

ROLE OF NIOBIUM IN THE PROGRESS OF GAMMA ALLOY DEVELOPMENT

Young-Won Kim

UES – Materials & Processes
4401 Dayton-Xenia Road
Dayton, OH 45432, USA

Keywords: gamma titanium aluminides, niobium, oxidation, carbon, creep, processing

Abstract

Nb has become a critically important alloying element in gamma aluminide alloys, as over the years their compositions have evolved from the initial composition of Ti-48Al-2Cr-2Nb to more recent ones based on Ti-45Al-(5-8)Nb. These changes, originally intended to raise the oxidation resistance, also resulted in increases in yield as well as creep strength levels; however, at the expense of damage tolerance. This paper discusses this complex phenomenon, along with the ways of further enhancing the resistance to oxidation as well as creep, and the processes that may improve the damage tolerance.

Introduction

Engineering gamma TiAl alloys can be grouped on the basis of their expected use temperature ranges, which have increased steadily with increases in Nb content. In the grouping, the major differences are Al and Nb contents, that is: over the years the Al content has been decreased from 48at% to 45at%, whereas the Nb content has increased from 2 to as high as 10 at%. Nb has been known to be the most effective in enhancing the oxidation resistance, and decreases in Al content may assist the Nb effect synergistically and stabilize the lamellar structures as well as raise the alloy strength levels. The compositional changes, however, reduced the material's damage tolerance, making inevitable the development of application-specific processes for finer and more uniform microstructures. Segregation of Nb around large boride needles and ribbons is another issue especially for most of alloy systems. Nb additions result in significant creep resistance improvement largely through solid solution hardening; however, more critical improvement depends on the precipitation hardening through small amounts of carbon and/or silicon additions. This paper discusses the above aspects using known information as well as experimental results.

Advances in Alloy Development

Gamma TiAl alloys, a class of intermetallic alloys based on gamma TiAl, are emerging as a pervasive structural materials technology [1, 2]. Over the years, many alloys have been developed and tested worldwide, and some representative alloys are listed chronically in Table 1. In general, the compositions of engineering gamma alloys lie within the range of Ti-(45-48)Al-(0-2)(Cr, Mn)-(1-8)Nb-xB-yC-zSi. A few observations can be made from the table. The aluminum content has been decreased while the Nb amounts have been increased, which reflects increases in use temperature. Additions of boron became common, with smaller

amounts in wrought alloys than cast alloys, as the means of grain refinement. Additions of small amounts of carbon add/or silicon also became increasingly common, which improve creep resistance additionally. The originally important alloying elements, Cr and Mn, begin to lose their importance as ductilizing alloying elements as the above compositional changes has happened; however, this empirical trends have yet to be evaluated.

Table 1. Evolution of Engineering Gamma Titanium Aluminide Alloys

Name	Composition (At %)	Uses	P	Ref
GE/ALD	Ti-48Al-2Nb-2Cr	LPT blade/Valve	C	3, 4
Howmet	Ti-(45,47)Al-2Nb-2Mn-0.8v%TiB2	Blade/Valve	C	5
GKSS/RR	Ti-47Al-1.5Nb-1Mn-1Cr-0.7 (Si,B)	HPC Blade	W	6
ABB	Ti-47Al-2W-0.5Si	LB Turbine Blade	C	7
Plansee	Ti-45.5Al-4(Cr,Nb,Ta,B)	Sheet	W	8
395MM	Ti-46Al-4(Cr, Nb, Mo)-x(B,C)	HPC Disk	W	
AF-K5	Ti-46Al-3Nb-2Cr-0.3W-0.2B-0.4(C,Si)	Disk, Valve	W	9, 10
MHI	Ti-46Al-6.5Nb-0.6Cr-0.2Ni	Turbocharger	C	11
Plansee/GKSS	Ti-45Al-5Nb-xB-yC	Valve/Sheet/Blade	W/C	12
Del West	Ti-44Al-6Nb-xW-xB	Valve	W	
AF	Ti-45Al-xNb-y(W,Hf,Mo)-z(B,C,Si)	Oxidation	W	

P: Processing; C: Cast; W: Wrought

The compositional changes in Table 1 reflect advances made in the temperature capability that is estimated by their resistance to oxidation and creep deformation. The alloys can then be classified into five groups in terms of use temperature, as listed below.

Group A: Ti-48Al-2Cr-2Nb (called 4822)

Ti-(45, 47)Al-2Mn-2Nb-0.9vol%TiB2 (called XD)

Group B: Ti-(45-47)Al-(1-2)Cr-(1-2.5)Nb-(0-2)(W, Mo, Ta)-(0-1)B-(0, 0.2)C

Group C: Ti-(45.5-46.2)Al-(1-2)Cr-3Nb-0.2W-xB-yC-zSi (K5 alloys)

Group D: Ti-45Al-(5-8)Nb-xB-yC

Group E: Ti-(44-45.5)Al-(0-1)Cr-(5-7)Nb-wRE-xB-yC-zSi

Group A alloys are essentially casting alloys with two distinguished microstructures: coarse casting duplex for 4822 alloy and refined nearly fully-lamellar for XD alloys. These are the first engineering alloys that received extensive evaluation during early 1990's. Industry focus has been on low cost implementation of these alloys for selected low-risk and low temperature applications in gas-turbine (low pressure turbine blades) and automotive (valves and turbochargers) engines [1, 3, 5]. Available data (oxidation resistance and creep properties) and engine test experience indicate that their long-term use may be appropriate at temperatures lower than 700°C. Nevertheless, these alloys are still under investigation for both applications [4]. These alloys were used worldwide for various basic and applied researches, which provided the basis and base information for fervent alloy design activities that followed.

