THE THERMAL STABILITY OF THE MICROSTRUCTURE OF 
\(\gamma\)-BASED TITANIUM ALUMINIDES

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Abstract—The thermal stability of the microstructure of \(\gamma\)-based titanium aluminides after aging at elevated temperatures was studied by transmission and analytical electron microscopy (TEM and AEM, respectively) as well as by optical microscopy. Predominantly lamellar microstructures of Ti-47Al (at.\%), Ti-47Al + B (140ppm) and Ti-47Al + B + W (0.5at.%.) alloys were produced by heat-treatment at 1400°C for 1h followed by furnace cooling. The average lamellar spacing of the as-heat treated was measurably finer in both alloys with solute additions (400nm in Ti-47Al, 170–190nm in alloys modified with B or W + B). The \(\alpha_2\) lamellae were especially finely distributed and continuous in the Ti-47Al + W + B alloy. These alloys were then aged at 800, 1000 or 1200°C for 168h. The alloys aged at 800°C did not show much change, whereas aging at 1000°C produced significant lamellar break-up and coarsening in the Ti-47Al and B-modified alloys, but not much change in the alloy modified with W + B. In addition, coarsening by grain boundary migration led to the formation of new coarser, irregular and branched lamellae. Aging for 168h at 1200°C significantly degraded the originally fine lamellar structure. In addition to the lamellar degradation processes observed at 800–1000°C, spheroidized precipitates of \(\alpha_2\) and \(\gamma\) were also observed at 1200°C. At all three aging temperatures the cut surfaces of the samples showed additional effects of aging, not observed in the bulk. These included the loss of Al from the sample during vacuum annealing, and the recrystallization and growth of new grains into the lamellar structure. After aging at 800 and 1000°C, these sub-surface microstructural changes were found to be greatest for the case of the binary alloy and least for the Ti-47Al + W + B alloy. The observed changes of the lamellar structure during aging were found by TEM to be related to the following defects present in the lamellar structure prior to aging: low angle grain boundaries, misoriented lamellae, lamellar interfaces and lamellar colony boundaries. Some of the observed modes of microstructural instability were fault and edge migration, the dissolution of lamellae at lamellar interfaces, and the formation of new \(\gamma\) grains within the lamellar structure. At 1200°C, even the coarsened lamellar structure ultimately spheroidizes, particularly in the binary alloy. Other kinds of microstructural instability observed included the migration of colony boundaries, the penetration of one colony into an adjacent colony, and the growth of \(\gamma\) grains found at colony boundaries into the lamellar structure of that same colony. The effect of alloying additions of B and W to the binary alloy is to refine the initial lamellar structure produced by heat-treatment at 1400°C, and then to stabilize that original lamellar structure (particularly B + W) during subsequent aging at 800–1000°C.

