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Effect of carbon addition on the glass-forming ability of mechanically alloyed Fe-based alloys

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The effect of carbon addition on the glass-forming ability (GFA) of mechanically alloyed Fe-based $\text{Fe}_{42}\text{M}_{28}\text{Zr}_{10}\text{B}_{20}$ (M=Ni, Al, or Ge) amorphous alloy systems was investigated. It was shown that when B was partially replaced by 10 at. % C in the Fe–Ni–Zr–B and Fe–Al–Zr–B alloy systems, the GFA of the systems had increased significantly, as determined by the reduced milling time required for amorphization. However, when carbon was added to the Fe–Ge–Zr–B alloy system, the GFA was decreased drastically and no amorphization was observed. The role of carbon on the GFA of alloy systems has been discussed from the thermodynamic and kinetic points of view.

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I. INTRODUCTION

Metallic glasses or amorphous alloys have an attractive combination of physical, chemical, mechanical, and magnetic properties. Consequently, they have already found commercial applications and other applications are being explored.^{1–5} However, during the last several years, more attention has been paid to the structure and properties of bulk metallic glasses (BMG's), glassy alloys that could be produced in large section thicknesses or diameters reaching several tens of millimeters.^{6–10} The largest section thickness (or diameter) that could be achieved is a function of the base metal and alloy composition, amongst other parameters. The largest rod diameter obtained in Fe-based BMG's is only about 16 mm,¹¹ while the largest diameter of 72 mm that has been achieved in any of the BMG's is in a Pd–(10 at. %) Ni–(30 at. %) Cu–(20 at. %) P alloy.¹² The ability to produce glassy alloys in larger section thicknesses enables exploitation of these advanced materials for a variety of different applications.

The production of BMG's by consolidation of amorphous powders is another promising method in the near-net-shape fabrication of amorphous materials. Such a situation is possible by producing the amorphous alloy powders via methods such as mechanical alloying (MA) and subsequently consolidating them to bulk specimens using conventional hot isostatic pressing, hot extrusion, or other methods.¹³ In comparison to the solidification route, MA is an inexpensive and simpler technique to process the material into an amorphous state. Further, it is easier to produce the amorphous phase in a much wider composition range by MA than by solidification methods. Additionally, since MA processing is carried out in the solid state, phase diagram restrictions do not seem to apply to the phases produced by the technique. However, due to the large number of variables in the MA process, optimization of parameters becomes essential for obtaining consistent results.^{14,15}

Amorphous alloy powders come into close contact with carbon dies and punches during hot pressing or other consolidation methods and therefore carbon contamination is likely an important concern. Thus, it is useful to evaluate the stability of amorphous alloys in presence of carbon by possibly designing alloy compositions which contain carbon as an additional alloying element. Thus, we report in this paper, the effect of carbon addition on the glass-forming ability (GFA) of Fe-based quaternary alloy systems with a base composition of $\text{Fe}_{42}\text{M}_{28}\text{Zr}_{10}\text{B}_{20}$ (where M=Ni, Al, or Ge). We had earlier reported that all these three systems could be easily amorphized by milling the blended elemental powders for 20, 10, and 10 h, respectively.¹⁶ Since the metalloid content in amorphous alloys is generally around 20 at. %, the compositions of alloys studied in this investigation were chosen to satisfy this requirement. Accordingly, part of boron was replaced by carbon (at 10 at. % level). This carbon content is at a much higher level in comparison to the 3, 4, and 5 at. % as studied earlier by Wang and co-workers,^{17,18} Kim *et al.*,¹⁹ and Inoue *et al.*,²⁰ respectively.

II. EXPERIMENTAL PROCEDURE

Alloy compositions of Fe–M–Zr–B (M=Ni, Al, or Ge) that were shown to amorphize by MA earlier were selected to study the effect of carbon addition. Elemental powders of 99.9% purity were blended to a composition of $\text{Fe}_{42}\text{M}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$. The subscripts represent the atomic percentage of elements in the powder mix. MA was carried out in a SPEX 8000 D shaker mill with a ball-to-powder weight ratio maintained at 10:1. The weighing, blending, and loading of the powders was carried out inside an argon-filled glove box so as to minimize powder contamination. For every composition, 5 g of blended elemental powder was loaded inside a stainless steel vial with 50 g of hardened steel balls. The ball diameters used were of 5 and 8 mm, so as to attain a better milling behavior. The evolution of structure during milling was monitored by collecting the milled powder samples at regular intervals of milling, and analyzed by the x-ray diffraction (XRD) and transmission electron mi-

