the ASM Symp. on Heat Treatment and Annealing of Ordered Alloys, Detroit, MI, Fall 1990.

2. "Observation of <001> Dislocations After Room Temperature Deformation and a Mechanism for Transgranular Cleavage in B2 FeAl", P.R. Munroe and I. Baker, to be presented at the TMS Fall meeting Detroit, MI, Fall 1990.

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* Indicates speaker.

[†] Invited review paper.

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