Class B alloys are represented by TAB (GKSS) and 395MM (PW/AF) alloys that are essentially wrought alloys, whose final microstructures are fully lamellar that are introduced by a final alpha heat treatment followed by controlled cooling. The isothermally shape-forged and alpha heat-treated TAB alloy blades have been tested for high-pressure compressor (HPC) blade applications [13]. Large- to full-scale alloy 395MM forgings were successfully produced through PAM ingot metallurgy (IM) as well as powder metallurgy (PM) routes [14]. Recently, the largest forging, weighing 210kg, ever made was produced through scaled-up PM processes, as exemplified in Figure 1. The technological progress in these areas have been remarkable in recent years thanks to the development of scaled-up processing for controlled wrought products on a large scale and advances in life prediction methodologies. Extensive efforts in characterizing defects (segregation, melting defects and inclusions) in forged 395MM disks have been made to understand the causes and to find ways of minimizing as well as handling them. The findings and understanding from these efforts are and will be significant for establishing the wrought gamma alloy technology. These B alloys have greater strength levels, slightly improved oxidation resistance and significantly better creep resistance as compared with Group A alloys. Nevertheless, their long-term exposure use temperatures for rotational components are not expected to be not much higher than 730°C. Therefore, it can be said that these alloys have played transitional (but important) roles as pathways toward the development of improved processing and higher temperature alloy materials. ABB alloy is a cast alloy also and had been extensively evaluated and tested for land-base turbine engine blade applications in a nearly-lamellar form [7]. Due to the high W addition and a substantial Si addition, this alloy has significantly improved creep and oxidation resistance. This in turn reduced its damage tolerance substantially, which is an overriding concern for the alloy.

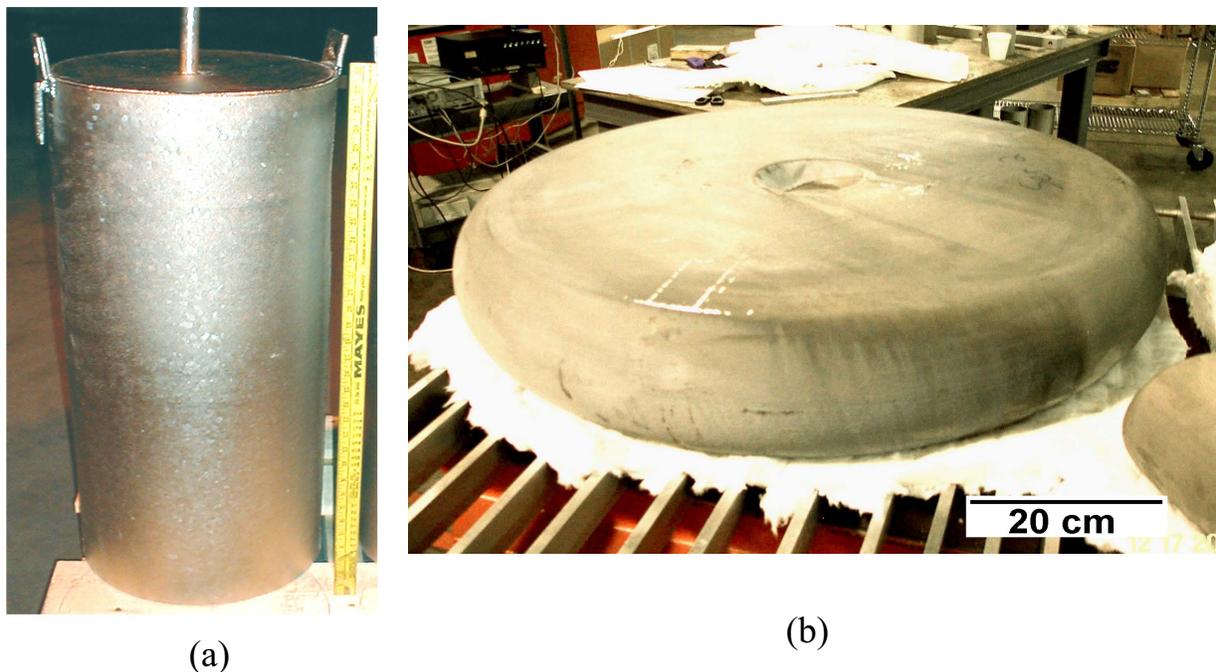


Figure 1. Powder metallurgy processing of full-scale (210kg) forged disks of alloy 395MM, involving (a) powder production and billet preparation (Crucible) and (b) isothermal forging (Ladish), followed by heat-treatments to produce the final microstructure of interest. The significance was that the complex heat treatment cycle was determined through a modeling process combining experimental data/information obtained from coupon specimens [AF/PW].

As higher performance/efficiency turbine engines and high temperature thermal protection systems (TPS) are increasingly needed or demanded and being designed as such, conventional gamma alloys (group A and B) often miss the opportunities of competition even for the components whose property requirements are ideally suitable for gamma alloys other than temperature requirements. The potential high-payoff higher-risk applications exist especially in wrought alloys, in such areas as compressor rotor disks, compressor blades, exhaust nozzle thin/large structures, space-vehicle skin structures including TPS, and high performance automotive exhaust valves. Accelerated efforts have been made worldwide in designing high temperature wrought gamma alloys [15-17].

The main thrust in high temperature alloy design at AFRL/UES was to enhance the resistance to high cycle fatigue deformation and failure, creep deformation and oxidation, and much of this has been accomplished through chemistry modification, refinement of lamellar structures and grains, process development, and/or their combinations. These efforts led to the introduction of a series of Group C alloys (K5 Series) that have potential to be used at temperatures up to 800°C for a long-term under given application stress conditions [9,10]. This increase in use temperature at least by 60°C is remarkable, and is due to the improvements of both oxidation as well as creep resistance. A small increase in Nb with a minor amount of W addition (3Nb-0.2W in at%) is apparently responsible for the slow rate with a parabolic kinetics, as will be shown later [18]. The apparent synergistic effect between Nb and W is significant but not understood. The original alloy K5, with no carbon and Si additions, was the first (engineering) alloy that was developed specifically for balanced properties in fully-lamellar forms through grain refinement by the alpha + beta treatment at first and by adding small amounts of boron [9]. The standard amount of boron addition (0.2at%) commonly used for most of current wrought alloys was determined during the development of K5 alloys by systematic boron-GS relationship studies [19]. Later on, carbon and silicon were added to improve the creep resistance, which has been achieved by controlled aging, which produces carbides and silicides [15]. These alloys, called K5BCS, exhibit dramatically increased resistance to both primary and secondary (minimum rate) creep deformation when properly aged after lamellar microstructure formation treatments especially [15]. However, there is growing evidence that further improvements of the creep resistance is possible when the alloy-specific optimum amounts and distribution of the particles and systematically understood and quantified. These complex phenomena are discussed in the next section. As is also discussed later, various processing routes have been investigated and developed to improve the ductility and strength of these alloys. It is important to note, however, that all achievements and developments described above have been made on materials from small to medium size ingots or billets.

