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CONTROLLING EMBRITTLEMENT AT GRAIN BOUNDARIES IN INTERMETALLIC COMPOUNDS

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Abstract - The present paper first presents some characteristics of structure and electronic (bond) nature for the grain boundary brittleness of the intermetallic compounds. Next, the phenomenological aspects for the grain boundary brittleness of the L1₂-type intermetallic compounds are reviewed with respect to the combination of the constituent atoms, the alloying effect, the stoichiometry effect, and the role of impurity or gaseous atoms. It is emphasized that the brittleness of grain boundaries in the intermetallic compounds are intimately related to the atomistic and electronic structures at grain boundary regions. Based on these systematic observations, it is suggested that the brittleness of the L1₂-type intermetallic compounds can be manipulated by appropriate control of composition and the corresponding electrochemical bond environment at grain boundary planes, and by control of the test environment. Finally, some instances pointing to the materials development are described.

1. INTRODUCTION

Material properties of intermetallic compounds and ordered alloys are regarded as being intermediate and having widely distributed spectra between metals and ceramics. Some intermetallic compounds have significant characteristics regarding the mechanical properties. Particularly, elevated mechanical properties are very attractive as new structural alloys since the intermetallic compounds occasionally show an increase in flow strength with increasing temperature and excellent corrosion and creep resistance. Also, many intermetallic compounds have been used by their own unique properties as functional materials. It has been, however, known that the intermetallic compounds are commonly very brittle. Therefore, they have not been fabricated into useful shapes and also have been alienated from practical uses. The brittleness of the intermetallic compounds is caused by strong tendencies to cleave in grain interiors or at grain boundaries. The former is attributed to a characteristic that the crystal structure of the intermetallic compounds usually has a large unit cell, compared with the ordinary metals and alloys, and then produces a
higher Peierls stress, resulting in a lack of operative slip systems. With regard to the grain boundary brittleness of the polycrystalline intermetallic compounds, the mechanism due to the segregation of impurities to grain boundaries [1], which has been applied traditionally to the ordinary metals and alloys, has been proposed. Further, a unique mechanism [2] applicable to the ordered structure has been demonstrated; the restriction of the number of available slip systems or cross-slip suppresses the deformability of the intermetallic compounds. However, neither of these mechanisms were successfully applied to a large number of the $L_1_2$-type intermetallic compounds, since AES measurements in some $L_1_2$-type compounds [3,4] have shown that the fractured grain boundaries were quite clean and free from any impurity atoms, and also since the sufficient number of slip systems (and sufficient cross-slip) is operative in these alloys. Recent study concerning intergranular fracture mechanisms suggests that grain boundary fracture directly relates to the structure and bond environment of the grain boundary itself [5]. If atomic knowledge of grain boundary in the intermetallic compounds could be correlated with their mechanical response, it may be possible to clarify these mechanisms and, furthermore, it may be possible to obtain high grain boundary cohesion so that plastic deformation of the grain interiors can be sustained.

The present paper first discusses the grain boundary brittleness of the $L_1_2$-type intermetallic compounds from the point of view of structural and electronic concepts on their grain boundary regions. Next, strength and the grain boundary fracture of the $L_1_2$-type intermetallic compounds will be reviewed in various kinds of alloy systems that have been currently investigated by the present author. Based on the phenomenological conclusion derived from this review, it will be suggested that the brittleness of grain boundaries can be manipulated by appropriate control of composition of the constituent atoms, alloying atoms, and intended or unintended impurity atoms, and further by control of test conditions. Also, some instances looking toward the development of improved materials will be shown.

2. STRUCTURAL (ELECTRONIC) ASPECT AND COHESIVE STRENGTH AT GRAIN BOUNDARIES

It is well known that many intermetallic compounds frequently show different properties even though they have quite similar crystal "structures". This is usually attributed to the difference of the bond nature of the constituent atoms. Here, emphasis will be placed on understanding the grain boundary brittleness from the point of view of the atomistic (and electronic) structure of the grain boundary region since this characteristic is uniquely responsible for the intermetallic compounds.