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FIG. 1. XRD patterns of the $\text{Fe}_{42}\text{Ni}_{28}\text{Zr}_{10}\text{B}_{20}$ powder (without carbon addition) in the as-blended condition and after milling for 20 h. The as-blended powder shows the presence of all the expected diffraction peaks. The powder milled for 20 h shows the presence of a broad diffuse halo indicating the presence of a fully amorphous phase.

crosscopy (TEM) techniques. XRD patterns were recorded using a Rigaku x-ray diffractometer with $\text{Cu K}\alpha$ radiation ($\lambda = 0.154\,06\text{ nm}$) in the 2θ range from 30° to 90° , and TEM studies were conducted using a Tecnai F30 transmission electron microscope operating at 300 kV and equipped with an energy dispersive analyzer.

III. RESULTS

Figure 1 shows the XRD patterns of the $\text{Fe}_{42}\text{Ni}_{28}\text{Zr}_{10}\text{B}_{20}$ powder (without carbon addition) in the as-blended condition (0 h of milling) and also after milling the powder for 20 h. While all the expected diffraction peaks of Fe, Ni, and Zr are seen in the as-blended powder, boron peaks are not seen because of its low atomic scattering factor and also because it is amorphous. On the other hand, the XRD pattern of the powder milled for 20 h clearly shows the presence of a broad diffuse halo indicating the presence of an amorphous phase.

To confirm the nature of the phase(s) in the powder milled for 20 h, TEM investigations were conducted. Figure 2 shows the TEM micrographs and diffraction patterns of the powder milled for 20 h. Figure 2(a) shows a diffraction pattern from the sample, which shows that the milled powder is fully amorphous in nature. However, on observation of the sample in the TEM, the intense electron beam appears to promote (at least partial) crystallization of the amorphous phase. Consequently, a few very tiny crystals of about 10–15 nm size (indicated by arrows) are also occasionally seen inside an amorphous matrix [Fig. 2(b)]. Figure 2(c) shows the corresponding diffraction pattern from this structure. It consists of mostly a diffuse ring indicating that the major phase is an amorphous phase, onto which some sharp diffraction spots are superimposed. These diffraction spots could be indexed to an Fe phase with a bcc structure and with the expected lattice parameter.

Figure 3 shows the structural evolution in the blended elemental powder mixture of $\text{Fe}_{42}\text{Ni}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$ as a function of milling time. The as-blended powder mix showed the presence of all the diffraction peaks expected from the me-

FIG. 2. Transmission electron micrographs of $\text{Fe}_{42}\text{Ni}_{28}\text{Zr}_{10}\text{B}_{20}$ (without carbon addition) powder milled for 20 h to confirm the presence of different phases. (a) Diffraction pattern from the fully amorphous phase. The diffuse ring confirms the presence of an amorphous phase. (b) Bright-field electron micrograph showing the presence of mostly an amorphous phase and a few tiny crystals about 10–15 nm in size. (c) Diffraction pattern from (b) showing the diffuse halo on which a few sharp diffraction spots are superimposed. These diffraction spots are indexed to the bcc iron phase.

tallic elements. Diffraction peaks of B and C were not seen due to their low scattering factor, and probably due to the amorphous nature of boron (and carbon). Formation of the amorphous phase is noted on milling the powder blend for 8 h which is evidenced by the appearance of a broad diffuse peak with a maximum corresponding to where the crystalline $(110)_{\text{Fe}}$ peak was expected to be present. The homogeneous amorphous phase in this powder blend is stable until about

FIG. 3. XRD patterns of blended elemental powder mix of Fe₄₂Ni₂₈Zr₁₀C₁₀B₁₀ as a function of milling time. Note that the glassy phase is formed on milling the powder for 8 h and is found to be stable till 15 h. Mechanical crystallization is noted to occur from around 20 h. The intensities of the crystalline peaks of the different phases increased beyond this milling time, and some of these intermetallics are identified in the pattern recorded from the powder milled for 30 h.