Group D alloys have been investigated and developed worldwide for the last several years in order to capitalize their high strength and oxidation resistance [11, 15, 16, 17, 20]. As the results, they are emerging as an important engineering gamma alloy class for expected high-temperature structural applications. Some alloys based on of Ti-45Al-5Nb began to be used for high performance automotive valve applications, as well as advanced sheet material fabrication [12]. It is noted that even within Group D composition range, increases in Nb contents appear to gradually increase the oxidation resistance [15, 18]. A selection of 5at% Nb containing alloys is an apparent compromise among several factors, including oxidation resistance, density, processibility and damage tolerance. Alloys containing higher Nb amounts (6-10 at%) have been further investigated in order to capitalize on several attributes including additional oxidation resistance improvement, improved wear resistance and higher density, the latter of which is needed for racing car engine valve applications along with a lower modulus.

These alloys, however, show an unusual oxidation behavior, that is: a period of rapid oxidation during cyclic exposure at 870°C, as discussed later. As was mentioned earlier, these alloys have substantially increased strength levels, which have been explained by either solid solution or structural refinement. As discussed later, our understanding of the strengthening mechanisms is far from being complete. Compared with conventional alloys (group B), these alloys tend to show expectedly lower ductility. This is the perhaps the most important issue to be resolved for these alloys to be used widely.

Recently, casting of these alloys into near-net shape has been experimented for automotive engine components with promising results showing reasonable castability and acceptable low porosity levels [13, 21].

Group E alloys have been developed and designed on the basis of empirical findings that adding small amounts of refractory elements (W in particular) considerably alleviate the abnormal oxidation behavior of Group D alloys [15]. The oxidation resistance of these new alloys is remarkably improved such that it approaches that of Ni-base superalloy 718 at 870°C in air for a long period of time [18]. This is discussed further in the next section. These alloys may show improved creep resistance over that of Group D alloys, due to additional solid solution hardening expected by the additions of refractory elements. Preliminary tests show, however, that the resistance (in both primary and minimum rate creep regimes) appears to be only comparable at best to that of Group C (K5BCS) alloys treated under appropriate aging conditions. This is an indication that precipitation hardening by carbides (and silicides) enhances the high temperature creep resistance more effectively and further than solid solution hardening does, and this rather complex phenomenon is detailed in the next section. Though not quantified, prevailing evidence indicates the added alloying elements in these alloys have resulted in the reduction of ductility and possibly damage tolerance. Because of this concern, alpha extrusion began to be tried to generate fine-grained in-situ lamellar structures, likely texture, that could improve the concerned properties as discussed later. Tremendous, thoughtful work has yet to be made in understanding and optimizing the alloy compositions and the processing routes before attempting any scaling-up effort.

Evaluation of Critical Attributes

Compositional Evolution

From the discussions made in the previous section, it is clear that multi-component engineering gamma alloys have progressively narrowed the range of their major alloying compositions effectively down to Ti-Al-Nb ternary systems, with minor additions of B, C, Si and selected refractory elements. A portion of the Ti-Al-Nb phase diagram was estimated for 1000°C as a best effort, as schematically shown in figure 2. The pathway for the compositional evolution from group A to E alloys is shown using the base compositions X and Y defined as

X: Ti-48Al-2Nb for group A alloys

Y: Ti-45Al-(5-10)Nb for group D and E alloys

It is also clear from Figure 2 that the Al content must have been reduced in order to maintain reasonable amounts of alpha-2. However, we do not know yet the precise phase boundaries in the system as a function of temperature especially low temperatures (below 1100°C). Because of this missing information, optimizing alloy compositions and understanding the strengthening mechanism have not been completed, as described later for the latter.

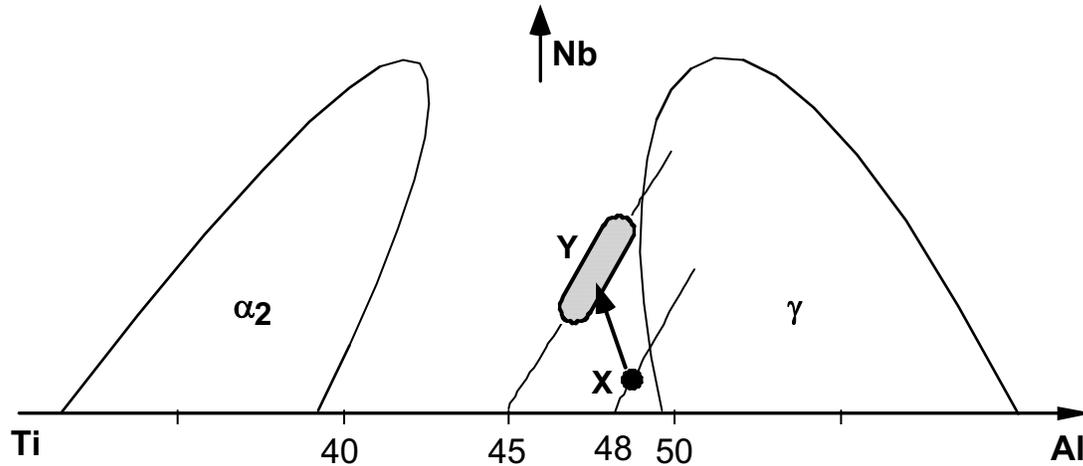


Figure 2: A Ti-Al-Nb isothermal phase field section (schematic) at about 1000°C shows the pathway for the compositional evolution of gamma alloys using the base compositions, X (for group A alloys) to Y (for D and E alloys).

Strengthening and Ductility

As discussed, high-Nb containing alloys (group D and E) have substantially increased strength levels over the conventional alloys, and this was explained by the solid solution hardening of the gamma phase by proposing some of Nb atoms taking the Al sites instead of Ti sites [22]. This interpretation has been recently disputed on the basis of a small size difference (~0.2%) between Ti and Nb atoms, and instead structural refinement such as abundant twinning activities was proposed to be responsible for the strengthening [13].

