1 of 1

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Ni₃Al ALUMINIDE ALLOYS

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Abstract

This paper provides a brief review of the recent progress in research and development of Ni₃Al and its alloys. Emphasis has been placed on understanding low ductility and brittle fracture of Ni₃Al alloys at ambient and elevated temperatures. Recent studies have resulted in identifying both intrinsic and extrinsic factors governing the fracture behavior of Ni₃Al alloys. Parallel efforts on alloy design using physical metallurgy principles have led to the development of Ni₃Al alloys with improved mechanical properties for structural use. Industrial interest in these alloys is high, and examples of industrial involvement in processing and utilization of these alloys are briefly mentioned.

Introduction

The nickel aluminide, Ni₃Al, has drawn enormous attention because of its technological and scientific interest [1-11]. Technologically, Ni₃Al is the most important strengthening constituent, generally referred to as γ phase, of commercial Ni-base superalloys used extensively as high-temperature structural material for jet-engine and aerospace applications. It is responsible for the strength and creep resistance of the superalloys at elevated temperatures. Ni₃Al containing about 25 at. % Al has the ability to form protective aluminum oxide scales, resulting in excellent oxidation resistance.

Scientifically, the aluminide shows some unusual mechanical properties, which are distinctly different from those of conventional solid-solution alloys (i.e., disordered alloys). It exhibits a strong yield anomaly behavior, that is, its yield strength increases strongly with increasing temperatures [12,13]. In contrast, conventional alloys exhibit a general decrease in yield strength with temperature. Polycrystalline Ni₃Al shows brittle grain-boundary fracture at ambient temperatures. Contrary to common experience with conventional materials, Ni₃Al grain boundaries are usually clean and free from impurities, as evidenced by Auger analyses [14]. As a result, considerable efforts have been devoted to understand the "intrinsic" grain-boundary brittleness in Ni₃Al as well as in other ordered intermetallics [15]. In addition, the physical and mechanical properties of Ni₃Al are extremely sensitive to a small change in composition around its stoichiometry [14]. The reason for this sensitivity is not well understood even at present.

In 1979, Aoki and Izumi [16] first reported that the tensile ductility of polycrystalline Ni₃Al can be effectively improved by doping with boron additions. The ductilizing effect of boron was soon confirmed by Liu et al. [14,17] and Taub et al. [18]. In the same year, Liu and Inouye [19] found that the tensile ductility of Co₃V can be dramatically increased from <1 to >40% at room temperature by alloying with iron additions, which control the ordered crystal structure and deformation behavior. These

discoveries together with the search for new high-temperature structural materials have stimulated enormous interest in Ni₃Al and other high-temperature intermetallics during the past ten years [1-11,20]. This effort has led to the understanding of brittle fracture and the development of ductile nickel, iron, and titanium aluminide alloys potentially suitable for structural applications in hostile environments at elevated temperatures.

This paper provides a brief summary of research and development of Ni₃Al alloys. In addition to the introduction section, it contains six additional sections: (1) crystal structure and lattice defect, (2) yield strength anomaly, (3) intergranular fracture and ductility improvement at ambient temperatures, (4) environmental embrittlement and ductility improvement at elevated temperatures, (5) alloy design of Ni₃Al alloys with improved properties, and (6) material processing and structural applications. Emphasis is placed on the progress made in recent years.

Crystal Structure and Lattice Defect

Long-Range Order

Ni₃Al is the first intermetallic phase formed at the nickel-rich end of the Ni-Al alloy system. It has the L₁₂ ordered structure below the peritectic temperature of 1395°C [21]. Its unit cell contains four atoms, with three nickel atoms occupying face-centered cubic (fcc) positions and one aluminum atom at corner positions. The L₁₂ crystal structure can be viewed as stacking of three close-packed planes in the stacking sequence abc. Ni₃Al has a homogeneous phase field of about 3 at. % around its stoichiometric composition. With off-stoichiometric compositions, anti-site substitutions are essentially formed on both nickel- and aluminum-rich sides of Ni₃Al. This is quite different from B2-NiAl, where anti-site defects are stable on the nickel-rich side and vacancies are formed on the aluminum-rich side [21-24].

It has been established by resistivity and high-temperature X-ray diffraction that long-range order in Ni₃Al is maintained to the peritectic formation temperature [21]. Recently, Cahn et al. [25] studied the ordering behavior of Ni₃Al alloys as functions of aluminum concentration and iron additions by electrical resistivity and dilatometric measurements. Based on extrapolation of limited data, they estimated the critical ordering temperature of stoichiometric Ni₃Al to be ~1450°C (about 65°C above the solidus).

Planar Faults

The L₁₂ structure of Ni₃Al is a derivative of fcc, in which Ni-Al nearest neighbors with a higher bonding energy are maximized and Al-Al nearest neighbors with a lower bonding energy are eliminated in the perfectly ordered state. Planar faults are created in Ni₃Al by nearest-neighbor violation [i.e., anti-phase boundaries (APB)]; stacking faults [i.e., superlattice intrinsic stacking faults

(SISF) and superlattice extrinsic stacking faults (SESF); or both [i.e., complex stacking faults (CSF)].

Planar faults can exist on different planes with different energies. Since only {111} and {100} are the important slip planes in Ni₃Al, planar faults on these planes have been carefully studied, and their fault energies have been measured and calculated [26-30].

Table I summarizes APB energies obtained from binary Ni₃Al with Al ≥ 23 at. %. The APB energy on {111}, γ_1 , is higher than that on {100}, γ_0 , as predicted from considering nearest-neighbor violations. The ratio of γ_1 to γ_0 is in the range 1.1 to 1.4, which agrees well with the predictions from first-principles calculations [29]. The SISF and SESF energies in Ni₃Al are estimated to be 5 to 40 mJ/m² [26,29] and are substantially smaller than APB energies. CSFs produced by one-sixth $\langle 112 \rangle$ have the energies comparable to or larger than APB energies [29]. The planar fault energies are affected by alloying additions in Ni₃Al alloys [31,32].

The superdislocations on {111} are not stable and have a tendency to cross slip onto {100} planes in Ni₃Al at temperatures above 70 K. The dislocations dissociation on {100} is of the APB type:

$$[\bar{1}01] \rightarrow \frac{1}{2}[\bar{1}01]_{\text{APB}} + \frac{1}{2}[\bar{1}01]. \quad (3)$$

According to the dislocation modeling by Paidar et al. [37] using computer simulations, {111} screw dislocations that dissociate into superpartials according to Eq. (1), have planar cores and are always glissile. {100} screw dislocations that dissociate according to Eq. (3), on the other hand, have nonplanar cores and become sessile. The prediction of dislocation core structures in Ni₃Al has been confirmed by Crimp [38] and Sun and Hazzledine [39] using high-resolution transmission electron microscopy (TEM).