15 h of milling time. At a later stage, from about 20 h, formation of some crystalline phases is noted as a result of mechanical crystallization of the amorphous phase. The mechanical crystallization is clearly evident on milling the powder blend for about 30 h, as indicated by the relatively intense diffraction peaks of the different phases. Formation of several phases along with the α -Fe solid solution is noted in the powder blend at 30 h of milling time.

Figure 4 shows the XRD patterns of the Fe₄₂Al₂₈Zr₁₀C₁₀B₁₀ powder mix as a function of milling time. Formation of α -Fe solid solution was clearly noted at short milling times, indicated by the presence of low intensity crystalline peaks at 5 h of milling time. Subsequent mill-

FIG. 4. XRD patterns of the blended elemental powder mix of Fe₄₂Al₂₈Zr₁₀C₁₀B₁₀ as a function of milling time. Note that the glassy phase formed on milling the powder for 5–8 h. Mechanical crystallization is observed to take place from about 15 h, evidenced by the formation of low intensity crystalline peaks, marked by arrows. It is noted that after crystallization, there is a tendency for reamorphization at 25 h (but a very small amount of crystalline phase may be present). Reasonably intense peaks of some crystalline phases are observed beyond this milling time, say, at 30 h.

FIG. 5. XRD patterns of the blended elemental powder mix of Fe₄₂Ge₂₈Zr₁₀C₁₀B₁₀ as a function of milling time. Formation of α -Fe-solid solution is observed at 5 h of milling time along with some compounds based on the other elements in the blend. Milling for long times (say, 30 h) transforms the alloy system into a combination of several interstitial compounds.

ing led to the formation of an amorphous phase. From the diffraction patterns, it appears that amorphization is occurring on milling the powder for a time between 5 and 8 h. On further milling, mechanical crystallization of the glassy phase commences at around 15 h, which is evidenced by the formation of low intensity crystalline peaks, marked by arrows. The stability of the amorphous phase does not appear to be significantly different between the Ni-containing and Al-containing compositions. For example, low intensity crystalline peaks of α -Fe start appearing in the Al-containing alloy after milling the powder for 15 h. A set of low intensity crystalline peaks start appearing around this time of milling in the Ni-containing alloy also. But, the nature of the crystalline phase(s) forming in this alloy composition appears to be different. This becomes more prominent at longer milling times. At the longest milling time recorded, the number and amount of crystalline phases seems to be more in the Ni-containing alloy than in the Al-containing alloy. The diffraction pattern for the Ni-containing alloy at 30 h of milling time shows more and also relatively high intensity peaks (compare Figs. 3 and 4). But, an important difference between these two alloy compositions is that the Fe₄₂Al₂₈Zr₁₀C₁₀B₁₀ alloy system shows a tendency for reamorphization on milling the powder for about 25 h. This is somewhat similar to the cyclic crystalline $\rightarrow$ amorphous $\rightarrow$ crystalline transformations observed in some alloy systems.^{21–25}

Figure 5 shows the XRD patterns for the Fe₄₂Ge₂₈Zr₁₀C₁₀B₁₀ blended elemental powder mixture as a function of milling time. Interestingly, the quinary system with carbon addition did not show any signs of amorphization. This should be contrasted with the Fe₄₂Ge₂₈Zr₁₀B₂₀ quaternary system (i.e., without carbon) in which the amorphous phase had formed in 10 h during MA.¹⁶ As evident from Fig. 5, at the initial stages of milling, the α -Fe solid solution is present along with some other phases. With continued milling, interstitial compounds based on Zr and Fe and some intermetallics seem to have formed. It is also ob-

TABLE I. Comparison of milling times required for amorphization with and without addition of C to $\text{Fe}_{42}\text{M}_{28}\text{Zr}_{10}\text{B}_{20}$ alloy system (M=Al, Ni, or Ge).

Quaternary system without carbon	Milling time for amorphization (h)	Quinary system with carbon	Milling time for amorphization (h)
$\text{Fe}_{42}\text{Ni}_{28}\text{Zr}_{10}\text{B}_{20}$	20	$\text{Fe}_{42}\text{Ni}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$	8
$\text{Fe}_{42}\text{Al}_{28}\text{Zr}_{10}\text{B}_{20}$	10	$\text{Fe}_{42}\text{Al}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$	5–8
$\text{Fe}_{42}\text{Ge}_{28}\text{Zr}_{10}\text{B}_{20}$	10	$\text{Fe}_{42}\text{Ge}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$	No amorphization

served from this figure that on milling for a long time, the solid solution formed at the early stages of milling, has transformed into carbides and borides. An unidentified phase (along with other phases) is observed at 30 h of milling time.