The explanations, however, neglect fundamental aspects that happen as Nb content increases from 2 to higher or X to Y in Figure 2. The changes will take place not only in phase field but also phase volume, as follows:

- 1) Increase of Nb, 2 to 5-10%, shift both phase boundaries ($\gamma/\gamma+\alpha_2$ and α_2/γ), altering the compositions as well as elemental partitioning in the constituent phases. The beta phase is not considered, since its small volume fractions should not affect the basic strengthening behavior.
- 2) Decreases in Al content (48 to 45%), combined with the phase boundary shifts, alter the relative volume fractions of constituent phases (γ and α_2). A small increase in the a_2/r volume ratio will increase the strength level significantly.

The shift of the $\gamma/\gamma+\alpha_2$ boundary is known toward the Ti side or away from the Al side (Figure 2), reducing the Al content in both γ and α_2 phases (see Figure 2). As these constituent phases become leaner with Al, Ti atoms will move in the Al sites, since Nb atoms are known to partition Ti sites. The consequential solid solution hardening of the gamma phase will take place due to Ti atoms (expelled by Nb atoms), and perhaps some of Nb atoms, substituting Al sites, instead of Nb atoms occupying the Ti sites. We do not know yet how the small atomic size differences between these incoming and Al atoms play to strengthen the phase. An assumption may be made that the size misfit for Nb in Al sites in the gamma phase would be substantial [Zhang] to strengthen the phase. The α_2/γ boundary appears to remain relatively unchanged (Figure 2). In this case, the incoming Nb atoms will simply move in the Ti sites resulting in little change in strength. This has yet to be verified by determining the phase boundaries and/or measuring the strength levels of both Ti-Al and Ti-Al-Nb materials having the boundary compositions. The role of the α_2/γ volume fraction can be significant depending on the degree of changes. Again, no quantitative information exists on the phase boundaries of the Ti-Al-Nb system around the compositions of interest at temperatures below 1000°C. Precise measurements of the phase boundaries and determination of the volume fractions must be made if this issue is to be resolved and clarified.

It is also possible that all reported group D alloy materials might have been in relatively non-equilibrium conditions since any stabilization (equilibration) treatments have been explicitly reported in the preparation of the investigated materials. Compositional equilibration will take time especially for high-Nb containing alloys because Nb additions are expected to slow down the diffusion process in the system. [15].

Due to their general tendency of showing low ductility, it has been suggested that fine-grained microstructures (duplex or near gamma) be favored for group D alloy materials [13]. On the other hand, recent experiments indicate that the ductility can be improved by introducing appropriate microstructures, such as nearly lamellar (NL) and in-situ lamellar microstructures [15]. When a NL microstructure was introduced in alloy 95G (Ti-45Al-10Nb) forging, the tensile ductility was measured to be over 1.0% upon an aging treatment [15]. Better ductility was achieved in some of group E alloys upon alpha extrusion as described in a later section. It may be thus more appropriate or meaningful to say ‘processing-specific ductility’ rather than just ‘ductility’ for group D and ED alloys.

Oxidation Resistance

For the last ten years, we have investigated to rank the oxidation resistance of various gamma alloys, both designed [15] and existing, under isothermal as well as cyclic exposure conditions at 760°C, 870°C and 925°C [18]. Some of the cyclic exposure results from representative alloys of each group are plotted in Figures 3. As shown, the oxidation resistance during the 1000h-500 cycles between RT and 760°C increases with Nb content from no-Nb (Ti-46Al) to 8Nb (alloy 98G or Ti-45Al-8Nb-0.2B-0.15C). Though not shown here, however, alloy 95G (Ti-45Al-10Nb) showed faster cyclic oxidation rates than 98G, in spite of its greater Nb content. It is then tempted to draw a conclusion that the beneficial effect of Nb additions on the cyclic oxidation resistance at 760°C appears to end below 10at%. Additional increases were achieved for group E alloys such as alloy 98D [Ti-45Al-6Nb-x(Cr, W, Mo, Hf)-y(B, C)]. The effects of increasing amounts of Nb (approximately to ~7at%) and small amounts of refractory elements on the oxidation resistance are not understood quantitatively, requiring basic studies for their mechanisms and kinetics. It is noted that Nb may reduce the oxygen penetration process, favoring the formation of Al₂O₃ [Taniguchi]; however, the basic mechanism for this process is not clearly explained either.

The responses to the cyclic exposures between 870°C and RT are shown in Figure 3b. It is particularly noted that alloy 98G gains weight rather rapidly initially up to about 400h exposure or 200 cycles, compared with other alloys, until the weight increases become parabolic. 95G also behaved similarly. Although increases in CTE differences between the Nb-containing oxides and the substrate may explain this, systematic studies are required to determine/quantify the precise mechanisms and diffusion-reaction kinetics. This rapid oxidation process, however, was eased significantly by small additions of refractory elements, as manifested by 98D. These empirical findings led us to the development of group E alloys, as described earlier.