Table I. Anti-Phase Boundary (APB) Energies in Ni₃Al

APB energy (mJm ⁻²)		γ_1 / γ_0	Method	Reference
γ_1	γ_0			
180	140	1.29	Measured	Veyssire et al. [26]
110	90	1.22	Measured	Douin et al. [27]
163	122	1.34	Measured	Dimiduk [28]
190	170	1.12	Measured	Dimiduk [28]
175	140	1.25	Calculated ^a	Fu et al. [29]
283	196	1.44	Calculated ^a	Sluiter et al. [30]

^aCalculated by first-principles methods.

Superdislocations and Dissociations

The deformation in Ni₃Al is governed by gliding superdislocations (or paired dislocations), which dissociate into various superpartials, based on relative energies of planar faults [33]. The dislocation dissociation schemes commonly observed on {111} planes in Ni₃Al are of the APB type:

$$[\bar{1}01] \rightarrow \frac{1}{2}[\bar{1}01]_{\text{APB}} + \frac{1}{2}[\bar{1}01], \quad (1a)$$

$$\begin{aligned} &\rightarrow \frac{1}{6}[\bar{1}\bar{1}2]_{\text{CSF}} + \frac{1}{6}[\bar{2}11]_{\text{APB}} + \frac{1}{6}[\bar{1}\bar{1}2] \\ &\quad + \frac{1}{6}[\bar{2}11]_{\text{CSF}} \end{aligned} \quad (1b)$$

Dissociation scheme [Eq. (1a)] is commonly observed in Ni₃Al at ambient temperatures, and the scheme [Eq. (1b)] has been observed in Ni₃Al deformed at elevated temperatures (e.g., 650°C) [34]. The SISF-type dissociations, as shown in Eq. (2), are not observed in Ni₃Al but have been found in other L₁₂ intermetallics with high APB energy [35,36].

$$[\bar{1}01] \rightarrow \frac{1}{3}[\bar{2}11]_{\text{SISF}} + \frac{1}{3}[\bar{2}11], \quad (2a)$$

$$\begin{aligned} &\rightarrow \frac{1}{6}[\bar{1}\bar{1}2]_{\text{CSF}} + \frac{1}{6}[\bar{2}11]_{\text{APB}} + \frac{1}{6}[\bar{1}\bar{1}2]_{\text{SISF}} \\ &\quad + \frac{1}{6}[\bar{2}11]_{\text{APB}} + \frac{1}{6}[\bar{1}\bar{1}2]_{\text{CSF}} + \frac{1}{6}[\bar{2}11]. \end{aligned} \quad (2b)$$

Yield Strength Anomaly

Cross-Slip Models

Ni₃Al, like many other L₁₂ ordered intermetallics [33], shows a strong yield strength anomaly, starting at a temperature as low as 70 K [40]. Above this temperature, its yield strength (σ_y) increases with increasing temperature until a peak is reached (see Fig. 1), after which σ_y decreases with temperature. The peak-strength temperature and its magnitude depend on crystal orientation and the sense of applied stresses (i.e., tension versus compression) [33]. The yield anomaly behavior has been explained based on the consideration of the instability of $\langle 101 \rangle$ screw superdislocations on {111} slip planes [13,41-49]. With the assistance of thermal activation, $\langle 101 \rangle$ screw superdislocations have the tendency to cross slip from {111} to {100} planes and become less mobile. The driving forces for cross slip have been identified as APB energy anisotropy [41] and elastic anisotropy [47].

The cross-slip model was developed quantitatively by Takeuchi and Kuramoto [44] and Lall, Chin, and Pope [45], and then further refined by Paidar, Pope, and Vitek [46], generally referred as the PPV model. According to the model, the cross-slip process is thermally activated, and the number of cross-slip pinning (CSP) points [see Fig. 2(a)] increases with increasing temperature, leading to an increase in the resistance to $\langle 110 \rangle$ {111} slip. The yield strength decreases again when $\langle 110 \rangle$ {100} slip becomes feasible. Yoo [47-49] has further modified the PPV model to include a torque term derived from consideration of elastic anisotropy. He has shown that cross-slip pinning will occur only when:

$$[3A / (A + 2)](\gamma_1 / \gamma_0) > \sqrt{3}, \quad (4)$$

Figure 1: Temperature dependence of the yield strength of Ni_3Al and $\text{Ni}_3(\text{Al,Hf})$ alloys [40].

Figure 2: Illustrations of successive positions for a superdislocation moving on the (111) plane by (a) bowing between cross-slip-pinning (CSP) points as envisioned in the CSP models and (b) lateral motion of superkinks between Kear-Wilford (KW) locks as envisioned in the superkink models [50].

where $A = 2C_{44}/(C_{11} - C_{12})$ is Zener's ratio of elastic anisotropy. In the case of Ni_3Al , γ_1/γ_0 is equal to 1.1 to 1.4. Equation (4) would not be satisfied for Ni_3Al if the elastic anisotropy is not considered (i.e., $A = 1$). The Zener anisotropy factor, which is not sensitive to temperature, was calculated to be 3.3 for Ni_3Al . Setting $A = 3.3$ and $\gamma_1/\gamma_0 = 1.2$, the condition for the yield strength anomaly is well satisfied for Ni_3Al . Thus, the torque term derived from elastic anisotropy plays a major role in the CSP process and yield strength anomaly observed for Ni_3Al [47,48].

Modification of the PPV Model

The PPV model [46] with the additional torque term developed by Yoo [47-49] has successfully explained many yield and flow phenomena in Ni_3Al , including the strong orientation dependence of the yield strength and tension/compression asymmetry [44,51,52]. However, the model has been troubled with some observations such as (1) a low strain-rate sensitivity in the anomalous regime below the peak-strength temperature [13,44,53]; (2) a large discontinuity in the activation volume at a

temperature well below that of the peak [54,55]; and (3) observed superkinks between long and straight sessile crew segments, analogous to Kear-Wilford (KW) locks [42], in deformed Ni_3Al [see Fig. 2(b)] [56-59]. Recently, several suggestions have been proposed to modify the PPV model in order to explain these observations. The modifications include (1) pinning of superdislocations by KW locks [60-62], instead of localized pinning points [46] [see Fig. 2(b)]; (2) thermally- and stress-assisted unpinning processes [50,63,64] in addition to thermally-assisted pinning processes; and (3) a statistically fluctuating (or random) distribution of pinning segments [50], instead of a uniform (or regular) distribution of pinning segments [46]. With these modifications, the PPV model is able to explain the low strain-rate sensitivity, large activation volume, high work-hardening rate, and observed superkinks in deformed Ni_3Al in the temperature range where the anomalous behavior is observed. In an in situ study of deformed Ni_3Al , Couret and Caillard [65] observed a jerky motion of straight screw superdislocations with jump distance approximately equal to the dissociation spacing (APB width) of superdislocations. None of the cross-slip models is capable of explaining this observation, which should not rule out the possibility of thin foil effects in the TEM observations [66].