Results of the effect of C addition with respect to the three quaternary systems studied in this investigation are summarized in Table I. The results clearly show the improvement of GFA in alloy systems containing Ni and Al, where the milling time for amorphization has decreased from 20 to 8 h in the case of Fe–Ni–Zr–C–B and from 10 to 5–8 h in the case of Fe–Al–Zr–C–B. In contrast, the Fe–Ge–Zr–C–B alloy system showed a reduced GFA, and amorphization was not very clearly observed.

IV. DISCUSSION

The XRD patterns presented in Figs. 3–5 and the results summarized in Table I show that addition of 10 at. % C has a significant effect on the GFA of the Fe-based alloy systems investigated here. While the Ni- and Al-containing alloys have shown improved GFA, the one with Ge showed a decreased GFA and did not show amorphization on addition of carbon.

Wang and co-workers^{17,18} showed that while a small amount of carbon addition (up to about 1 at. %) increased the GFA of Zr-based glassy alloy systems, devitrification (crystallization) was noted on increasing the C content to about 3 at. % or higher. That is, higher carbon concentrations in the alloy led to crystallization of the amorphous phase formed. In another study, Kim *et al.*¹⁹ fabricated Misch metal-based BMG's and noted that the GFA of these alloys was enhanced with the addition of 4 at. % carbon. In comparison to the above studies, we find our results quite novel and important for processing of Fe-based amorphous alloys. In our study, the effect of carbon addition is seen to be varying depending on the constituent elements in the alloy. From Table I it is noted that the carbon addition enhances the GFA of Ni- and Al-containing Fe-based alloys and that it has a much more significant effect on the GFA of the Ni-containing system than on the Al-containing system. This is because the time required for the formation of the amorphous phase was brought down from 20 to 8 h in the Ni-containing system, whereas it was brought down to 5–8 h from 10 h in the Al-containing system.

Even though the effect of carbon addition is seen to increase the GFA of the two alloy systems containing Al and Ni, an important difference noted is that reamorphization seems to take place in the Al-containing alloy system. In other words, the Al-containing system possesses a tendency to resist crystallization. Further, it is also noted that the num-

ber of phases is much less and with relatively low intensity peaks at the end of milling for 30 h, as seen in Fig. 4. Since crystallization is a kinetic process and it requires time for the crystalline phases to form, this reamorphized phase would require a much longer time for sufficient amount of the crystalline phases to form. Milling for a longer time, say, for 50 h, would perhaps produce a much larger amount of the crystalline phases in the Al-containing alloy. This could explain why the number and intensity of the crystalline peaks is higher in the Ni-containing alloy and less in the Al-containing alloy.

In the system containing Ge, however, the results were unexpected since, in our earlier study, the quaternary Fe–Ge–Zr–B showed good GFA,¹⁶ whereas a detrimental effect on glass formation was observed when C was added.

The role of carbon on the GFA of alloys can be discussed on the basis of its kinetic and thermodynamic aspects. It is known that the presence of metalloid atoms disrupts the short-range order of the type necessary for the formation of crystal nuclei and thus can lead to formation of an amorphous phase.²⁶ This seems to be due to the small size of metalloid atoms as compared with the metal atoms among which they are mixed. Also, due to the strong attractive bonds present between neighboring metal-metalloid pairs,^{27,28} the metalloid atoms tend to avoid each other as neighbors. That is why the metalloid atoms are surrounded by metal atoms forming clusters. Furthermore, the addition of carbon atoms increases the degree of dense random packing of the alloy system. Additionally, since the diameter of the carbon atom (0.142 nm) is much smaller than that of Fe (0.248 nm), Zr (0.318 nm), Ni (0.250 nm), or Al (0.286 nm), this large difference in atomic size suppresses the long-range interdiffusion required for crystallization. All these factors enhance the GFA, as indicated by the reduced milling time required for amorphization.

