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POINT DEFECT CONCENTRATIONS AND SOLID SOLUTION
HARDENING IN NiAl WITH Fe ADDITIONS

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Abstract

The solid solution hardening behavior exhibited when Fe is added to NiAl is investigated. This is an interesting problem to consider since the ternary Fe additions may choose to occupy either the Ni or the Al sublattice, affecting the hardness at differing rates. Moreover, the addition of Fe may affect the concentrations of other point defects such as vacancies and Ni anti-sites. As a result, unusual effects ranging from rapid hardening to solid solution softening are observed. Alloys with varying amounts of Fe were prepared in Ni-rich (40 at. % Al) and stoichiometric (50 at. % Al) compositions. Vacancy concentrations were measured using lattice parameter and density measurements. The site occupancy of Fe was determined using ALCHEMI. Using these two techniques the site occupancies of all species could be uniquely determined. Significant differences in the defect concentrations as well as the hardening behavior were encountered between the Ni-rich and stoichiometric regimes.

1.0 Introduction

Point defects, such as vacancies and anti-site defects, are known to strongly affect the properties of intermetallic compounds with the B2 crystal structure. For example, the off-stoichiometric hardening observed in NiAl has been attributed to the presence of constitutional defects (1). Thermally generated defects have also been reported to strongly affect the mechanical properties of such B2 compounds as FeAl, CoAl, and FeCo (2-4). In a

similar manner, the addition of ternary or higher order alloy additions can strongly affect the mechanical and physical properties of B2 compounds. e.g. NiAl (5,6). The addition of a ternary element to an ordered binary compound can drastically affect the concentrations of point defects. These defects are often directly responsible for the observed property changes. It is clear that any investigation of the effects of ternary alloying in this class of materials would benefit from a knowledge of the point defect concentration which result from this alloying. It will be the purpose of this investigation to examine the effect of substitutional ternary alloy additions on both the point defect structure and the resultant solid solution hardening behavior of an ordered B2 compound. The system chosen to investigate is NiAl with ternary Fe additions. This system is of interest since it has been reported that Fe additions affect the room temperature ductility (7) and high temperature creep properties (8) of NiAl.

2.0 Experimental

The experiments discussed in this paper are part of a larger study which will be presented elsewhere. For this reason only a brief review of the experimental details will be given here. Twelve alloys were prepared from high purity starting elements by arc melting. Alloys with up to 12 at. % Fe were made in two stoichiometric: Ni-rich (40 at. % Al) and stoichiometric (50 at. % Al). The compositions of the eight alloys are given in Table I. The as-cast alloys were then subjected to homogenizing heat

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Table I. The Compositions and Properties of the Alloys in this Study, as Quenched from 700 and 1000°C

% Ni	% Fe	% Al	Lattice Parameter (Å)		Bulk Density (g/cm ³)		Vacancy Concentration		Hardness (GPa)	
			700°C	1000°C	700°C	1000°C	700°C	1000°C	700°C	1000°C
60	0	40	2.8674	2.8663	6.494	6.476	-0.002	0.002	4.17	4.14
57.6	2.4	40	2.8710	2.8707	6.439	6.437	0.001	0.002	3.78	3.90
54	6	40	2.8770	2.8765	6.385	6.377	0.001	0.003	3.37	3.53
48	12	40	2.8836	2.8832	6.319	6.314	0.001	0.002	2.98	2.98
50	0	50	2.8876	2.8881	5.901	5.886	0.001	0.003	2.50	2.55
48	2	50	2.8881	2.8873	5.883	5.878	0.003	0.004	2.64	2.69
45	5	50	2.8878	2.8868	---	5.842	---	0.009	3.06	3.13
40	10	50	2.8878	2.8861	5.813	5.800	0.010	0.013	3.66	3.77

treatments. Fine powders were prepared from these ingots and given a heat treatment (in vacuum) of 5h/1000°C + water quench (WQ). The lattice parameters of these powders were measured using x-ray diffraction. The remaining portion of the ingots were given a heat treatment of 5h/1000°C + WQ. Sections of the ingots (approximate dimensions: 10 x 10 x 5 mm) were then taken for hardness measurements. The other sections of the ingots were crushed into a coarse powder for density measurements. The density measurements were made using the hydrostatic method. After the lattice parameter, density and hardness measurements, the samples were given a subsequent anneal for 3d/700°C + WQ. The measurements were then repeated. From the lattice parameter and density data, the vacancy concentration of the alloys could be calculated. One discussion of this type of calculation can be found in the review by Chang and Neumann (9).