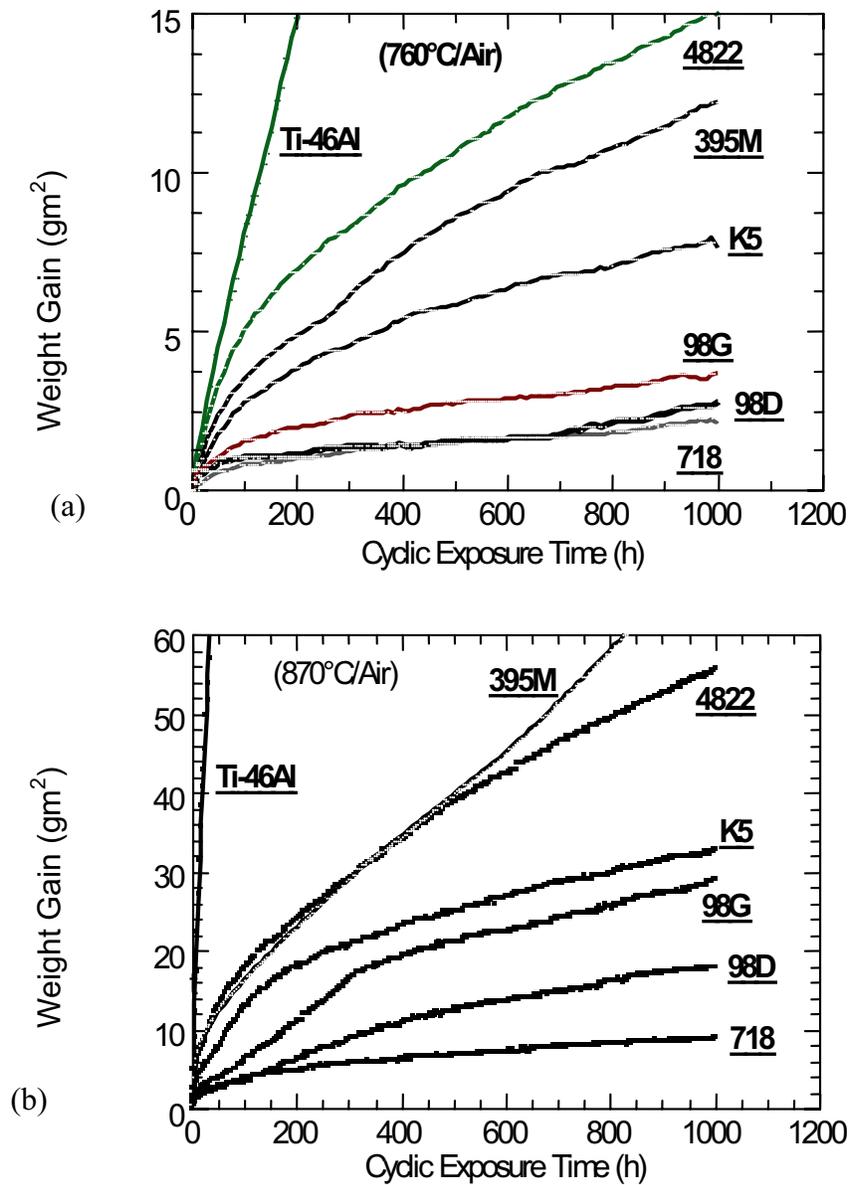


Figure 3: Cyclic oxidation behavior of various gamma alloys, as measured by the total weight changes, following cyclic exposures between 760°C and RT (a) and between 870°C and RT (b). 98G is a class D alloy and 98D is a class E alloy. At 760°C, the oxidation resistance increase rather monotonically with increasing Nb content (from 2Nb for 4822 to 8at%Nb for 98D), although further increases are achieved with 6at%Nb + x(W,Hf,Mo) for 98D. At 870°C, 98G shows initially high oxidation rates which are substantially suppressed in 98D.

Creep Resistance

It had been known for a long time that small amounts of carbon additions produce coherent carbides (Ti_3AlC) in the γ phase thereby increasing the creep resistance of the alloy system. These coherent particles are precipitated out of the supersaturated γ phase upon aging at relatively low temperatures ($<800^\circ C$) and effectively retard the creep deformation process. Apparently, however, they become ineffective, under dynamic loading conditions, at temperatures higher than $750^\circ C$ due to the tendency of coarsening and apparent weakening of coherency. Recognizing this deficiency of the coherent carbides at elevated temperatures, efforts in utilizing incoherent carbide particles were made in wrought K5 alloys, $Ti-46.2Al-2Cr-3Nb-0.2W-(0, 0.2)B-(0-0.2)C-(0, 0.2)S$ [9]. In the investigations, the wrought-processed materials were annealed at temperatures in the α field and cooled controlled rates to produce fully lamellar microstructures, followed by appropriate aging treatments at temperatures between 800 and $1000^\circ C$. Figure 4 shows creep test results for alloys K5 ($Ti-46Al-2Cr-3Nb-0.2W$) and K5SC ($K5-0.2Si-0.15C$) in fully-lamellar (FL) microstructure forms having coarse ($\sim 400\mu m$) grain sizes. Both materials were aged at $900^\circ C$ for 0h and 50h and crept at $760^\circ C$ under step-increment stress conditions, and their minimum creep rates are plotted as a function of applied stress. As shown, the creep resistance is reduced in K5-FL (no C+Si) upon aging, whereas it is considerably enhanced in K5SC-FL (with Si+C). Apparently, the aging increases lamellar spacing to reduce strength of the lamellar materials in K5-FL/50. This loss (high temperature strength) is much smaller in K5SC/50 than the precipitation hardening caused by the formation of carbides and silicides. (It is important to note that the aging may reduce the RT tensile strength even in K5SC-FL.) Upon aging, the $\alpha-2$ plates in the FL grains undergo an equilibration process in composition and shape. As shown in Figure 5, the $\alpha-2$ plates thin down non-uniformly through a diffusion process involving saturation and segregation of carbon, silicon, and beta-phase forming elements, Figure 5a and c. Eventually, the $\alpha-2$ plate is replaced by an array of carbide (Ti_2AlC), silicide (Ti_5Si_3), B2 and residual $\alpha-2$ particles.