Alloy-Stoichiometry Effect

The yield strength of Ni_3Al is sensitive to alloying variables such as sublattice occupation of solutes, atomic-size misfit, and deviations from alloy stoichiometry. Figure 3 shows the effect of off-stoichiometry on the yield strength of Ni_3Al as a function of temperature [67]. Aluminum-rich deviations from stoichiometry produce more significant hardening and greater temperature dependence of strength than do nickel-rich deviations at low and elevated temperatures. Recently, Dimiduk [28,32] pointed out that the difference in symmetry of anti-site defects forming in aluminum- and nickel-rich alloys may affect the defect hardening in off-stoichiometric alloys. Alloy stoichiometry apparently does not affect the temperature of peak strength of Ni_3Al , located around 600°C. However, deviation from stoichiometry shifts the peak-strength temperatures of Ni_3Si , Ni_3Ge , and Ni_3Ga [68].

Solid-Solution Hardening

Figure 4 shows the plot of the solid-solution hardening rate ($d\sigma/dx$) measured at 77 K versus the lattice-parameter change rate (da/dx) in polycrystalline Ni_3Al alloys containing ternary additions [40,69]. For B-subgroup elements, a linear relationship holds well, indicating that the elastic solute-dislocation interaction [70] mainly governs the solid-solution hardening in these alloys. For transition-metal elements with relatively small atom sizes (such as Mo), their data fall into the same curve for the B-subgroup elements [24,40]; however, for the transition-metal elements with large atom sizes (such as Zr and Hf), their data

Figure 3: Effect of aluminum concentration on yield strength of Ni_3Al as a function of temperature [67].

Figure 4: Relation between the rate of solid-solution hardening at 77 K, $d\sigma_{0.2\%}/dx$, and that of lattice parameter change, da/dx , with additions of various ternary elements to $Ni_3(Al, X)$ polycrystals [69].

points fit a separate curve located above the straight line for the B-subgroup elements. One possible interpretation of this deviation is that the hardening in the transition-metal containing alloys is not controlled by the elastic solute-dislocation interaction alone. In view of the fact that the anomalous yield and flow behavior have been observed at temperatures as low as 77 K [40], it is possible that the transition-metal solutes with large atom sizes affect the superdislocation core structure and cross-slip processes and thus provide an additional hardening effect.

Like binary Ni_3Al , its ternary alloys also show an increase in yield strength with increasing temperature until a peak is reached [40,69]. This is indicated in Fig. 1 where the 0.2% yield strength is plotted as a function of temperature for Ni_3Al alloyed with hafnium. In this case, the solute is more effective in hardening Ni_3Al at elevated temperatures than at ambient temperatures. This unusual hardening behavior can not be explained from classic solid-solution hardening theories, which predict a decrease (rather than increase) with increasing temperature.

The thermally-activated cross-slip models are also used to explain the yield anomaly in Ni_3Al . It is thus logical to assume that the solid-solution hardening anomaly is related to the promotion of the cross-slip processes by ternary solutes. Recently, several studies have been focused on understanding the solid-solution hardening in single-crystal Ni_3Al alloys. Their findings indicate a lack of general correlation between the APB-energy anisotropy ratio, a driving force for cross slip of superdislocations, and the strength of ternary alloys at elevated temperatures. The torque term derived from the anisotropic elastic theory has been proposed to be the dominant driving force for screw dislocations to cross slip onto {010} planes [47,49]; nevertheless, experimental data show that the elastic properties of Ni_3Al alloys are generally insensitive to solute concentration in the range of a few percent. Based on these findings, Dimiduk [28,32] suggests that the solute may affect the activation barrier to core transformation of superdislocations, thereby promoting the cross-slip process, resulting in the solute-hardening anomaly. Further studies are required to verify this thought.

Intergranular Fracture and Ductility Improvement at Ambient Temperatures

Intrinsic Grain-Boundary Brittleness

Ni_3Al exhibits {111}<110> slip and has sufficient slip systems for extensive plastic deformation. Single crystals of Ni_3Al are highly ductile, whereas polycrystals are brittle at ambient temperatures. The brittleness originates at grain boundaries, which were considered to be intrinsically brittle mainly because no appreciable impurities at grain boundaries had been detected by Auger

analyses [14,71,72]. In 1991, Liu et al. [73,74] made a first attempt to link brittle intergranular fracture with environmental embrittlement in binary Ni_3Si and Ni_3Al . Table II shows the tensile properties of recrystallized polycrystalline Ni_3Al (23.5% Al) produced by repeated cold forging and 1000°C annealing of a cast ingot [73]. The aluminide showed only 2.5% elongation in air but 8.2% in dry oxygen, an increase in ductility by a factor of ~3. Later, George et al. [75] studied the environmental effect in polycrystalline Ni_3Al (23.4% Al) carefully prepared from recrystallization of cold-worked, single-crystal material. As shown in Table II, the polycrystalline specimens exhibited a room-temperature tensile ductility of 3.1 to 4.8% in air and 12.6 to 15.8% in dry oxygen. These results clearly demonstrate that binary Ni_3Al alloys are prone to environmental embrittlement at room temperature. Thus, the extrinsic factor—environmental embrittlement—is a major cause of low ductility and brittle intergranular fracture in binary Ni_3Al [73].

Moisture-Induced Hydrogen Embrittlement

Environmental embrittlement in aluminides, as well as other ordered intermetallics containing reactive elements (such as Al, Si, V, and Ti), is explained by the following chemical reaction [77,78]:

Aluminum atoms in the aluminides react with moisture in air, resulting in the generation of atomic hydrogen that penetrates into crack tips and causes brittle grain-boundary crack propagation and premature failure. This embrittlement is similar to hydrogen embrittlement [79-81], except that here hydrogen is generated by moisture in air, rather than hydrogen sources such as hydrogen gas or cathodically charged hydrogen. Consistent with Eq. (5), the best ductility is generally obtained in dry oxygen environments, which suppress hydrogen generation by direct formation of Al_2O_3 .

The environmental embrittlement is apparently not the sole source of grain-boundary brittleness in Ni_3Al because the elimination of the environmental effect by testing in dry oxygen does not lead to extensive ductility (e.g. > 30%) and suppression of intergranular fracture in Ni_3Al [73]. Another factor governing the brittle intergranular fracture is the poor grain-boundary cohesion due to high ordering energy and the large difference in electronegativity and valence electron between nickel and aluminum atoms. Atomistic simulation calculations suggest that the high ordering energy in Ni_3Al reduces atomic relaxation in the grain-boundary region and causes a formation of columnar "cavities" [82-85]. These cavities are expected to reduce grain-boundary cohesion and to serve as suitable sites for nucleation and growth of intergranular cracks.