Selected compositions from above were chosen for analysis with the TEM technique, ALCHEMI. Rods were prepared by arc melting followed by drop casting. These ingots were given homogenizing heat treatments followed by a treatment of 3d/700°C + WQ. TEM samples were prepared from the quenched rods. The details of the ALCHEMI technique will not be presented here, but follow the methods used by Anderson et. al. (10).

3.0 Results and Discussion

3.1 Point Defect Hardening in Binary B2 Compounds

Before attempting to discuss solid solution hardening in a ternary B2 compound it is helpful to review the defect structure and point defect hardening in binary B2 compounds, especially for the NiAl system itself. It has been shown that the defect structure of NiAl can be approximated with the triple defect model (9,11). This model is based on the assumption that the majority of defects present are either vacancies on the Ni sublattice or Ni atoms on the Al sublattice (Ni anti-sites). Defects can be present either as constitutional or thermal defects. Constitutional defects consist of Ni anti-sites at Ni-rich compositions and vacancies at Al-rich compositions. Thermal defects consist of vacancies and Ni anti-sites occurring in a two to one ratio. This ratio maintains the site balance of the crystal. The concentrations of vacancies as well as Ni anti-site defects in NiAl as a function of composition are shown in Fig. 1. The lines in this figure are values estimated from the thermodynamic

model of Chang and Neumann (9). The data points are experimentally measured vacancy concentrations taken from Pike et. al. (4). Good agreement is seen between the model and the experimental values. Off-stoichiometric hardening occurs at both Ni-rich and Al-rich compositions. This hardening has been attributed to Ni anti-sites and vacancies, respectively (1). A recent study by Pike et. al. examined the point defect hardening in several B2 compounds including NiAl. It was found that the hardness could often be correlated to the square root of the hardening defect concentration, following the relationship derived for yield stresses in one solid solution hardening (SSH) theory (12). In Fig. 2 the off-stoichiometric hardening in NiAl is plotted vs. the square root of the hardening defect concentration for both Ni-rich and Al-rich compositions. The hardening rate of vacancies is seen to be significantly greater than that of the Ni anti-sites. For all of the B2 compounds considered by Pike et. al., the hardening rates ($\gamma = dH/d\sqrt{c}$ where H is hardness and c is the hardening defect concentration) of defects fell into two distinct groups. Those of anti-site defects fell in the range G/9 to G/75 and could be linearly related to the term $1/a|da/dc|$ which is a measure of the elastic dilation of the lattice (where G is the shear modulus and a is the lattice parameter). This suggests that the elastic "size effect" is at least partially responsible for the hardening caused by anti-site defects. The hardening rates of vacancies fell in the range G/3.1 to G/4.1. Thus, vacancies are considerably more effective hardeners than are anti-site defects. A correlation between the hardening rates of vacancies and the lattice dilation was not found. The hardening rate of vacancies may be affected by such factors as clustering or association with impurities.

3.2 Point Defect Concentrations in Fe-doped NiAl

To understand the SSH which results from ternary alloying Fe to NiAl, it is important to first develop an understanding of the point defect concentrations which arise as a result of this alloying. Firstly, the Fe additions may occupy either the Ni or the Al sublattice (or both), affecting the hardness in significantly different fashions. Secondly, the addition of Fe may affect other point defect concentrations such as vacancies or Ni anti-site defects. For this reason the first step in our analysis will be to determine the defect concentrations which result from ternary alloying Fe to NiAl.

In a triple defect compound such as NiAl, it is assumed that the Al anti-site (Al^{Ni}) and Al-site vacancy ($\square^{Al}$) concentrations are negligible (we have introduced the notation A^b to indicate the

Figure 1: The vacancy and Ni anti-site concentrations in NiAl at 1000°C as calculated from the thermodynamic model along with experimentally determined vacancy concentrations of samples quenched from 1000°C (4).