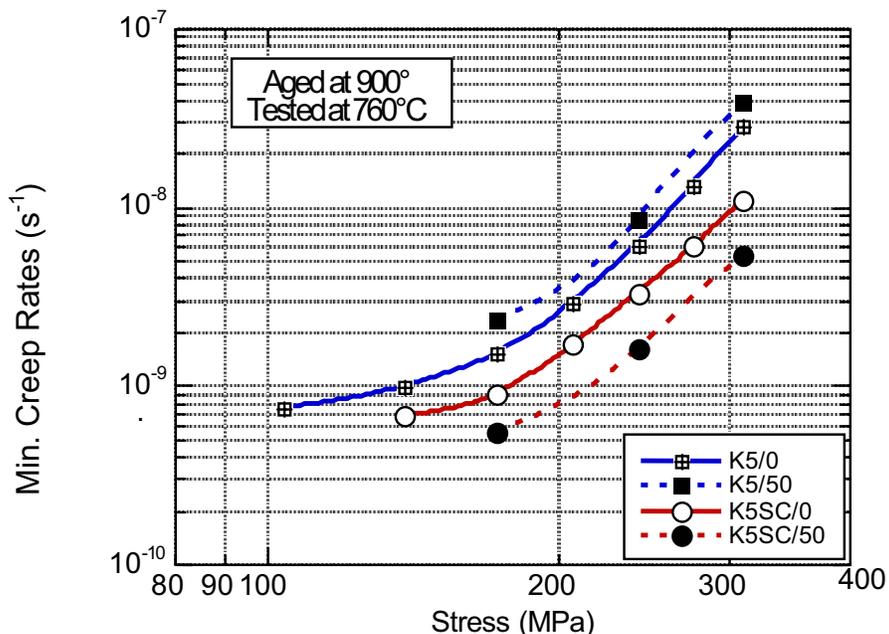


Figure 4: Effect of aging treatments on the creep resistance at $760^\circ C$, as measured by the minimum creep rate against applied stress, in K5 fully lamellar alloys (K5 and K5SC) with/without (C+Si) additions. Isothermal aging was conducted at $900^\circ C$ for 0 (K5/0 and K5SC/0) and 50h (K5/50 and K5SC/50).

Figure 5b, where the process is almost complete. It is noted that silicide particles alone are relatively ineffective in retarding creep deformation; however, their combined effect appears to be more effective than carbides alone. This apparent synergistic behavior needs to be understood for further compositional optimization. In addition, we do not know yet the optimum distributions of the strengthening particles and also the upper limits of their effectiveness. It is thus critically important to develop the processes and post-processing aging treatments that will optimize the size and spatial distribution of carbides and silicides. Critical for this optimization are the quantitative information on the solubility limits of carbon and silicon as a function of specific alloy composition.

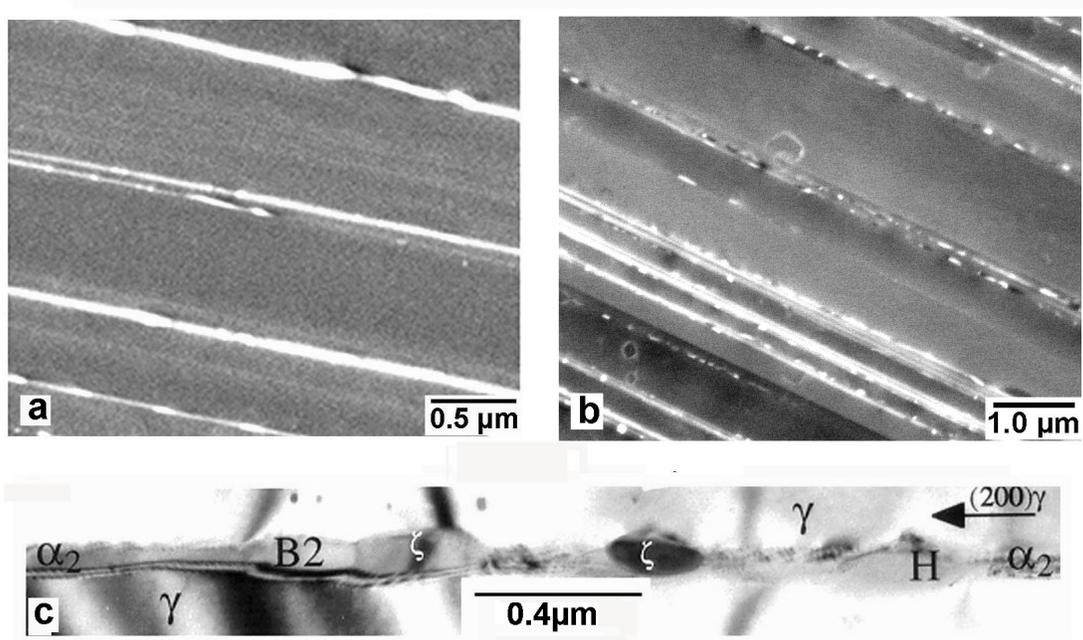


Figure 5: Images taken by the BEI technique showing the decomposition of alpha-2 plates (imaged bright) in an alloy K5SC FL grain that takes place upon aging at 900°C for 6h (a) and 50h (b). The equilibration process leads to the transformation of the originally perfect alpha-2 plate into an array of particles. The EDP analysis (c) identified the decomposed particles are mixture of H carbide (Al_2AlC), ξ silicide (Ti_5Si_3), beta phase, and alpha-2 phase (c).

Process Development and Mechanical Properties

As discussed earlier, increases in Nb content and decreases in Al content tend to decrease ductility and increase strength and tend to be more notch-sensitive. Consequently group D and particularly group E alloys are prone to develop cracking during hot working, and are particularly sensitive to defects and compositional/microstructural inhomogeneity. Processing of these alloys requires the elimination of those sources. Once a sound, defect-free and relatively uniform (compositionally and microstructurally) raw material is selected or obtained, the next step is to refine grains, hopefully, in a narrow size distribution. A boron addition is a good way of achieving this, and again small amounts of boron are commonly used for alloying in order to produce refined lamellar grain structures. Even if the average lamellar grain size (GS) is reduced by the borides, however, the GS's can vary widely depending on the size and spatial distribution of borides, which is directly dependent on the cast structure because borides

are formed in the last liquid during solidification. Not every microstructure is refined with the presence of borides. For example, fine grains such as duplex and nearly lamellar microstructures do not need borides for refinement. In fact, the grain refinement by borides is minimal for duplex microstructure material; however, there is qualitative evidence that borides help duplex microstructures become uniform in grain size.