In the strongly ordered $A_3B$ $L_1_2$ structure, the bonds will be characterized by their 'directionality' and the 'heteropolarity' of their electronic charge distribution. Concerning the former feature, these intermetallic compounds have a contribution to the bond due to hybridization between the d electrons of the transition metals (A atom) and the p electrons of the b-subgroup elements (B atom) and therefore are called a "heteroelectronic" structure [6-8]. It has been pointed out that this bond is directional and shortens the nearest neighbor distance between the A and B atoms [8]. In the weakly ordered $L_1_2$ structure, i.e. the "homoelectronic" ordered $L_1_2$ alloys, such a directional character is less pronounced. With regard to the latter feature, in the heteroelectronic $A_3B$ $L_1_2$ alloy, the chemical bonds between A and A atoms, or B and B atoms or A and B atoms possess different spatial electron charge densities, depending on their individual bond natures. Consequently, the electron density will be decreased or enhanced. However, in the homoelectronic $A_3B$ $L_1_2$ alloy, the transfer of electron charge will be less substantial, resembling those in pure metals, metallic bonds.

The grain boundary structure in the $L_1_2$-type intermetallic compounds has been shown to have various kinds of configurations, i.e. of the bond pairs crossing the grain boundary plane, depending on misorientation, rotation axis and type of grain boundary [9,10]. However, as generally described, a majority of bond pairs are A-A bonds and a minority are A-B bonds. The B-B bonds are scarce. As an example, Fig. 1 shows the atomic configuration and the bond environment geometrically constructed in the $\Sigma = 5$, 53.13 CSL boundary with [100] rotation axis [9].

Now, the correlation between the electronic nature of grain boundaries region and grain boundary brittleness is considered. First, from the point of view of the
The spatial electronic character of the bonds in the grain boundaries of the \( \text{L}_{12} \)-type alloys leads to promising methods for controlling the grain boundary brittleness if the heterogeneity of charge distribution along grain boundaries is minimized by a combination of proper constituent atoms, or by alloying with the proper elements, or if the electron charge efficiency of the A-A pairs is filled up by adding the elements acting as electron donors, or if the heterogeneity of electron charge by the harmful impurity element is excluded. When this idea is applied to material development in an engineering sense, the technique based on this may be called the "macro-alloying" and the "micro-alloying" methods, respectively.

3.1 Selection of Alloy Constituents

Very systematic observation for the grain boundary brittleness of the \( \text{L}_{12} \)-type \( \text{A}_3\text{B} \) alloys was performed in terms of the combination of the constituent atoms.
Table 1 shows the alloy systems which were investigated [16]. The alloys in which the A component consisted of 1st transition group VIII metals and the B component consisted of b-subgroup elements failed through intergranular cracking with no measurable bend ductility (Fig. 2a), whereas those for which the B component was another transition metal failed through transgranular cracking with appreciable bend ductility (Fig. 2b). To find a parameter that would separate the two group of alloys many comparisons between various parameters such as grain size, grain morphology, the strength of matrix were made. However, neither of these success-fully predicted the grain boundary brittleness.

To evaluate more quantita-tively the grain boundary brittleness, the electronic parameters may be considered. Ex-amples are the valence value between the component atoms, the electro-negativity, and parameters in Miedema's model for the heat of formation of the intermetallic compounds [17]. The evaluation by the valence difference between two constituent atoms was the most successful, although this criterion is less clear in terms of the electronic structure mentioned above. The valence differences in the heteroelectronic alloys were usually larger than in the homeoelectronic alloys as shown in Table 1. Thus, it is predicted in the Ni-based A3B alloys that the degree of the grain boundary fragility is in the sequence of Ni3Ge = Ni3Si > Ni3Ga = Ni3Al > Ni3Mn > Ni3Pd. However, when the valence differences are compared among the alloys with different A atoms, the quantitative evaluation will be less successful.

This argument predicts that the L12-type intermetallic compounds consisting of two transition metals must be ductile. It was indeed shown that Co3Ti was noticeably ductile over a wide range of temperatures [18,19]. Also, another example can be found in the h.c.p.-type ordered Co3V which has been known to be brittle. When this alloy was altered into f.c.c. (L12)-type structure by the replacements of Co with both of Ni and Fe [20,21], extensive ductilities were found [22]. The origin (mechanism) of this ductilisation should be attributed not only to the change of crystal structure but also to the fact that two components consisted of transition metals.