Figure 2: The hardness of Ni-rich and Al-rich NiAl vs. the square root of the Ni anti-site and vacancy concentrations, respectively (4).

species A located on the b sublattice; the symbol $\square$ represents a vacancy). Therefore, there are six site occupancies which must be determined to fully characterize the point defect concentration of the crystal. These are the concentrations of Ni^{Ni} , Fe^{Ni} , $\square^{\text{Ni}}$, Ni^{Al} , Fe^{Al} , and Al^{Al} . Through mass and site balances the number of unknowns is reduced to two. Thus, there are two experimental variables which must be determined for any given composition and heat treatment in order to determine all point defect concentrations. The first variable we will measure is the vacancy concentration which can be determined by combining lattice parameter and bulk density measurements. This has been done for a number of binary B2 compounds (4, 9). A second experimental variable can be determined using the TEM technique ALCHEMI. Many

experimental difficulties which plagued the accuracy of early work using this technique have recently been overcome. Using this improved technique Anderson (13) was able to determine the site preference of several ternary additions to the FeAl system. In the present study ALCHEMI was used to determine the $\text{Fe}^{\text{Ni}}/\text{Ni}^{\text{Ni}}$ ratio of several alloys. This information, combined with the vacancy concentration measurements was enough to fully determine the site occupancies resulting from Fe additions to NiAl.

3.2.1 Fe added to Ni-rich NiAl. First we will consider the case where Fe is added to replace Ni in Ni-40Al. Alloys were made from 0 to 12 at. % Fe, i.e. $\text{Ni}_{60-x}\text{Fe}_x\text{Al}_{40}$ where x varies from 0 to 12. The lattice parameters of these alloys as quenched from 1000°C are shown in Fig. 3. The lattice parameter is seen to increase with increasing Fe concentration. The bulk densities of these alloys are listed in Table I. The vacancy concentrations as quenched from 1000°C, as calculated from the lattice parameter and bulk density, are shown in Fig. 4. Here the vacancy concentration is defined as the number of vacancies per the total number of atoms in the crystal. It can be seen that the vacancy concentration is around 0.002 (0.2 %) for all compositions. This is a relatively small value and approaches the threshold of detection. The vacancy concentration of these alloys as quenched from 700°C was also found to be very small. The $\text{Fe}^{\text{Ni}}/\text{Ni}^{\text{Ni}}$ ratio of samples quenched from 700°C was determined using ALCHEMI. This data, combined with the vacancy concentration data, provided the information needed to determine the concentrations of Ni and Fe on the Al sublattice (which we will refer to as Ni and Fe anti-site defects, respectively). These are shown in Fig. 5. From this figure it is clear that as Fe is added to Ni-40Al, Fe anti-site defects begin to replace the Ni anti-site defects. This trend continues until around 10 at. % Fe where the anti-site concentrations begin to level off. At this point Fe begins to occupy the Ni sublattice in significant numbers. The Fe^{Ni} concentration is also shown in Fig. 5 and is significantly lower than the Fe anti-site (Fe^{Al}) concentration for all of the alloys. These results clearly indicate that Fe atoms preferentially occupy the Al sublattice in Ni-rich NiAl. The above results are in qualitative agreement with predictions based on thermodynamic considerations (14).

Figure 3: The lattice parameter vs. Fe concentration.

Figure 4: The vacancy concentration vs. Fe concentration.

Figure 5: The concentrations of NiAl, FeAl, and FeNi defects vs. Fe concentration in the 40 at. % Al alloys.

3.2.2 Fe added to Stoichiometric NiAl. We next consider the case where Fe is added to replace Ni in Ni-50Al. Alloys were made from 0 to 10 at. % Fe, i.e. $\text{Ni}_{50-x}\text{Fe}_x\text{Al}_{50}$ where x varies from 0 to 10. The lattice parameters of the alloys as quenched from 1000°C are shown in Fig. 3 and are seen to decrease slightly as the concentration of Fe is increased. The bulk densities are listed in Table I. The vacancy concentrations are shown in Fig. 4 for samples quenched from 1000°C. These are seen to monotonically increase from 0.003 to 0.013 (0.3 to 1.3%) as the Fe concentration increases from 0 to 10 at. %. A similar trend was observed in the samples quenched from 700°C. The Ni and Fe anti-site concentrations are shown in Fig. 6. These were calculated using the vacancy concentration measurements of the present study combined with the ALCHEMI data from Duncan (15). For "stoichiometric" alloys such as these, any anti-site type defects are thermal defects. From this data it can be seen that the anti-site defect concentrations are small (relative to the number of Fe^{Ni} defects), and that those anti-site defects that are present are primarily Fe anti-sites. The number of Fe anti-sites is seen to increase slightly as the concentration of Fe is increased. However, the

majority of the Fe is located on the Ni sublattice. The concentration of these Fe^{Ni} defects are also shown in Fig. 6. Thermodynamic calculations by Kao et. al. yielded estimates qualitatively similar to the above results (14,16).