One of the most potent ways to refine lamellar structures even without the presence of borides is alpha processing [15]. This process involves canning of a gamma alloy billet (either cast or mill product), soaking it at temperature in the alpha-phase field, transfer of the hot canned billet in RT (or warm) press dies with an appropriate dwell time, hot-working of the canned billet (while the billet is in the alpha status) at a controlled rate to an appropriate reduction (thickness or cross section), followed by controlled cooling through the dynamic lamellar-formation temperature. Since the billet material (in the alpha status) may have to be in a relaxed condition prior to the formation of stable lamellae, hot working must end reasonably well above the lamellar formation. On the other hand, achieving fine alpha grains require a shortest possible resident time in the single alpha-phase. All these complex competing conditions/parameters are alloy-specific as well as route-specific, and must be synergistically coordinated to obtain desired microstructures through modeling and validation experiments. There are three routes for this processing, that is, alpha extrusion, alpha forging and alpha rolling. The initial alpha grain size/structure during hot working is expected to be controlled by dynamic recrystallization. When properly executed, the resulting microstructure is typically called 'alpha processed' or 'situ' lamellar structure that consists of fine textured lamellar grains. These lamellar structures should be controlled by strain rate, total reduction, unloading temperature (with respect to the lamellar formation temperature) and cooling rate, and the combined effects must be understood to quantify the processes and optimize the microstructures.

Some experiments conducted recently show that desirable microstructures and improved tensile properties can be achieved through any of the alpha processing routes that employed fundamentally sound but empirical conditions and parameters [15]. Figure 6 compares an alpha extruded lamellar (AEL) microstructure in alloy K5 and a TMTL produced in alloy K5 by the conventional TMT method (hot-working in the $\alpha+\gamma$ field + alpha annealing). Limited number of tensile tests show alpha processed material exhibit both tensile strength and ductility improved

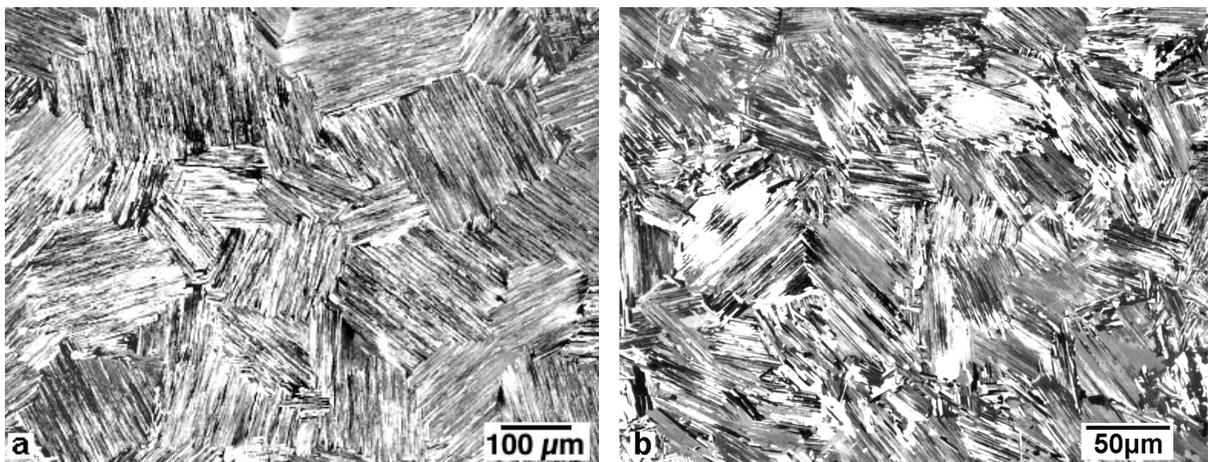


Figure 6. Two typical lamellar microstructures in a K5 alloy (Ti-46Al-2Cr-3Nb-0.2W-0.2B-0.2C), showing: (a) a random TMTL microstructure produced by the usual hot working + alpha anneal and (b) an alpha-extruded lamellar (AEL) structure which is finer and appears somewhat directional.

significantly over those of TMTL material, as shown in Figure 7a. The improved strength levels are retained over the entire test temperature range to 1000°C, Figure 7b. Experimental evidence indicates that post-processing aging treatments affect the tensile properties significantly as well as, as expected, creep properties when the alloy contains carbon.

Technically, alpha extrusion can be best employed to produce blades made of HT gamma alloys, and alpha forging can be ideal for producing HT gamma disks requiring high strength as well as improved low-cycle fatigue resistance. The properties in Figure 7 were obtained from the alpha-processed of material small alloy scale ingots or billets. It would be important to take steps toward scaling up the process to see if the beneficial microstructure/property combinations can be demonstrated in application-specific size materials.