It is difficult to assess the relative importance of the two causes of brittle grain-boundary fracture in Ni_3Al . Takasugi et al. [81] reported the environmental effect on ductility reduction in beryllium-doped Ni_3Al but not in binary Ni_3Al (24% Al). The Ni_3Al alloy doped with 1 at. % Be showed a tensile ductility of 5% in vacuum but 1% in air. On the other hand, no environmental effect was detected in cast binary Ni_3Al , which exhibited very limited plastic deformation at room temperature. Manganese at a level of 15 at. % was added for enhancement of the grain-boundary cohesion in Ni_3Al [86]. Masahashi et al. [86] found that the tensile ductility of $Ni_3(Al, Mn)$ is highly susceptible to environmental embrittlement at room temperature at different strain rates. Note that the environmental effect was thought previously to be associated with the ternary alloying additions added to Ni_3Al (rather than in connection with binary Ni_3Al) until the recent finding of a clear environmental effect in fabricated and recrystallized Ni_3Al and Ni_3Si [73,74].

More recently, George et al. [76] studied the environmental effect in a Zr-doped polycrystalline Ni_3Al (Ni-22.65% Al-0.26% Zr) prepared from single-crystal material. Zirconium additions were

Table II. Effect of test environment on room-temperature tensile properties of binary Ni₃Al (Ni-23.4% Al and Ni-23.5% Al) and Zr-doped Ni₃Al (Ni-22.65%-0.26% Zr) alloys

Test environment	Tensile ductility (%)	Yield strength (MPa)	Ultimate tensile strength (MPa)
Ni-23.5% Al [73]			
Air	2.5	193	230
Oxygen	8.2	194	351
Ni-23.4% Al* [75]			
Air	3.1 to 4.8	308 to 327	392 to 401
Oxygen	12.6 to 15.8	336 to 345	642 to 681
Ni-22.65% Al-0.26% Zr [76]*			
Water	6.3 to 8.7	322 to 231	473 to 528
Air	10.7 to 13.2	324 to 341	661 to 603
Oxygen	47.8 to 50.6	326 to 327	1438 to 1451

*Polycrystalline materials produced from single crystals and tested at 0 and 45° to the rolling direction.

reported to improve the room-temperature ductility of Ni₃Al [86]. Table II also shows the room-temperature tensile ductility of the Zr-doped Ni₃Al alloys tested in a different environment. The polycrystalline specimens exhibited tensile ductilities of 6.3 to 8.7% in water, 10.7 to 13.2% in air, and 47.8 to 50.6% in dry oxygen. These results demonstrate that, by adding zirconium to possibly eliminate the intrinsic source of grain-boundary brittleness, the Ni₃Al alloy becomes very sensitive to test environment. The ductility increases from 6 to 50% by testing the alloy in a dry environment. Moisture-induced hydrogen environment has been observed in many other L1₂ intermetallics [88].

Beneficial Effects of Boron

Boron has been found to be most effective in improving the tensile ductility of Ni₃Al (<25% Al) tested in air at room temperature [14,16]. In view of the recent finding of environmental embrittlement, the ductilizing effect of boron was reassessed by testing boron-doped Ni₃Al (24% Al) in various environments. As shown in Table III [89], all specimens, including the one tested in water, are ductile with tensile elongation more than 35% and fracture mainly by a transgranular mode. These results indicate that boron-doped Ni₃Al is not sensitive to test environment at room temperature, consistent with the earlier data reported by Takasugi et al. [81]. Thus, boron is effective in eliminating environmental embrittlement in Ni₃Al. Since boron and hydrogen both occupy interstitial sites, it is reasonable to assume that the strong segregation of boron to Ni₃Al grain boundaries would reduce the diffusivity of hydrogen along the boundaries and thus alleviate hydrogen embrittlement. Boron-free Ni₃Al (24% Al) showed a ductility of only 8% in dry oxygen while boron-doped

Ni₃Al exhibited >40%. This comparison suggests that boron segregation also enhances the intrinsic grain-boundary properties of Ni₃Al. Atomistic-simulation calculations by Chen et al. confirm the increase in grain-boundary cohesion by segregation to defects at Ni₃Al grain boundaries [90].

Improvement in Room-Temperature Ductility

Low ductility and brittle grain-boundary fracture are major problems for structural use of polycrystalline Ni₃Al. In recent years, substantial progress has been made in improving the ductility and fracture resistance of Ni₃Al. Table IV summarizes the effect of alloying additions on the room-temperature ductility and fracture behavior of Ni₃Al alloys. Boron is most effective in suppressing grain-boundary fracture and improving tensile ductility of Ni₃Al in air [14,16-18]. It is interesting to note that another interstitial element, carbon, does not produce any beneficial effect in Ni₃Al [91], possibly because of no strong segregation to Ni₃Al grain boundaries.

Takasugi and Izumi [92,93] conducted a systematic study of the effect of substitutional elements on mechanical properties of Ni₃Al. They found that alloying with iron and manganese to replace aluminum significantly improves the tensile ductility and lowers the propensity for grain-boundary fracture in Ni₃Al. The beneficial effect of iron and manganese was attributed to their reduction of the average valency and atom-size differences between nickel and "modified aluminum" atoms, thereby increasing the grain-boundary cohesion. Dimiduk et al. [94], on the other hand, showed that iron improved the ductility only of two-phase alloys but not single-phase alloys. Ni₃(Al, Fe) doped with boron had excellent tensile ductilities in air at room temperature.

Table III. Effect of boron addition on tensile properties of Ni₃Al (24% Al) tested at room temperature in different environments [89]

Test environment	Elongation (%)	Strength (MPa)	
		Yield	Ultimate
Oxygen	42.8	289	1315
Air	39.3	280	1241
Water	36.8	288	1199

Table IV. Effect of alloying addition on room-temperature ductility and fracture behavior of Ni₃Al alloys prepared by conventional melting and casting

Alloying element	Alloy composition (at. %)	Tensile ductile (%)	Fracture mode
	Ni ₃ Al (24% Al)	1 to 3	Intergranular
B	Ni-24 Al-0.5 B	35 to 54	Transgranular
B, Fe	Ni-20 Al-10 Fe-0.2 B	50	Transgranular
Mn	Ni-16 Al-9 Mn	16	Transgranular
Fe	Ni-10 Al-15 Fe	8	Mixed
Pd	Ni-23 Al-2 Pd	11	Intergranular
Pt	Ni-23 Al-2 Pt	5	Intergranular
Co	Ni-23 Al-2 Co	4	Intergranular
Cu	Ni-23 Al-2 Cu	6	Intergranular
Zr	Ni-22.65Al-0.26 Zr	13	Intergranular

Zirconium additions also gave a moderate improvement in the ductility of Ni₃Al [76,87]. Note that Ni₃(Al, Mn) [86] and Ni₃Al+Zr [76] showed moisture-induced hydrogen embrittlement at room temperature, indicating that Mn and Zr do not effectively reduce hydrogen penetration along the grain boundary. The beneficial effect of B, Fe, and Mn has also been observed in Ni₃Al alloys prepared by rapid solidification using melt-spinning [18,95,96] or the in-rotating-water method [97,98].