Figure 6: The concentrations of NiAl, FeAl, and Fe^{Ni} defects vs. Fe concentration in the 50 at. % Al alloys. This data is calculated using the present vacancy concentration combined with the ALCHEMI data of Duncan (15).

3.3 Solid Solution Hardening in Fe doped NiAl

As discussed in section 3.2, the addition of Fe to NiAl can significantly affect the numbers and types of point defects. Each of these defect types will have a unique hardening rate. In addition, as was seen in the preceding section, there is often more than one type of defect present in significant numbers. Based on information garnered from SSH in binary B2 compounds, much useful qualitative information can be found for the SSH in Fe-doped NiAl. However, if a quantitative method of examining the solid solution hardening behavior is desired when there is more than one defect type present, a method of adding the two effects must be considered. Numerous methods for adding the effects of multiple hardening sources have been suggested (a list of these has been compiled by Gypen and Deruyttere (17)). One of these proposed by Koppelaar and Kuhlmann-Wilsdorf (18) is the "mean square law". This assumption suggests that the square of the total increase in the shear yield stress is equal to the sum of the squares of the changes in the shear yield stress if the two defects were considered separately. In a computer simulation, Foreman and Makin (19) found this rule to be applicable to a wide range of situations. Furthermore, using theoretical arguments Gypen and Deruyttere (17) found the mean square law to be expected when the individual defect hardening follows a square root relationship. If the shear yield stress is assumed to be roughly proportional to the hardness, the mean square law can be extended to the change in hardness:

$$\Delta H = \sqrt{(\gamma_1 c_1^{1/2})^2 + (\gamma_2 c_2^{1/2})^2 + \dots + (\gamma_n c_n^{1/2})^2} \quad [1]$$

where H is the hardness, γ_x and c_x are the hardening rate and defect concentration of the defect of type x , and n is the total number of different types of defects. The proportionality of the hardness to the yield stress is strictly valid only for perfectly plastic materials.

Unfortunately, the brittle nature of many of these compounds often makes direct measurement of the yield stress difficult. In the following sections we will investigate the SSH behavior of Fe-doped Ni-rich and stoichiometric NiAl.

3.3.1 Fe added to Ni-rich NiAl. The hardness of the $\text{Ni}_{60-x}\text{Fe}_x\text{Al}_{40}$ alloys is shown in Fig. 7. The hardness is seen to decrease with increasing Fe concentration. That is, Fe actually causes solid solution softening in Ni-rich NiAl. A rational explanation for this unusual observation can be offered after consideration of the effect of Fe on the point defect concentrations in these alloys. As discussed in section 3.2.1, the two major defects in these alloys are Ni and Fe anti-site defects. The former being replaced by the latter as the concentration of Fe is increased. The Goldschmidt radii of Ni, Fe, and Al are 1.24, 1.27, and 1.43, respectively (20). A smaller Ni atom on the Al sublattice would obviously cause a contraction of the lattice and therefore be expected to harden the material. In a similar manner, an Fe atom on the Al sublattice would be expected to cause hardening. However, as the Fe atom is somewhat larger than the Ni atom, the Fe anti-site would be expected to be a less potent hardener than the Ni anti-site. Therefore, as Fe is added to Ni-40Al, Ni anti-sites (potent hardeners) are being replaced by Fe anti-sites (less potent hardeners). The net result would be a decrease in hardness. This decrease is most likely responsible for the solid solution softening observed in Fig. 7.

Figure 7: The hardness vs. Fe concentration.