INTRODUCTION

The increasing need for high temperature materials for structural applications has generated considerable interest in ordered intermetallics, and particularly in \(\gamma\)-based titanium aluminides [1–3]. \(\gamma\)-based TiAl alloys offer an attractive combination of low density, good creep resistance, high-temperature strength and oxidation resistance. These alloys are generally two phase structures, consisting of the \(\alpha_2\) (D0₅, structure) and \(\gamma\) (L1₂ structure) phases at room temperature (Fig. 1). At temperatures above the eutectoid temperature (approximately 1120°C), the \(\alpha_2\) phase disorders to the \(\alpha\) (h.c.p.) phase. It has been established recently [1] that the importance of these alloys in engineering applications arises mainly from the attractive mechanical properties of the lamellar colonies which consist of plates of the \(\alpha_2\) and \(\gamma\) phases. When these alloys are furnace cooled, the lamellar structure forms as a result of precipitation of plates of \(\gamma\) from the \(\alpha\) phase field, and the remaining \(\alpha\) phase then orders to the \(\alpha_2\). In addition, equiaxed \(\gamma\) grains are also often formed at lamellar colony boundaries. Considerable effort has been devoted to optimizing the mechanical properties of these alloys through alloying additions and by thermomechanical processing. These efforts have been directed towards control of the size, shape and relative volume fractions of the lamellar colonies and \(\gamma\) grains. Within a lamellar colony, the interlamellar spacing and the relative volume fraction and distribution of the \(\alpha_2\) and \(\gamma\) phases can also be altered, and attempts have also been made to alter the structure of the boundaries between lamellar colonies [1].
It has been observed that the mechanical properties of these alloys deteriorate when the original lamellar structure and the \( \gamma \) grains become unstable, as a result of additional thermal and thermomechanical exposure. One such example is the increase during creep testing of the minimum creep rate caused by the spheroidization of the lamellar structure [4]. Another example of the importance of the stability of the microstructure in these alloys is the observation that superplastic forming is adversely affected by the formation of a surface layer which forms as a result of Al loss and other microstructural changes at the surface [5]. Theoretical and experimental studies have been performed to study the mechanisms of instability in other predominantly lamellar structures, such as pearlite in steel or several aligned eutectic or eutectoid in-situ composites prepared by directional solidification techniques [6-23]. Although the various microstructural changes occurring on aging at elevated temperatures occur concurrently in order to lower the total free energy of the system, it is convenient to conceptually divide these changes into the following three categories: (1) the change of the volume fraction and composition of the phases towards the equilibrium volume fraction and phase boundary compositions by dissolution/precipitation of the relevant phases (called phase boundary driven changes); (2) the reduction of interfacial energy by a reduction in interfacial area through coarsening either by lattice diffusion or pipe diffusion (called continuous coarsening); or (3) coarsening to reduce the interfacial area by grain boundary migration (discontinuous coarsening). In experimental systems, combinations of all three of these effects are often observed. For example, grain boundary migration to reduce the interfacial area is often accompanied by changes in volume fraction and chemistry of the phases dictated by phase diagram (volume free energy) considerations.

The general features of the instability of the microstructure on aging in \( \gamma \)-based TiAl alloys has also been reported recently [24-27]; however, no systematic work on the effect of alloying additions and temperature has been reported in the past. This paper reports the results of an investigation, using optical microscopy, transmission electron microscopy (TEM) and analytical electron microscopy (AEM), to study the mechanisms of microstructural changes on aging at temperatures of 800, 1000 and 1200°C in Ti-47Al (in at.\%), Ti-47Al-140ppm B and Ti-47Al-140ppm B-0.5W alloys. The aim of this investigation is to study the effects of aging temperature and of alloying element additions on the thermal stability of the lamellar microstructure in these \( \gamma \)-TiAl alloys. These alloying additions of B and W were chosen because the addition of boron has been reported to stabilize the lamellar structure [20-30] and to improve the mechanical properties of these alloys [30]. The addition of tungsten has also been reported to improve the creep resistance [3] and to stabilize the lamellar structure [1].