Looking at the thermodynamic aspects, carbon has positive heat of mixing with Fe (+40 kJ mole⁻¹), Ni (+51 kJ mole⁻¹), and Al (+54 kJ mole⁻¹), and a large negative heat of mixing with Zr (−98 kJ mole⁻¹).²⁹ Also, other constituent elements (e.g., Zr and B) have a negative heat of mixing, thus satisfying Inoue's empirical criteria for glass formation.^{6,9} Consequently, the presence of elements with a combination of positive and negative enthalpies of mixing frustrates the formation of competing crystalline phases, as seen in the studies by Park *et al.*^{30,31} Therefore, we observe an increase in the GFA for the Fe–Ni–Zr–C–B and Fe–Al–Zr–C–B alloy systems.

It has been observed that in Fe-based alloys containing metalloids, the major crystalline phases coexisting with the

glassy matrix are always the Fe-metalloid-type compounds.³² In our study, the atomic size differences among the constituent elements are much more in the Al-containing system than in the Ni-containing system. Accordingly, the dense random packing of atoms is higher in the Al-containing system, which is appropriate for amorphization. Further, the strong bonding between Al and B/C atoms causes formation of Al–C and Al–B atomic pairs in the glassy phase. The presence of these Al-metalloid pairs prevents the atomic rearrangement required for the precipitation of the major crystalline phases of the Fe-metalloid compounds in the Fe–Al–Zr–C–B system. Thus, the tendency to resist crystallization, or increased tendency for amorphization, can be rationalized in the Fe–Al–Zr–C–B alloy system.

Sinclair *et al.*³³ have proposed a mechanism, wherein the crystallization temperature of an amorphous phase could be significantly reduced in the presence of some metals, particularly those that form a eutectic phase diagram with Si and Ge. This is known as metal-mediated crystallization. It was shown³⁴ by high-resolution transmission electron microscopy methods that the mechanism involves dissolution of the amorphous element (e.g., carbon) in the matrix thereby supersaturating it. Subsequently, precipitation of the solute atoms in the metal matrix, followed by the diffusion of atoms from the amorphous to crystalline nuclei, initiates the reaction at much lower temperature than can occur in the elemental matrix. In our case, Fe–Ge alloy system features a eutectic reaction as seen from the binary phase diagram³⁵ and thus, addition of carbon which is amorphous in nature, induces crystallization, thus explaining the nonamorphization behavior of the Fe–Ge–Zr–C–B alloy system.

V. CONCLUSIONS

The glass-forming ability (GFA) of some mechanically alloyed Fe-based alloy systems was improved by the addition of carbon. It was also seen that the amount of carbon can be increased to as high as 10 at. % without any negative effect on glass formation in some of the systems. This aspect becomes important in consolidation of amorphous alloy powders where carbon dies are commonly used and carbon pickup by the powder compact is inevitable. Alloy systems with the composition $\text{Fe}_{42}\text{Ni}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$ and $\text{Fe}_{42}\text{Al}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$ show an improvement in GFA which is indicated by the shorter milling times (in contrast to the quaternary systems without carbon) required for amorphization. On the other hand, the alloy composition containing Ge ($\text{Fe}_{42}\text{Ge}_{28}\text{Zr}_{10}\text{C}_{10}\text{B}_{10}$) showed a drastic decrease in GFA with the carbon addition, and no amorphous phase formation was noted. The role of carbon addition has been attributed to the atomic size effects, its positive heat of mixing with some constituent elements, and metal-mediated crystallization which occurs in the Ge-containing system.