Let us now consider whether the observed softening can be rationalized in a quantitative manner. From the study on binary B2 compounds the hardening rate of Ni anti-sites was around $G/12$ (4). However, the hardening rate of Fe anti-sites in NiAl is not directly measurable due to the presence of more than one type of defect. Nevertheless, it is possible to estimate this hardening rate by recalling that the hardening rate of anti-site type defects has been shown to correlate to the lattice dilation of the hardening defect (4). Thus, it becomes necessary to determine the lattice dilation of Fe anti-site defects. From lattice parameter data of binary NiAl, the lattice dilation of Ni anti-sites was found to be -0.076 . The increase in the lattice parameter with increasing Fe concentration seen in Fig. 3 represents the sum of the effects of removing Ni anti-sites while adding Fe anti-sites. If it is assumed that lattice dilations add linearly when there is more than one type of defect present (a

reasonable assumption considering most B2 compounds follow Vegard's law (9)), then the dilation of Fe anti-sites can be estimated from Fig. 3 to be -0.018 . Thus, the lattice dilation of Fe anti-sites is considerably less than that of Ni anti-sites. With this dilation, an estimate for the hardening rate of Fe anti-sites can be made from the observed relationships between hardening rates and lattice dilations in binary B2 compounds (4). A value in the range of $G/25$ to $G/600$ is found, making the Fe anti-site a relatively weak hardener.

Taking the hardening rate of Ni anti-sites ($G/12$) and the data of Fig. 7, Eq. 1 can be solved to estimate the hardening rate of Fe anti-sites. A value of $G/35$ is found. This value lies well within the range estimated using the lattice dilation. The lattice dilations and hardening rates are summarized in Table II.

Table II The Lattice Dilations and Hardening Rates of Point Defects Important to this Study

Defect Type	Sub-lattice	Lattice Dilation $1/a(da/dc)$	Estimated Hardening Rate ^a γ (GPa)	Experimental Hardening Rate ^a γ (GPa)
vacancy	Ni	-0.081 (4)	—	$G/4.1$ (4)
Ni anti-site	Al	-0.076 (4)	—	$G/12$ (4)
Fe atom	Ni	0.007 (21)	$G/40$ to 0 ^b	less than $G/60$ ^c
Fe atom	Al	-0.018	$G/25$ to $G/600$ ^b	$G/35$ ^d

^a The shear modulus, G , of NiAl is 70 GPa (22).

^b These estimates are based on the lattice dilation and the data available for the relationship between the lattice dilations and hardening rates of anti-site defects in binary B2 compounds (4).

^c From applying Eq. [1] to the hardening data for the addition of Fe to stoichiometric NiAl.

^d From applying Eq. [1] to the hardening data for the addition of Fe to Ni-rich NiAl.

3.3.2 Fe added to Stoichiometric NiAl. The hardness of the $\text{Ni}_{50-x}\text{Fe}_x\text{Al}_{50}$ alloys is shown in Fig. 7 to increase as the concentration of Fe is increased. From the discussion in section 3.2.2, we must consider three types of point defects in the stoichiometric alloys: Fe anti-sites, Fe on the Ni sublattice (Fe^{Ni}), and vacancies. All of these defects increase in number as the concentration of Fe increases. The question arises as to how do these different defect types contribute to the observed hardening. The hardening rate of vacancies in NiAl is around $G/4.1$ (4), while the hardening rate of Fe anti-sites was found in the previous section to be around $G/35$. Thus, the only unknown hardening rate is that of the Fe^{Ni} defects. The lattice dilation of these defects has been estimated to be 0.007 using lattice parameter data of Al-rich NiAl (21). Following the relationship between hardening rates and lattice dilations in binary B2 compounds (4), a hardening rate of $G/40$ to 0 can be estimated from this dilation. Solving Eq. [1] using the hardness data in Fig. 7, a hardening rate for Fe^{Ni} defects of less than $G/60$ is calculated. This is well within the range estimated above. Consideration of the terms in Eq. [1] for this case reveals that the potent hardening rate of vacancies is responsible for making the vacancy term the dominant term in the equation. In effect, the observed hardening can be attributed solely to the increasing vacancy concentration. To

illustrate this point consider Fig. 8 where the hardness of the stoichiometric alloys is plotted against the square root of the vacancy concentration. The data is seen to fall along a line. The slope of this line is $G/4.1$, the same hardening rate found for vacancies in binary NiAl (4). The strong role vacancies play in hardening these stoichiometric alloys has also been explored by Duncan (15).

Figure 8: The hardness vs. the square root of the vacancy concentration in the $\text{Ni}_{50-x}\text{Fe}_x\text{Al}_{50}$ alloys.