Recently, Chiba, Hanada, and Watanabe [99-102] found that the tensile ductility and cold fabricability of Ni₃Al were improved by alloying with small amounts (1 to 3 at. %) of Pd, Pt, Cu, and Co, and 0.25% Ag. For instance, the study by Chiba et al. [99] indicated that palladium has a strong tendency to occupy nickel sites, independent of composition of the host alloys. For instance, the alloy Ni₇₅Pd₂₅Al₂₃ was successfully cold-rolled by 10 to 40% reduction without cracking, and it exhibited a tensile elongation of 11% at room temperature. The results cannot be rationalized based on consideration of the electronegativity or valency difference; instead, they can be explained in terms of reducing the ordering energy by palladium additions. The ordering-energy consideration is supported by observation of APBs in palladium-doped Ni₃Al but not in palladium-free Ni₃Al. In comparison, Pt, Cu, and Co are less effective in improving the ductility of Ni₃Al. The effect of these elements on environmental embrittlement has not been studied so far.

Directionally-Solidified Ni₃Al Alloys

The ductility of binary Ni₃Al alloys can be substantially improved by control of grain shape. Liu [103] first reported that the tensile elongation of cast Ni₃Al (24 at. % Al) increased from 1.2 to 14.1% with a change in grain shape from equiaxed to columnar. In this case, a columnar grain structure with a strong <100> texture was produced by a directional levitation-zone-remelting technique. The beneficial effect of the columnar grain structure tested along the growth direction is attributed to minimizing the normal stress on boundaries, thus reducing nucleation and propagation of cracks along the boundaries [104]. Recently, Hirano and his associates [105-107] reported even higher tensile ductilities of stoichiometric and hyperstoichiometric Ni₃Al alloys of columnar grain structures with weak <100>+<111> or <110>+<111> textures (Fig. 5). For instance, a tensile ductility of >100% was obtained in stoichiometric Ni₃Al tested along the growth direction. Surprisingly, the same material tested perpendicular to the growth direction also showed high ductility (28 to 31%) and transgranular fracture. These striking results can be rationalized from analyses of grain-boundary character, which shows very high frequency of low-energy boundaries, such as 33.3% of Σ1 (low-angle boundaries), 28.5% of Σ3, 7.3% of Σ9, and 1% of Σ19 [108].

Figure 5: Room-temperature stress-strain curves of directionally solidified Ni₃Al alloys grown by a floating-zone method [107].

Environmental Embrittlement and Ductility Improvement at Elevated Temperatures

Test environments affect the ductility and fracture in polycrystalline Ni₃Al alloys not only at ambient temperatures but also at elevated temperatures. As discussed in the foregoing section, ambient-temperature embrittlement is associated with hydrogen released from the reaction of moisture in air with aluminum atoms in Ni₃Al. The embrittling agent at elevated temperatures is oxygen that penetrates along grain boundaries and causes brittle intergranular fracture [109-111]. Single crystals of Ni₃Al also show a ductility minimum at intermediate temperatures [13], which may also be related to environmental embrittlement. However, the loss in ductility has not been studied in detail at the present time.

Effects of Test Temperature and Aluminum Concentration

Environmental embrittlement of Ni₃Al alloys is strongly affected by test temperature and aluminum concentration [111]. Figure 6 shows the tensile ductility of an Ni₃Al alloy (Ni-21.5Al-0.5 Hf-0.1B, at. %) as functions of test temperature and environment (air versus vacuum). The alloy shows distinctly lower ductility in air than in vacuum (10⁻³ Pa) above 300°C, and severe embrittlement occurs in the range 600 to 850°C. The similarity in the profile of the two curves in Fig. 6 suggests that a conventional vacuum in the range of 10⁻³ Pa is not good enough to completely suppress the environmental embrittlement. The environmental effect in oxygen-containing environments is vividly demonstrated in the study of the tensile ductility of a higher aluminum alloy (Ni-23Al-0.5Hf-0.07B) tested at 760°C as a function of air pressure [33]. The alloy exhibits only about 1% elongation when tested in air;

Figure 6: Comparison of tensile elongation of Ni-21.5Al-0.5Hf-0.1B (at. %) tested in vacuum and air [111].

however, the elongation increases steadily with decreasing air pressure in a vacuum system, and it reaches 25% at air pressures below 10^{-3} Pa.

The embrittlement is sensitive to aluminum concentration in Ni₃Al alloys containing 21 to 24% Al [111]. All the alloys exhibit an excellent (>45%) ductility at 600°C when tested in vacuum, whereas the 600°C ductility is dramatically lowered in air. The corresponding fracture mode changes from ductile transgranular in vacuum to brittle intergranular in air. In both cases, the alloys show small but steady increases in ductility with decreasing aluminum concentration. The alloys are less embrittled in air if they have lower levels of aluminum. Alloy specimens with fine grain sizes exhibit better ductility when preoxidized at 1100°C.

Again, the ductility of preoxidized specimens increases with decreasing aluminum concentration [109]. Alloy specimens with coarse grain sizes (e.g., 200 μ m), however, show embrittlement rather than improvement by preoxidation. Environmental embrittlement was observed in spray-formed Ni₆₆Co₁₀Al₂₄ alloy containing 0.25 at. % B [112].

The environment affects not only tensile properties but also fatigue behavior. Figure 7 is a plot of the fatigue life as a function of temperature for a boron-doped Ni₃Al (24 at. % Al) prepared by powder metallurgy [113]. The alloy showed a sharp drop in fatigue life above 500°C. Note that the fatigue tests were conducted in vacuum (10^{-3} Pa), suggesting again that a conventional vacuum is not good enough to suppress the embrittlement in alloys containing as high an aluminum level as 24 at. %. The drop in the fatigue life is accompanied by a change in fracture mode from transgranular to intergranular.

Figure 7: Change in high-cycle fatigue life with temperature for Ni₃Al doped with boron and tested in vacuum [113].

Dynamic Embrittling Mechanism

The embrittlement described here is mainly caused by a dynamic effect of simultaneous, high localized stress concentration; elevated temperature; and gaseous oxygen [111]. Although the embrittlement takes place when samples are exposed in oxidizing environments, it is not necessarily associated with formation of oxide scales on specimen surfaces or with grain-boundary oxidation. As mentioned before, preoxidation can cause a slight increase rather than a decrease in ductility of fine-grained Ni₃Al when tested in air at 600°C. Also, alloy specimens that fractured in a brittle manner at elevated temperatures (600 to 800°C) in air remain ductile in subsequent bend tests at ambient temperatures. Such a dynamic effect basically involves repeatedly weakening and cracking of the grain boundary as a result of oxygen absorption and penetration at crack tips. The reduction in the ductility of Ni₃Al alloys is always accompanied by a change in fracture mode from ductile transgranular to brittle intergranular.