Table 1 Various kind of the L12-type A3B alloys which were investigated for the grain boundary brittleness. The valence differences between the constituent atoms are shown as a criterion to predict the grain boundary brittleness.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Valence of B atom</th>
<th>Valence difference between constituent atoms</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni3Ge</td>
<td>4 (4)</td>
<td>4 (3.75)</td>
</tr>
<tr>
<td>Ni3Si</td>
<td>4 (4)</td>
<td>4 (3.75)</td>
</tr>
<tr>
<td>Ni3Ga</td>
<td>3 (3)</td>
<td>3 (2.75)</td>
</tr>
<tr>
<td>Fe3Ga</td>
<td>3 (3)</td>
<td>2.8 (2.08)</td>
</tr>
<tr>
<td>Ni3Al</td>
<td>3 (3)</td>
<td>3 (2.75)</td>
</tr>
<tr>
<td>Co3Ti</td>
<td>4.2 (3.92)</td>
<td>3.4 (3.20)</td>
</tr>
<tr>
<td>Cu3Pd</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Ni3Mn</td>
<td>0.9 (1.35)</td>
<td>0.9 (1.10)</td>
</tr>
<tr>
<td>Ni3Fe</td>
<td>0.2 (0.92)</td>
<td>0.2 (0.67)</td>
</tr>
</tbody>
</table>

Table 2 Substitution nature of the ternary elements in the Ni3Al alloy and the valence difference as a criterion to predict the alloying effect on the grain boundary brittleness of the Ni3Al.

<table>
<thead>
<tr>
<th>Third Electronic Substitution Valence element nature</th>
<th>Side</th>
<th>difference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sb Vb Al</td>
<td>-2</td>
<td></td>
</tr>
<tr>
<td>Si IVb Al</td>
<td>-1</td>
<td></td>
</tr>
<tr>
<td>Ge IVb Al</td>
<td>-1</td>
<td></td>
</tr>
<tr>
<td>Al IIIb A1</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>Ga IIIb Al</td>
<td>0</td>
<td></td>
</tr>
<tr>
<td>Cu Ib Ni</td>
<td>? (+0.25)</td>
<td></td>
</tr>
<tr>
<td>Ni VIII A1</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>Co VIII Ni</td>
<td>-0.8(-0.47)</td>
<td></td>
</tr>
<tr>
<td>Fe VIII A1</td>
<td>+2.8(+2.08)</td>
<td></td>
</tr>
<tr>
<td>Mn VIIa A1</td>
<td>+2.1(+1.65)</td>
<td></td>
</tr>
<tr>
<td>Cr Vla Al</td>
<td>+1.8(+1.31)</td>
<td></td>
</tr>
<tr>
<td>Mo Vla Al</td>
<td>0</td>
<td></td>
</tr>
<tr>
<td>W Vla Al</td>
<td>0</td>
<td></td>
</tr>
<tr>
<td>V Va Al</td>
<td>+1.5(+0.81)</td>
<td></td>
</tr>
<tr>
<td>Nb Va Al</td>
<td>?</td>
<td></td>
</tr>
<tr>
<td>Ta Va Al</td>
<td>?</td>
<td></td>
</tr>
<tr>
<td>Ti Iva Al</td>
<td>-1.2(-0.02)</td>
<td></td>
</tr>
<tr>
<td>Sc IIIa Ni</td>
<td>?</td>
<td></td>
</tr>
</tbody>
</table>
3.2. Alloying of Substitutional Elements

Systematic observations for the alloying effect on the grain boundary brittleness of the Li$_2$-type structure have been reported in two alloys of Ni$_3$Al [23] and Co$_3$Ti [24].