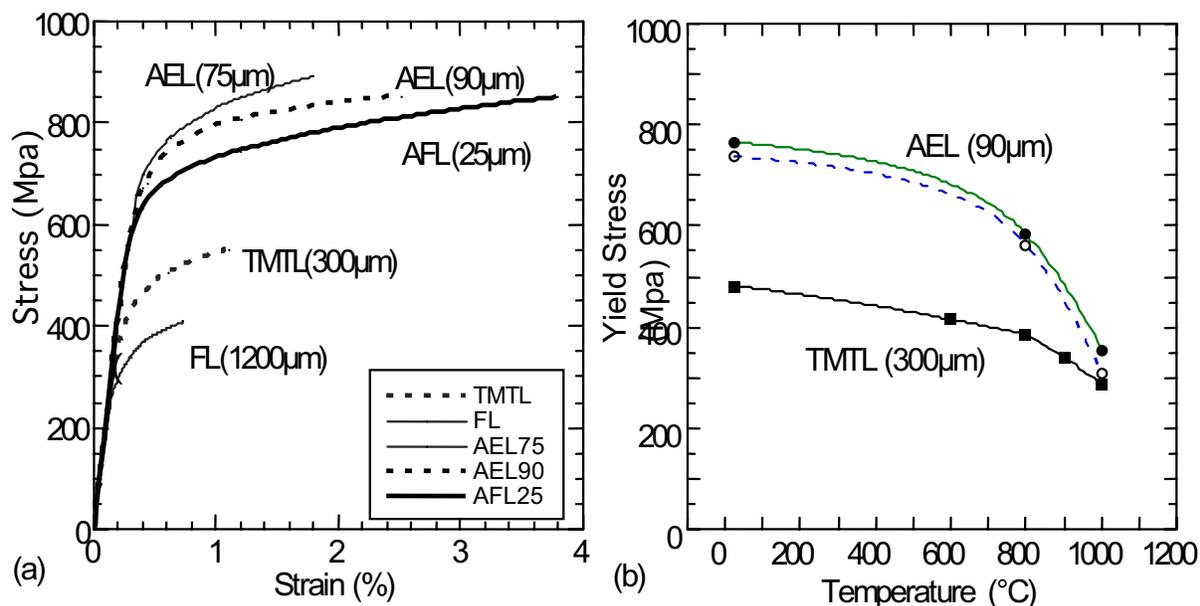


Figure 7. Tensile properties of alloy K5 lamellar materials widely vary depending on processing. (a) Alpha extruded lamellar (AEL) and alpha forged lamellar (AFL) materials show both RT tensile strength and ductility substantially increased over those for conventionally processed TMTL and FL materials. In general, the RT tensile properties are improved with decreasing grain size (shown in the parentheses); however, this is not the case within the alpha processed lamellar materials group. (b) The superior tensile strength levels in alpha processed materials are retained at least up to 1000°C under a strain rate of $2 \times 10^{-4} \text{ s}^{-1}$, and this trend is exemplified with yield strength for the AEL (90µm) against TMTL (300µm). It appears that they converge at a higher temperature around 1100°C.

Summary

Technological challenges ahead are enormous in making the oxidation- and creep-resistant high-Nb containing alloys technologically viable. The first step is to understand the fundamentals of oxidation and creep resistance mechanisms and to identify both individual and synergistic effects of alloying elements. The second is to optimize the compositions of high Nb-containing alloys, and this requires further improved knowledge and information of phase

relations, oxidation resistance, creep resistance and processibility. When conventionally processed in the two-phase ($\alpha + \gamma$) field, both forgings and extrusions of these alloys show lower ductility than those of current gamma alloys. This has been a concern, and fine-grained duplex microstructures are generally favored over fully-lamellar microstructures for these alloys. Recently, however, alpha processing has shown that in-situ lamellar microstructure can be produced in these alloys to enhance both ductility and strength as well. Further work on the development and optimization of alpha processes is warranted for these promising but brittle alloys.

References

1. Y-W. Kim, "Ordered Intermetallic Alloys, Part III: Gamma Titanium Aluminides," *JOM*, 46 (1994), 30-39 and 7.
D.M. Dimiduk, P. L. Martin, and Y-W. Kim, *Mater. Sci. & Eng. A* 243 (1998), 66-76.
3. C.M. Austin, et al., *Gamma Titanium Aluminides*, eds., Y-W. Kim, R. Wagner and M. Yamaguchi, (TMS, Warrendale, PA, USA, 1995), 21-32.
4. M. Blum, et al., *Gamma Titanium Aluminides*, eds., Y-W. Kim, D. Dimiduk and M. Loretto (TMS, Warrendale, PA, USA, 1999), 35-39.
5. P.A. McQuay and D. Larsen: "Manufacturing-Performance-Cost Relationships in Gamma TiAl Castings," *Structural Intermetallics 1997*, eds. M. V. Nathal, R. Darolia, et al., (TMS, Warrendale, PA, USA, 1997), 523-529.
6. R. Wagner, F. Appel, B. Dogan et al., *Gamma Titanium Aluminides*, eds., Y-W. Kim, R. Wagner and M. Yamaguchi, (TMS, Warrendale, PA, USA, 1995), 387-404.
7. V. Lupinc, M. Marchionni, M. Nazmy, et al., *Structural Intermetallics 1997*, eds. M. V. Nathal, R. Darolia, et al., (TMS, Warrendale, PA, USA, 1997), 515-522.
8. H. Clemens, et al., *Gamma Titanium Aluminides*, eds., Y-W. Kim, D. Dimiduk and M. Loretto (TMS, Warrendale, PA, USA, 1999), 209-223.
9. Y-W. Kim and D.M. Dimiduk, US Patent 5,226,985, (13 July 1993).
10. Y-W. Kim, *Gamma Titanium Aluminides*, eds., Y-W. Kim, R. Wagner and M. Yamaguchi, (TMS, Warrendale, PA, USA, 1995), 509-524.
11. T. Tetsui and S. Ono, *Intermetallics*, 7 (1999), 689-697.
12. H. Kestler, N. Eberhardt and S. Knippscheer, *Nb for High Temperature Applications*, eds. Y-W. Kim and T. Carneiro, (TMS, Warrendale, PA, 2004), In Press.
13. F. Appel, et al., *Nb for High Temperature Applications*, eds. Y-W. Kim and T. Carneiro, (TMS, Warrendale, PA, 2004), In Press.
14. D.M. Dimiduk, P. Martin, and R. Dutton, *Gamma Titanium Aluminides* eds. Y-W. Kim, H. Clemens and A. Rosenberger (TMS, Warrendale, PA, 2004), In press.
15. Y-W. Kim, Unpublished results.
16. J.D.H. Paul, F. Appel and R. Wagner, *Acta Mater.* 46, 1075 (1998), p.1075.
17. F. Appel, M. Oehring, J.D.H. Paul, and U. Lorenz, in: *Structural Intermetallics 2001*, eds. K.J. Hemker, D.M. Dimiduk, H. Clemens, R. Darolia, H. Inui, J.M. Larsen, V.K. Sikka, M. Thomas, J.D. Whittenberger (TMS, Warrendale, PA, 2001), 63-72.
18. M. Yoshihara and Y-W. Kim, *Gamma Titanium Aluminides*, eds., Y-W. Kim, D. Dimiduk and M. Loretto (TMS, Warrendale, PA, USA, 1999), 753-760.
19. Y-W. Kim and D.M. Dimiduk, *Structural Intermetallics*, eds. M.V. Nathal, R. Darolia, et al. (TMS, Warrendale, PA, 1997), 531-543.
20. G.L. Chen, W.J. Zhang, Z.C. Liu, S.J. Li, and Y-W. Kim, *Gamma Titanium Aluminides* eds. Y-W. Kim, D.M. Dimiduk and M.H. Loretto (TMS, Warrendale, PA, 1999), 371-380.
21. X. Wu and M. Loretto, *Nb for High Temperature Applications*, eds. Y-W. Kim and T. Carneiro, (TMS, Warrendale, PA, 2004), In Press.
22. W.J. Zhang, S.C. Deevi, and G.L. Chen, *Intermetallics*, 10 (2002), 403-406.