Recently, Hipsley and DeVan [114] proposed a model of stress-assisted grain-boundary oxygen penetration to explain the dynamic embrittlement of Ni₃Al in oxidizing environments at elevated temperatures. This model is similar to the stress-assisted grain-boundary oxidation (SAGBO) suggested for oxygen embrittlement [115] in superalloys. According to their model (Fig. 8), dynamic embrittlement involves the following four sequential steps: (1) occurrence of surface cracks at the initial stage of deformation; (2) chemisorption of gaseous oxygen to the crack tips where a high localized stress field is involved; (3) oxygen penetration in its atomistic form to the stress field ahead of tips; and (4) inward development of surface cracks preferentially along the grain boundaries, leaving some secondary cracks. Steps (2) to (4) proceed continuously and repeatedly during deformation, leading to premature fracture and severe loss in ductility at elevated temperatures.

Improvement in Elevated-Temperature Ductility

Oxygen-induced embrittlement is the main cause of the low ductility and brittle fracture in Ni₃Al at elevated temperatures. This problem has to be overcome in order to use Ni₃Al alloys in oxidizing environments. Recent studies show that the embrittlement can be alleviated or reduced by (1) control of surface conditions, (2) control of grain shape, and (3) alloy additions. Control of surface conditions is a simple way to

Figure 8: Schematic illustration of steps in proposed fracture mechanism of dynamic embrittlement at elevated temperatures [114].

alleviate environmental embrittlement from surface reactions. In several cases, preoxidation to form protective oxide scales was proven to be beneficial in reducing oxygen embrittlement at elevated temperatures [111]. Unfortunately, the oxide films are cracked by stretching a couple of percents, and their protective effect disappears. Surface coating should also be useful in protecting underlying materials; however, this effect has not yet been well demonstrated.

Columnar-grained structures reduce environmental embrittlement effectively in Ni_3Al at elevated temperatures [104]. They can be produced by directional solidification, for example, using levitation-zone remelting. Table V compares the tensile properties of boron-doped Ni_3Al with equiaxed and elongated grain structures tested at 600 and 760°C in air or vacuum. The specimens with equiaxed grain structure showed distinctly lower ductility in air, but good ductility in vacuum, while the specimens with the columnar grain structure exhibited good tensile ductility in air. The beneficial effect of the columnar-grained structure with grain boundaries parallel to the stress axis is attributed to minimizing the normal stress across the boundaries and consequently suppressing nucleation and propagation of cracks along the boundaries weakened by oxygen penetration.

Figure 9: Effect of chromium additions on tensile elongation of Ni_3Al alloys tested at room temperature to 1000°C in air [110].

In these aluminide alloys, chromium at a level of 6 to 9% is added for reducing environmental embrittlement in oxidizing environments at elevated temperatures. Zirconium or hafnium additions are most effective in improving the high-temperature

Table V. Tensile elongations of boron-doped Ni_3Al with equiaxed grains and columnar grains tested at 600 and 760°C in air and vacuum [104]

Grain shape	Tensile elongation (%)			
	600°C		760°C	
	Air	Vacuum	Air	Vacuum
Equiaxed grains	0.6	45	0.7	11
Columnar grains	33	.*	7.3	.*

*No data available.

Liu and Sikka [110] found that chromium is effective in alleviating elevated-temperature embrittlement in oxidizing environments. Figure 9 shows that alloying Ni_3Al with 8 at. % Cr substantially reduces the environmental embrittlement at 600 to 800°C in air. The chromium-containing alloys, however, exhibit better ductility in vacuum than in air, indicating that this embrittlement is not completely eliminated by chromium additions. The beneficial effect of chromium is considered to be associated with self-healing films of chromium oxides that reduce the penetration of oxygen into grain boundaries. The chromium effect is sensitive to composition, with the best result from alloys containing 10 to 15 vol. % of the disordered γ phase.

Alloy Design of Ni_3Al Alloys with Improved Properties

Mechanical properties of Ni_3Al depend on deviation from stoichiometry and alloying additions. The aluminide is capable of dissolving a substantial amount of alloying elements, and its mechanical and metallurgical properties can be improved through control of solute concentration and second-phase formation.

Alloy Design of Polycrystalline Material

Recent efforts on alloy design have led to the development of Ni_3Al -base alloys with the following composition range for structural use at elevated temperatures in hostile environments [110,116,117]: Ni-14 to 18Al-6 to 9Cr-1 to 4Mo-0.1 to 1.5Zr/Hf-0.01 to 0.20B (at. %).

strength via solid-solution hardening effects. Molybdenum additions are used for improving strength at ambient and elevated temperatures. Microalloying with boron reduces moisture-induced hydrogen embrittlement and enhances grain-boundary cohesive strength, sharply increasing the ductility at ambient temperatures. In some cases, certain amounts (<20%) of cobalt and iron are added to replace Ni or Al and Ni, respectively, in order to further gain in hardening and corrosion resistance. The alloys with optimum properties usually contain a certain amount of second phases, such as the disordered γ phase, for further improving mechanical properties. For instance, the γ phase at a level of 5 to 15 vol. % reduces environmental embrittlement in oxidizing atmospheres and improves creep properties at elevated temperatures.

Cast aluminide alloys usually possess a coarse grain structure that lowers the yield strength at ambient temperatures. The strength can be increased by alloying with molybdenum via solid-solution hardening [117]. Figure 10 compares the tensile properties of a cast Ni_3Al alloy (Ni-16.0Al-8.0Cr-1.7Mo-0.5Zr-0.03B at. %) with commercial alloys, such as the cast superalloy IN-713C that is commonly used at high temperatures by industry. The strength of the advanced aluminide alloy is comparable to that of IN-713C at room temperature, but it is higher at elevated temperatures because of the significant increase in yield strength with temperature. The aluminide alloys prepared by investment casting with minimum cast defects showed excellent fatigue resistance. The high-cycle fatigue life of an aluminide alloy (IC-221M: Ni-15.9Al-8.0Cr-0.8Mo-1.0Zr-0.03B, at. %) is higher than that of IN-713C by more than two orders of magnitude at 650°C in air [118]. The cast aluminide alloys usually have tensile elongations of 10 to 30% at room and elevated temperatures.

Figure 10: Comparison of yield strength of Ni_3Al alloys (advanced aluminide alloy and $\text{Ni}_3\text{Al} + 0.2 \text{ B}$) with commercial alloys (IN 713C, Hastelloy X and 316 stainless steel).

The aluminide alloys with the compositions $\text{Ni}-16 \pm 2\% \text{ Al}-6 \pm 2\% \text{ Cr}-0.6 \pm 0.5\% \text{ Zr}-0.10 \pm 0.09\% \text{ B}$ (at. %) can be fabricated by cold-rolling or hot extrusion. The wrought alloys with fine-grain structures have excellent strengths at ambient temperatures ($\sigma_y > 600 \text{ MPa}$) but lower creep resistance because of grain-boundary sliding [119] above 700°C . The alloys generally showed good fatigue properties and crack-growth resistance [120]. The fatigue and crack-growth behaviors of Ni_3Al alloys are sensitive to test environment; nevertheless, the aluminide alloys containing chromium show fatigue-crack resistance better than chromium-free alloys when tested in air at elevated temperatures.