Ni$_3$Al has been previously known to be extremely brittle in its polycrystalline form. The preferential site occupations of the substitutional ternary elements in Ni$_3$Al are summarized in Table 2. The grain boundary brittleness of the ternary Ni$_3$Al alloys has been investigated by mechanical test and fractographical observation [23]. The ternary elements were added to Ni$_3$Al with stoichiometric composition at the expense of Ni or Al content, obeying the preferential site occupancies of the elements. The alloys added with the b-subgroup elements (Sb, Si, Ge and Ga) failed by intergranular cracking with no measurable ductility. This propensity for brittle grain boundary fracture was not improved over their entire solubility ranges. For the alloys added with true metals or noble metal of Cu(Ib), Co(VIII), Fe(VIIa) and Mn(VIIa), the alloys added with former two elements (Cu and Co) failed via intergranular cracking with no measurable ductility, whereas the alloys added with latter two elements (Fe and Mn) began to show the ductility at about 6 at.% addition and then their ductilities increased with increasing solute contents. Correspondingly, the fracture mode varied from intergranular cracking, through mixed modes, to transgranular cracking. The results of fractographs and bend ductilities for the alloys are shown in Fig. 3. On the other hand, the alloys added with many remaining tran
sition elements again failed by intergranular cracking with no measurable ductility, similar to the binary Ni$_3$Al.

Depending on the electronic character and substitutional nature of the ternary atoms, three types of grain boundary structures may be schematically demonstrated. In Fig. 4a, it is shown that the additions to Al sites of the b-subgroup elements and many transition elements strengthen the covalent bonds between the Ni and the b-subgroup elements and thereby lead to a more heterogeneous electronic charge distribution. This would be responsible for the further reduction of the grain boundary cohesive strength of Ni$_3$Al. Fig. 4b illustrates that with additions of Cu or Co atoms, it is unlikely that the electronic charge distribution will be modified, because the Cu(or Co)-Al bond is similar to the original Ni-Al bonds. As a result, the improvement of the grain boundary cohesive strength is not expected. However, Fig. 4c demonstrates that additions of Mn or Fe create a more homogeneous charge distribution, accounting for the increased grain boundary cohesive strength.

It is again possible to evaluate quantitatively the alloying effect on the grain boundary cohesive strength of Ni$_3$Al if we take a criterion based on the valence value. The valence difference between the third element and the solvent element substituted by the former element is proposed as this criterion (see Table 2). It is predicted that the grain boundary cohesive strength is enhanced when the valence difference is positive and vice versa. The valence differences in the alloys with Mn and Fe additions were found to have the greatest, i.e. +2.1 and +2.8, respectively. On the other hand, those with Cu and Co additions were found to be +0.25 and -0.8, respectively, which are sufficiently close to zero as to have no effect on the modification of the electronic environment. Most ternary elements in b-subgroup are inherently "embrittling atoms" in Ni$_3$Al because of the negative values of their valence differences.

Similar effects of alloying on the grain boundary brittleness was seen in Co$_3$Ti [24]. The preferential site occupations of the ternary elements (X) in Co$_3$Ti have been determined by the extended Y' phase lobe in the Co-Ti-X phase diagrams, as shown in Fig. 5 [25]. In this case, the alloys added with many transition metals which substitute exclusively for Ti sites showed almost identical values to or lower values than that of the tensile elongation of pure Co$_3$Ti. However, the alloys added with VIII metals of Fe which substitute for both sites of Ti and Co showed higher values of the tensile elongation than unalloyed Co$_3$Ti, as shown in Fig. 6. The improvement of the ductility observed in the ternary alloy with Fe additions can be attributed to alternating a portion of Co-Ti bond pairs with Co-Fe bond pairs.

The fact that this phenomenon depended not only on the electronic nature of the third elements but also on the substitution nature implies that the atomic accommodations in the grain boundary region are much fewer in comparison with pure metals or disordered alloys. If this is fully established, the third elements with similar
electronic nature should show similar effect on the intergranular brittleness.

Through the principle deduced from the alloying effect on the ductilization in the L12-type intermetallic compounds, the materials development work is under way for Ni3(Al,Mn) [26,27] and Co3Ti+X [24] because these alloys have outstanding mechanical, physical and chemical properties.

3.3 Control of Stoichiometry

It is well known that stoichiometry in intermetallic compounds is a very important crystallographic parameter affecting their physical and mechanical properties. In the L12 structure, the substitutional defect has been investigated in off-stoichiometric Ni3Al [28] and Co3Ti [29].

Recent work on several L12-type A3B intermetallic compounds in which component A atom consists of VIII metals showed that as A atoms became excess from stoichiometry, elongation always increased and correspondingly the transgranular fracture mode became dominant [30]. Figure 7 represents this behavior for each alloy. Also, it is very evident from these curves that the compositional dependence on the grain boundary brittleness (or ductility) is more remarkable in the sequence of Ni3Al > Ni3Ga > Co3Ti > Ni3(Al,Mn) although the former two alloys were doped with boron and then ductilized (see later).