4.0 Conclusions

The effects of Fe additions on the defect concentrations and the hardness of NiAl were investigated. Two compositional ranges were considered: Ni-rich and stoichiometric. It was found that there were significant differences in the hardening characteristics between the two regimes. An understanding of the site occupancy of the ternary Fe addition as well as the effect of this addition on the concentrations of other point defects, such as vacancies and Ni anti-site defects, was essential in rationalizing the observed hardening behaviors. To determine these defect concentrations, the vacancy concentration was experimentally determined on the Fe doped alloys. In addition, ALCHEMI was performed to determine the site occupancy of the Fe additions as well as the Ni anti-site concentrations. The lattice dilation of Fe defects on both the Ni and Al sublattices was estimated from lattice parameter data. Using the correlation between lattice dilation and hardening rates observed in binary B2 compounds (4), estimates were made of the hardening rates of these Fe defects.

Apparent solid solution softening by Fe was observed in the Ni-rich alloys. This softening was attributed to the replacing of Ni anti-site defects (potent hardeners) with Fe anti-sites (less potent hardeners). The net result being a decrease in hardness with increasing Fe concentration. In the stoichiometric alloys a strong increase in the hardness was observed with increasing Fe concentration. This hardening was found to be stronger than would be expected from the Fe defects alone. This hardening was attributed to the increase in the vacancy concentration which occurs with the addition of Fe.

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References

1. J. Westbrook, *J. Electrochem. Soc.*, **54** (1956), 103.
2. P. Nagpal and I. Baker, *Metall. Trans.*, **21A** (1990), 2281.
3. Y. A. Chang, L. M. Pike, C. T. Liu, A. R. Bilbrey, and D. S. Stone, *Intermetallics*, **1** (1993), 107.
4. L. M. Pike, Y. A. Chang, and C. T. Liu, *Acta Metall.*, in press.
5. R. D. Noebe, R. R. Bowman, and M. V. Nathal, *Inter. Mat. Reviews*, **38** (1993), 193.
6. D. B. Miracle, *Acta Metall.*, **41** (1993), 649.
7. R. Darolia, D. F. Lahrman, and R. D. Field, *Scr. Metall.*, **26** (1992), 220.
8. I. Jung, M. Rudy, and G. Sauthoff, *High Temperature Ordered Intermetallic Alloys II*, vol. **81**, ed. N. S. Stoloff, et al. (MRS Symp. Proc., Pittsburgh, PA 1987) pp. 263-74.
9. Y. A. Chang and J. P. Neumann, *Prog. Solid State Chem.*, **14** (1982), 221.
10. I. M. Anderson, A. J. Duncan, and J. Bentley, in *High Temperature Ordered Intermetallic Alloys VI*, vol. **364**, ed. J. A. Horton, et al. (MRS Symp. Proc., Pittsburgh, PA 1994) pp. 443-54.
11. A. J. Bradley and A. Taylor, *Proc. R. Soc. Lond.*, **A159** (1937), 56.
12. R. L. Fleischer, in *The Strengthening of Metals*, ed. by D. Peckner (Reinhold Publishing Corp., New York, 1964), pp. 93-140.
13. I. M. Anderson, *Acta Metall.*, in press.
14. C. R. Kao, L. M. Pike, S.-L. Chen and Y. A. Chang, *Intermetallics*, **2** (1994), 235.
15. A. J. Duncan, Ph.D. Thesis, University of Florida, 1995.
16. L. M. Pike, C. R. Kao, S. L. Chen, and Y. A. Chang, in *Design Fundamentals of High Temperature Composites, Intermetallics, and Metal-Ceramics Systems*, ed. R. Y. Lin et al. (TMS Publication, Warrendale, PA 1996) p. 171.
17. L. A. Gypen and A. Deruyttere, *J. Mater. Res.*, **12** (1977), 1028.
18. T. J. Koppenaal and D. Kuhlmann-Wilsdorf, *Appl. Phys. Letters*, **4** (1964), 59.
19. A. J. E. Foreman and M. J. Makin, *Phil. Mag.*, **22** (1966), 911.
20. F. Laves, in *Theory of Alloy Phases* (ASM Publication, Cleveland, OH, 1956).
21. L. M. Pike, unpublished data.
22. G. Simmons and H. Wang, *Single Crystal Elastic Constants and Calculated Aggregate Properties: A Handbook*, 2nd ed., (MIT Press, Cambridge, MA, 1971).