Alloy Design of Single Crystals and Directionally Solidified Materials

Polycrystalline Ni_3Al alloys, just like polycrystalline nickel-base superalloys, are not strong in creep because of the grain-boundary sliding problem at high temperatures ($>0.5 T_m$). An effective way to eliminate or reduce the grain-boundary weakness is to produce single-crystal materials or columnar grain structures by directional solidification. At present, limited effort has been devoted to design of single-crystal Ni_3Al alloys for structural applications [121-125].

Effects of alloy stoichiometry and ternary alloy additions on the creep behavior have been studied for single-crystal, single-phase Ni_3Al alloys. The alloys containing Hf, Ta, Ti, and Cr additions often show an inverse type of creep curve, i.e. low initial creep rates followed by accelerated creep deformation [122,126,127]. All these alloying elements substantially reduce the creep rate and extend the rupture life of Ni_3Al , and the solid-solution strengthening is more effective in nickel-rich alloys than for aluminum-rich alloys. TEM studies reveal that the creep deformation at the first and second creep stages is mainly controlled by viscous gliding of superdislocations with a stress exponent (n) of about three.

Shah and Duhal [121,125] attempted to design single-phase Ni_3Al alloys based on γ compositions [e.g., $\text{Ni}-14.40\text{Al}-5.19\text{Ta}-2.73\text{Co}-2.52\text{Ti}-0.89\text{W}-1.94\text{Cr}$ (at. %) of PWA 1480, a high-strength, single-crystal nickel-base superalloy]. The creep properties were evaluated as functions of Ni/Al ratio, tantalum concentration, and other additions of refractory elements. The multi-component γ alloys are much stronger than unalloyed Ni_3Al ; nevertheless, they provide no significant advantages over the precipitation-hardened superalloys. In addition, the single-phased Ni_3Al alloys showed relatively lower impact resistance at ambient temperatures (in particular, for aluminum-rich compositions).

Han et al. [124] have recently developed a strong Ni_3Al alloy prepared by directional solidification (DS). The alloy (density = 7.91 g/cm^3) has a simple composition $\text{Ni}-16.3\text{Al}-8.2\text{Mo}-0.2\text{B}$ (at. %), with molybdenum as the major solid-solution strengthener. It shows a yield strength of 990 MPa at 700°C , 600 MPa at 1000°C , and 520 MPa at 1050°C , all of which appear to be higher than those of strong nickel-base superalloys in single-crystal or DS forms. The creep resistance of this alloy is better than that of PWA 1422, particularly at temperatures above 1000°C . The excellent mechanical properties of the Ni_3Al alloy are attributed mainly to solid-solution hardening by molybdenum and second-phase strengthening by 15 to 20% γ phase (with a lattice misfit of 1.24% between γ and γ' phases). This result has demonstrated the possibility of the development of promising Ni_3Al -base alloys for structural applications above 1000°C . Additional development work is certainly needed for further optimizing alloy composition, microstructure, and properties.

Material Processing and Structural Applications

The unique strength and ductility characteristics of Ni_3Al alloys present correspondingly unique challenges and opportunities in the processing of the aluminide alloys. This section includes a brief description of conventional and innovative ways of alloy preparation, cold and hot fabrication, welding and joining, and structural applications of Ni_3Al alloys.

Alloy Melting and Casting

It has been demonstrated that Ni_3Al alloys based on $\text{Ni}-14$ to $18 \text{ Al}-6$ to $9 \text{ Cr}-0$ to $4 \text{ Mo}-0.1$ to $1.0 \text{ Zr/Hf}-0.01$ to 0.2B (at. %) can be prepared by induction melting in air, arc melting in inert gas, and electron-beam melting in vacuum. Because of their high aluminum concentration, molten alloys are capable of forming a protective aluminum-oxide surface layer, which reduces the oxygen contamination to less than 50 wt ppm in air-melted materials. Large Ni_3Al alloy heats have been successfully melted using Al_2O_3 , ZrO_2 , and graphite molds [128]. The quality of primary melted ingots can be enhanced by electroslag remelting (ESR) and vacuum-arc remelting (VAR), which improve the surface condition, grain size, grain shape, and impurity levels in ingots.

Near-net-shape casting has been used to prepare Ni_3Al alloys. Alloy rods with a diameter of $<3 \text{ cm}$ and sheets with a thickness of $<3 \text{ mm}$ can be prepared by direct-casting techniques [119,128]. The directly cast materials show excellent ductility at room temperature, and surface roughness and marks due to solidification processes can be minimized simply by cold-rolling or swaging. Structural parts with complicated shapes can be fabricated by investment casting such as is commonly used for nickel-base superalloys. Figure 11 shows turbocharger rotors for diesel-engine applications made by investment casting. Hot-cracking, cast porosity, and interaction with shell materials are common problems in connection with investment casting of Ni_3Al alloys; however, these problems can be solved by controlling casting parameters and adjusting alloy composition and mold material. For instance, cast porosities in Ni_3Al alloys were minimized by increasing the cooling speed and zirconium concentration to $>0.7 \text{ at. \%}$ [117,129]. Test bars and turbine blades with single-crystal or directionally solidified structures were also successfully produced by the same industrial processes used for nickel-base superalloys.

Cold and Hot Fabrication

Cast Ni_3Al alloy ingots generally show good ductility at ambient temperatures; however, their hot ductility is sensitive to alloy composition, grain size, and test temperature [110,119,128]. In general, the hot ductility increases with decreasing zirconium concentration and refining grain structure. The Ni_3Al alloys containing $<0.37 \text{ at. \% Zr}$ can be successfully hot forged around 1050 to 1150°C [116]. Sheet fabrication by hot-rolling is possible

Figure 11: Photograph of investment-cast turbocharger rotor of Ni₃Al alloy IC-396 (Ni-16.4 at. % Al-8.0% Cr-15% Mo-0.5% Zr doped with 300 wt ppm B). Casting was carried out at Precision Castparts Corporation (PCC) [128].

but not easy even for the low-zirconium alloys. Hot extrusion at 1100-1200°C is feasible for Ni₃Al alloys containing <0.7% Zr. Because of superplastic behavior at temperatures above 900°C, isothermal forging at controlled strain rates offers excellent possibilities for hot fabrication of fine-grained Ni₃Al alloys. A maximum elongation of 638% had been reported for fine-grained IC-218 tested at 1100°C at an initial strain rate of $8.3 \times 10^{-4} \text{ s}^{-1}$ [119]. Cold fabrication is effective for materials directly cast into sheet or rod. In these cases, final products can be cold fabricated without repeated annealing and recrystallization treatments.