Since non-stoichiometric (i.e. excess) atoms are considered a sort of ternary atom in binary A3B intermetallic compounds, the degree (i.e. the slope) of the compositional dependence on the grain boundary cohesive strength may be evaluated by the valence differences between two constituent atoms. As shown in Table 1, when Ni atoms substitute for Al atoms (i.e. in Ni-rich Ni3Al) this value takes the highest value of +3, while when Al atoms substitute for Ni atoms (i.e. in Al-rich Ni3Al) the value yields the lowest value of -3. On the other hand, in the case of Ni3Mn, Ni-rich Ni3Mn takes the value of +0.2 while Mn-rich Ni3Mn does the value of -0.2, implying similar grain boundary cohesive strengths at both off-stoichiometric compositions.

Similar stoichiometry effects on the grain boundary brittleness were reported in the boron-doped Ni3Al [31], and also the boron-doped L12-type alloys [32]. Thus, the control of stoichiometry should be a useful method to get a desired ductility for the material.

3.4 Effect of Interstitial Elements

Traditionally, the intergranular failure in conventional metals and alloys have been believed to be associated with significant segregation of some trace impurities to grain boundaries. On the other hand, recently, several alloys showed intergranular failure without the segregation of embrittling impurity to grain boundaries and then these intrinsically brittle alloys were ductilized by increased grain boundary cohesion resulting from segregation of particular trace elements.

Thus, the segregated atoms can be separated whether they are harmful or beneficial for the grain boundary cohesive strength. Sulfur and oxygen are embrittling elements also in the intermetallic compounds. Indeed, the segregations of sulfur [3,31] and oxygen [33] into the grain boundaries have been reported in L12-type alloys. Hence, it has to be noted that the elimination or control of these impurity elements is still an important technical method for developing ductile intermetallic compound. Boron was first shown to be a beneficial segregant in the tensile duc-
tibility of polycrystalline cast Ni$_3$Al [34] (Fig. 8). The ductility increase was associated with a dramatic change in fracture mode, from intergranular fracture in the undoped Ni$_3$Al, to almost completely transgranular fracture in the doped Ni$_3$Al. Subsequently, ductility improvement was observed for contents from 0.02 wt.% to about 0.2 wt.% boron additions, and was attributed to boron segregation to grain boundaries [31]. Recently, it was shown that the doping of a small amount of beryllium element into Ni$_3$Al has similar ductilization effect although the increased ductilities were less than 10 pct in tensile elongation [35] and also this element behaves as substitutional in the grain interior [36].

The fact that boron acts to increase the cohesive strength of grain boundaries in Ni$_3$Al by segregation to grain boundaries is widely accepted. However, models which are proposed for this effect are very conflicting. Liu et al [31], based on the thermodynamics for unusual segregation behavior of boron of grain boundary plane relative to free surface, proposed the enhanced grain boundary cohesive strength in boron-doped Ni$_3$Al. Schulson et al [37,38] proposed that the role of boron is to ease the transmission of slip between grains and hence eliminate crack initiation at grain boundaries.

Electronic considerations could again be responsible for the effect of interstitial atoms, boron although the atomic structure and resultant electronic structure at grain boundaries segregated by these atoms are not so easily drawn in a geometrical frame. It has been suggested that in pure nickel, boron atoms act as electronic donors and thereby strengthen the atomic bond [12,13]. Consequently, it is very likely that boron segregated at grain boundaries of Ni$_3$Al modifies the electronic environment of grain boundaries and enhances the grain boundary cohesion. However, there are still unratified experimental results for these proposed models. One of them is the interaction between alloy stoichiometry and the doping of B atoms; a small amount of boron addition into Ni$_3$Al was effective only at Ni-excess composition [34,35], as shown in Fig. 7.

The C atom has a similar electronic structure to B atom because of its similar position in the periodic table. Indeed, the segregation of C atom to the grain boundary planes in rapidly solidified Ni$_3$Al was detected by AES [39]. However, the experimental result indicated no improvement of ductility in C-doped Ni$_3$Al [39]. Thus, carbon has not been characterized as a harmful or a beneficial species.