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- ¹F. E. Luborsky, *Amorphous Metallic Alloys* (Butterworths, London, 1983).
- ²C. Suryanarayana, *Bull. Mater. Sci.* **6**, 579 (1984).
- ³*Metallic Glasses: Production, Properties and Applications*, edited by T. R. Anantharaman (Trans Tech, Aedermannsdorf, Switzerland, 1984).
- ⁴R. W. Cahn, in *Materials Science and Technology: A Comprehensive Treatment*, edited by J. Zarzycki (Wiley VCH, Weinheim, 1991), Vol. 9, pp. 493–548.
- ⁵*Rapidly Solidified Alloys: Processes, Structures, Properties, and Applications*, edited by H. H. Liebermann (Marcel Dekker, New York, 1993).
- ⁶A. Inoue, *Acta Mater.* **48**, 279 (2000).
- ⁷W. L. Johnson, *MRS Bull.* **24**(10), 42 (1999).
- ⁸J. F. Löffler, *Intermetallics* **11**, 529 (2003).
- ⁹A. Inoue, *Bulk Amorphous Alloys: Preparation and Fundamental Characteristics*, Materials Science Foundations Vol. 4 (Trans Tech, Zurich, Switzerland, 1998).
- ¹⁰A. Inoue, *Bulk Amorphous Alloys: Practical Characteristics and Applications*, Materials Science Foundations Vol. 6 (Trans Tech Publications, Zurich, Switzerland, 1999).
- ¹¹J. Shen, Q. Chen, J. Sun, H. Fan, and G. Wang, *Appl. Phys. Lett.* **86**, 151907 (2005).
- ¹²A. Inoue, N. Nishiyama, and H. Kimura, *Mater. Trans., JIM* **38**, 179 (1997).
- ¹³Q. Li, *Mater. Sci. Eng., A* **471**, 75 (2007).
- ¹⁴C. Suryanarayana, *Prog. Mater. Sci.* **46**, 1 (2001).
- ¹⁵C. Suryanarayana, *Mechanical Alloying and Milling* (Marcel Dekker, New York, 2004).
- ¹⁶S. Sharma, R. Vaidyanathan, and C. Suryanarayana, *Appl. Phys. Lett.* **90**, 111915 (2007).
- ¹⁷W. H. Wang, Q. Wei, and H. Y. Bai, *Appl. Phys. Lett.* **71**, 58 (1997).
- ¹⁸W. H. Wang and H. Y. Bai, *J. Appl. Phys.* **84**, 5961 (1998).
- ¹⁹J. H. Kim, J. S. Park, H. T. Jeong, W. T. Kim, and D. H. Kim, *Mater. Sci. Eng., A* **386**, 186 (2004).
- ²⁰A. Inoue, Y. Shinohara, and J. S. Gook, *Mater. Trans., JIM* **36**, 1427 (1995).
- ²¹M. Sherif El-Eskandarany, K. Aoki, K. Sumiyama, and K. Suzuki, *Scr. Mater.* **36**, 1001 (1997).
- ²²M. Sherif El-Eskandarany, K. Aoki, K. Sumiyama, and K. Suzuki, *Appl. Phys. Lett.* **70**, 1679 (1997).
- ²³M. Sherif El-Eskandarany, K. Aoki, K. Sumiyama, and K. Suzuki, *Acta Mater.* **50**, 1113 (2002).
- ²⁴W. C. Johnson, J. K. Lee, and G. J. Shiflet, *Acta Mater.* **54**, 5123 (2006).
- ²⁵M. Sherif El-Eskandarany and A. Inoue, *Phys. Rev. B* **75**, 224109 (2007).
- ²⁶T. A. Weber and F. H. Stillinger, *Phys. Rev. B* **31**, 1954 (1985).
- ²⁷G. S. Cargill III, *J. Appl. Phys.* **41**, 12 (1970).
- ²⁸J. Dixmier and A. Guinier, in *Physics of Non-Crystalline Solids*, edited by J. A. Prins (North-Holland, Amsterdam, 1965), p. 67.
- ²⁹F. R. de Boer, R. Boom, W. C. M. Mattens, A. R. Miedema, and A. K. Niessen, *Cohesion in Metals: Transition Metal Alloys* (North Holland, Amsterdam, 1988).
- ³⁰E. S. Park and D. H. Kim, *Acta Mater.* **54**, 2597 (2006).
- ³¹E. S. Park, D. H. Kim, T. Ohkubo, and K. Hono, *J. Non-Cryst. Solids* **351**, 1232 (2005).
- ³²A. Inoue, Y. Shinohara, and J. S. Gook, *Mater. Trans., JIM* **36**, 1427 (1995).
- ³³R. Sinclair, T. Itoh, and R. Chin, *Microsc. Microanal.* **8**, 288 (2002).
- ³⁴R. Sinclair and T. J. Konno, *Ultramicroscopy* **56**, 225 (1994).
- ³⁵*Binary Alloy Phase Diagrams*, edited by T. B. Massalski (ASM International, Materials Park, OH, 1986).