**EXPERIMENTAL PROCEDURE**

Alloy ingots (25mm in diameter and 50mm long, weighing 175g) of nominal composition Ti-47Al, Ti-47Al-40ppm B and Ti-47Al-40ppm B-0.5at.%W alloys were prepared by arc melting and drop casting, and the difference in weight after melting was less than 0.5%. A small fraction of the tungsten was not completely melted in the quaternary alloy, and settled to the bottom of that ingot. Sectioning of these ingots was carried out by a slicer or by a low speed saw. Samples from these sections were heat treated in a vacuum furnace at 1400°C for 1h, followed by furnace cooling. These samples were then further aged at either 800, 1000 or 1200°C for 1 week followed by furnace cooling to room temperature. Specimens from these samples, corresponding to longitudinal sections of the rod, were prepared for optical microscopy by conventional grinding and polishing using 0.5\( \mu \)m diamond paste followed by etching with a solution of 1% HF and 10% HNO\(_3\) in distilled water. TEM foils were prepared in a twin jet electropolishing unit from 200\( \mu \)m thick, 3mm diameter disks using a solution of 600ml methanol, 340ml 2-butoxy-ethanol (butyl cellulose), 60ml perchloric acid and 5ml glycerine and operating at −20°C and 32V. The foils were immediately rinsed, in baths of methanol and ethanol after thinning. TEM and AEM were carried out using a Philips EM400T AEM equipped with a field emission gun (100kV, probe size <10nm), a CM30 (300kV) and a CM12 (120kV) AEM. The CM12 and EM400T microscopes were interfaced to EDAX 9900 and EDAX 9100 computer analyzers, respectively. Quantitative measurements of the average interlamel-
lar spacing were carried out by tilting the specimens so that all lamellar ($\gamma$ and $\alpha_2$ phase) boundaries were parallel to the electron beam, and visible, and boundary thicknesses were minimum. Quantitative analysis of the X-ray energy dispersive spectroscopy (XEDS) data was carried out on some of these specimens, and another binary $\gamma$-Ti–48Al sample of known chemical composition was used to check $k$ factors for Al concentration. Quantification of the concentration of W in the $\alpha_2$ and $\gamma$ phases using a standardless method with calculated $k$ factors for W concentration, indicated an overestimate of up to 0.4at.% in the analysis of tungsten levels. The use of experimental standards is important in the case of elements such as Al, where there is known to be a significant difference between the proportionality constant relating weight percent to the observed intensity ($k$ factor) calculated from first principles, and those $k$ factors which have been experimentally measured in particular microscopes.

Fig. 2. (a and b).
RESULTS

Optical microscopy observations

The as-cast Ti–47Al alloy (prepared by drop-casting) consisted of lamellar colonies (with average size 100 \( \mu \text{m} \) or less) and a small volume fraction of equiaxed \( \gamma \) grains mostly formed at the colony boundaries [Fig. 2(a)]. The size of the lamellar colonies in the as-cast structure is small because these small drop-castings have a higher cooling rate than larger ingots. Solutionizing of the as-cast structure in the \( \alpha \) phase field by heat-treatment at 1400°C for 1h (followed by furnace cooling) produced larger lamellar colonies (average size 700 \( \mu \text{m} \)), which had formed from the prior \( \alpha \) grains, as well as fine equiaxed \( \gamma \) grains along the colony boundaries [Fig. 2(b)] (shown later in the TEM results). A significant fraction of the as-heat-treated colonies have serrated colony boundaries, with lamellae from one colony penetrating directly into the adjacent colony [Fig. 2(b)]. Further aging at 800°C for 168h showed dissolution of some lamellae within the
lamellar colonies, and growth of the $\gamma$ grains found along or near colony boundaries to a size of about 30$\mu$m [Fig. 2(o)]. Beneath the cut surfaces of the sample, a layer of recrystallized $\gamma$ grains was observed in many regions, ranging from 20 to 24$\mu$m deep below the surface [Fig. 2(d)]. Such a layer indicates that near-surface deformation was introduced during specimen cutting. The lamellar structure just below this sub-surface layer generally showed more change (coarser and more irregular lamellar structure) than other regions of the bulk specimen. Aging for 168h at 1000°C resulted in the growth of equiaxed and elongated $\gamma$-grains at colony boundaries, the breakup of some lamellae into segments, and the thickening of some lamellae at the expense of their neighboring lamellae [Fig. 2(e)]. At the edge of the sample, growth of $\alpha$ phase (hereafter called the $\alpha$ case) takes place due to the loss of aluminum from the surface during aging.
in vacuum. Deeper beneath that surface, a 60–100 μm thick layer of equiaxed γ grains was found, with a significant fraction of those γ grains containing new α\textsubscript{2} precipitates within the grains and at grain boundaries [Fig. 2(f)]. The overall sub-surface region comprised of the α case, the layer of recrystallized γ grains and the growth and penetration of some of these large γ grains into the lamellar structure directly
underneath, and is referred to as the "damaged layer". After aging at 1200°C for 168h, the bulk structure consisted of much coarser, irregular and branched segments of lamellae [Fig. 2(g)]; the interlamellar spacing after such aging was much coarser than at lower temperatures, ranging from 0.5 to 4μm, and varying considerably from one grain to another. At 1200°C, coarse spheroidized and plate-shaped α2 phase particles were also observed within the large non-lamellar γ grains [Fig. 2(h)].