3.5 Hydrogen and Oxygen Embrittlement (Environmental Effect)

Concentrated efforts to find intrinsically ductile intermetallic compounds and to make the intermetallic compounds ductile have resulted in some fabricable intermetallic compounds. However, another problem has been shown recently; the grain boundaries in ductilized compounds were usually very susceptible to hydrogen and oxygen embrittlement.

Hydrogen injected by electrolytic charging or hydrogen gas exposure led to brittleness by assisting the intergranular failure. Increased embrittlement evaluated by tensile testing has been demonstrated in the L$_1_2$-type (Fe,Ni)$_3$V [40], the orthorhombic Ni$_5$Cr [41] and the Ni$_3$Al micro-alloyed with boron [42]. Recent works for Co$_3$Ti [19] and Ni$_5$(Al,Mn) [23] also exhibited that their tensile ductilities strongly depend upon the test environment, with much lower ductilities observed in air than in vacuum, and of strain rate, with lower ductilities observed at low strain rate than at rapid strain rate, as shown in Fig. 9. Similarly, the marked test environment effect on the ductilities was observed in the Ni$_3$Al doped with Be [35] and in Co$_3$Ti alloyed with the ternary elements [24] (see Fig. 6). The embrittlement was generally shown to be reversible after a degassing treatment
4. SUMMARY AND REMARKS

Therefore, the reduced ductility could not be attributed to the introduction into the sample interior of such permanent damage as hydride or swelling voids. From these documented cases, it is likely that hydrogen could be injected from outer circumstance to the sample interior and also removed from the sample interior by vacuum pumping. The results that the samples tested in air and hydrogen-charged conditions usually showed more brittle intergranular failure along the sample perimeter and that the ductilities were very sensitive to the test strain rate suggest that the samples are embrittled by the "dynamic" effect involving weakening of atomic bonds across grain boundaries as a result of hydrogen adsorption and high localized-stress concentration at a propagating crack tip [19].

Although more research is needed to characterize this phenomenon, the details of dynamic intergranular embrittlement due to the hydrogen would be understood again based on the electronic concept for the effect of hydrogen atom on the grain boundary cohesive strength.

From the point of view of controlling (or suppressing) the hydrogen embrittlement in the intermetallic compounds, some useful conclusions can be stated. The first example is that boron-doped Ni3Al which was preoxidized at a high temperature showed improved ductility [42]. A thin adherent oxide film was suggested to be responsible for this modification. The second example is that the elements of Fe and Al were the most resistant to the hydrogen embrittlement in the ternary Co3TiX alloys, as shown in Fig. 6 [24], although its mechanism has not been clarified as yet.

Recently, the oxygen-related grain boundary embrittlement has been reported in the boron-doped Ni3Al alloys [44-46]. The tensile ductilities measured at high temperatures were found to be a strong function of test environment, i.e. whether air or vacuum. The samples tested in air showed severe loss of the ductility. Also, the occurrence of the strain rate effect on this oxygen-related embrittlement was phenomenologically very similar to the hydrogen gas embrittlement [46].

4. SUMMARY AND REMARKS

In this paper, the phenomenological features for the grain boundary brittleness of the L12-type intermetallic compounds were reviewed with respect to the combination of the constituent atoms in the alloys, alloying effects, stoichiometry effects, and the role of interstitial and gaseous atoms. Based on these systematic observations, it was suggested that the brittleness of grain boundaries of these L12-type intermetallic compounds can be manipulated by appropriate control of the chemical bond environment of the boundaries through selecting proper alloy composition, and by control of test condition.

Some alloy systems ductilized by this promising principle are candidates for material development. A few pioneer attempts in this direction are being done. Indeed, the mechanical properties of the "advanced aluminide" [47], (Fe,Co,Ni)3V [22] and (Fe,Co,Ni)4Ti [48] surpassed those of the traditional high temperature alloys such as superalloys. Different efforts are also being made to improve the ductility of some intermetallic compounds through grain refining such as thermomechanical treatment [31] and rapid solidification [49]. These developments which have been successful in the laboratory scale may lead to promising transfer into the industrial scale.

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