Powder Metallurgy

Ni₃Al alloy powders can be produced by atomization in argon or nitrogen. The powders are then canned within steel billets and hot extruded at 1100 to 1200°C to near-fully-dense bar stock with a reduction ratio ≥8:1 [116,119]. The hot extruded material generally has a fine-grained structure (10 to 20 μm) and contains a very small amount of microporosity. Instead of direct extrusion of powder, alloy compacts can be produced by consolidation under atmospheric pressure. These compacts generally contained large pores but could be further consolidated by hot extrusion. Because of the superplastic behavior, the fine-grained material is suitable for near-net shaping by isothermal forging.

Reaction Synthesis

Reaction sintering [130-133] is a novel process to produce ordered intermetallics such as Ni₃Al. With this method, it is possible to obtain dense intermetallics from elemental powders using a self-sustaining reaction (sometimes referred to as combustion synthesis or self-propagating high-temperature synthesis). Reaction sintering offers advantages such as near-net shapes, low cost, and energy savings. Powder compacts produced by cold pressing are first preheated to 300 to 650°C in controlled atmospheres (e.g., vacuum). Because of the exothermic nature of the reaction, heat is generated and a local increase in temperature triggers the combustion reaction. Aluminide alloys with a density of up to 97% of the theoretical value have been produced by optimizing reaction-synthesis parameters, including particle size, heating rate, and atmosphere [131]. The products contain, nevertheless, uniformly distributed small pores and relatively large shrinkage cavities in the center. Recently, Nishimura and Liu [132,133] have found that Ni₃Al with a high density (>99%) can be produced by reaction synthesis under compressive stresses. Reaction-synthesized Ni₃Al alloys showed tensile elongations of more than 10% at room temperature [131].

Welding and Joining

Ni₃Al-base alloys are susceptible to weld cracking; however, if done with care, sound welds can be made in most of the alloys [134-137]. Welding speed should be low, and the boron level has to be limited to less than 0.1 at. % to avoid hot-cracking. Oxygen

is particularly detrimental: oxide scales on alloy surfaces should be removed prior to welding, and the atmosphere must be controlled to reduce oxygen concentrations during welding. Certain alloying additions (iron for example) have been found to promote weldability.

Structural Applications

Industrial interest in Ni₃Al-based alloys is high at the present time [119,138]. This interest arises because the alloys possess a unique combination of properties including high strength and good oxidation and corrosion resistance at elevated temperatures, and relatively low density compared with many nickel-base, high-temperature superalloys. A detailed description of the industrial interest in Ni₃Al alloys is given by Sikka et al. [139]. Here, only a brief summary of these applications is outlined.

1. Diesel-engine turbocharger rotors: Cummins Engine Company, PCC Airfoils, and Oak Ridge National Laboratory (ORNL) are jointly pursuing the use of Ni₃Al alloys for turbocharger rotors in diesel-engine trucks [110]. The aluminide alloy IC-221M (Ni-15.9 Al-8.0 Cr-0.8 Mo-1.0 Zr-0.03 B, at. %) with a good castability has been selected for this application. Aluminide rotors have been successfully produced by investment casting at PCC (Fig. 11). This castable aluminide alloy is expected to replace IN-713C with the major benefits of improved fatigue life and potential for lower cost.
2. High-temperature dies and molds: The good high-temperature oxidation resistance, together with excellent strength at high strain rates, makes Ni₃Al alloys attractive as die material for isothermal forging and mold material for glass processing. At present, Ni₃Al alloys are being evaluated for these applications in several companies.
3. Furnace fixtures for heat treating auto parts: Because of its resistance to carburizing and oxidizing atmospheres, the Ni₃Al alloy IC-221M is currently being evaluated by General Motors (Saginaw Division) and ORNL for use as a fixture material for heat treatment of auto parts in high-temperature furnaces. The Saginaw Division has the world's largest heat-treatment facility with a heat-treatment capability of about 6000 tons per day. The HU alloy (Fe-40 Ni-20 Cr-2.5 Si-2.0 Mn-0.5 Mo-0.4 C, wt %) is currently used for furnace fixtures, which cracked badly after ~500 thermal cycles. Ni₃Al alloy furnace fixtures have been successfully made by sand casting, and the evaluation of their performance is in progress [140].
4. Rollers for steel slab heating furnaces: The high-temperature strength, together with good oxidation and corrosion resistance of Ni₃Al alloys, can produce significant savings in energy costs by not requiring water cooling and in material costs by extending the life four to six times over the current material in use [139].
5. Hydroturbines: Nickel aluminides, which offer excellent vibration cavitation resistance in water, are being considered for either repair of current hydroturbine rotors or even replacement of the entire units with these alloys.
6. Cutting tools: Nickel-aluminide-bonded tungsten-carbide composites have better low- and high-temperature strength and cutting properties than cobalt-bonded tungsten-carbide materials.
7. Turbine blades and vanes for jet engines: Directionally solidified Ni₃Al alloy with the composition of Ni-16.3Al-8.2Mo-0.26B (at. %) has superior strength and creep resistance at temperatures above 1000°C in oxidizing environments. Hence, this alloy has the potential to be used for turbine blades and valves for jet engines [124].

Concluding Remarks

For the past 15 years, considerable effort has been devoted to the study of Ni₃Al, a model material for ordered intermetallics. As a result, a great deal of knowledge has been gained in understanding the physical metallurgy and mechanical behavior of Ni₃Al alloys, especially the flow and fracture properties. Brittle failure and poor fracture resistance are the major concerns for structural use of the aluminide. Recent studies have led to identifying both intrinsic and extrinsic factors governing the fracture behavior of Ni₃Al alloys [73,76]. Several promising Ni₃Al-base alloys have been developed, and industrial interest in these alloys with improved mechanical and metallurgical properties is high. Examples of industrial involvement in processing and utilization of these alloys are briefly mentioned.

In spite of the concentrated effort on Ni₃Al and its alloys, there are areas needing additional research. For structural applications, there is room for further improvement. Suggested future research and development of Ni₃Al alloys includes: (1) mechanisms governing the solid-solution hardening behavior at ambient and elevated temperatures, (2) modification of cross-slip-pinning models in order to fully explain dislocation structures observed experimentally, (3) effect of alloy stoichiometry on moisture-induced hydrogen embrittlement and grain-boundary cohesion, (4) environmental embrittlement in single-crystal Ni₃Al alloys at elevated temperatures in oxidizing environments, (5) effects of minor additions on creep resistance and hot ductility and fabricability, (6) improvement in hot cracking and weldability of B-doped Ni₃Al alloys, (7) near-net-shape processing via reaction synthesis, (8) understanding of solidification processing and cavity formation in cast Ni₃Al alloys, and (9) development of strong single-crystal and directionally solidified Ni₃Al alloys for jet-engine applications.

At present, there is world-wide interest in research and development of ordered intermetallic alloys for structural use at elevated temperatures in hostile environments. The knowledge and experience gained from the Ni₃Al studies have been extended to other aluminides, silicides, and other intermetallics. A prominent example is the improvement in ductility of Ni₃Al alloys by boron additions [16-18], which have a similar beneficial effect in iron aluminides and other intermetallics [15,141].

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