The as-cast microstructure of the Ti-47Al + B (140ppm) alloy (not shown) was similar to that found in the binary alloy [Fig. 2(a)], but with more equiaxed γ between lamellar colonies. After the heat-treatment at 1400°C for 1h, the initial microstructure consisted of fully-lamellar colonies similar to those found in the binary alloy. A
sub-surface “damage layer” consisting mainly of equiaxed and elongated $\gamma$ grains was also observed at the edge of the specimen. Slight coarsening of the lamellar structure at the colony boundaries was accompanied by displacement of the boundary from one colony to the other [Fig. 3(a)]. After aging at 800°C, no appreciable change in the microstructure was observed optically [Fig. 3(b)]. After aging at 1000°C for 168h, the microstructure showed the growth of new $\gamma$ grains along the colony boundaries and dissolution or coarsening of some of the lamellae within the colonies [Fig. 3(c)]. After aging at 1000°C, the layer of recrystallized grains in the “damage layer” beneath the cut surface was 10–45μm thick, significantly less than found in the binary alloy [Fig. 2(f)]. The lamellar microstructure of the sample aged at 1200°C showed considerable change, with the resultant microstructure being similar to that found

Fig. 3. Optical microstructures of Ti-47Al-40wppm B alloy (a) heat treated at 1400°C for 1h followed by furnace cooling, note lamellar colony interpenetration producing secondary lamellae; (b) aged at 800°C for 1 week; (c) aged at 1000°C for 1 week; and (d) aged at 1200°C for 1 week. An example of discontinuously coarsened regions can be observed in the lower left of the figure.
in the binary alloy [Fig. 3(d)]. The coarse lamellar grains also contained many discontinuously coarsened regions [e.g. in the lower left of Fig. 3(d)].

The lamellar colony surface of the Ti–47Al–0.5W + B (TiAl + W + B) alloy heat-treated at 1400°C was also similar to that of the Ti–47Al and Ti–47Al + B alloys given the same heat-treatment [Fig. 4(a)]. Isolated lamellae [see, e.g. upper left of Fig. 4(a)] are coarser near the lamellar colony boundaries, but there is also significant interpenetration of lamellae across the boundary from one colony to another, and no apparent equiaxed γ grains. Aging at 800°C had little or no effect on the lamellar colonies and the colony boundaries with interpenetrating lamellae, at least optically [Fig. 4(b)]. Aging at 1000°C produced some small changes, including the formation of small equiaxed γ grains (average size of 20μm) [Fig. 4(c)]. In contrast to the other alloys, the W + B modified alloy aged at 1000°C had only an α-case in the "damaged layer" beneath the cut

Fig. 4. (a and b).
surface, and not a measureable layer of recrystallized γ grains. Aging at 1200 °C produced coarse, irregular lamellae throughout the microstructure, with interlamellar spacings of up to 6 μm [Fig. 4(d)]. However, relative to the other alloys, the Ti-47Al + W + B alloy had less discontinuous coarsening and spheroidized α₂.

Transmission and analytical electron microscopy observations

The lamellar microstructure of the Ti-47Al alloy heat-treated for 1 h at 1400 °C and furnace-cooled had an average interlamellar spacing of 400 nm [Fig. 5(a) and Table 1] and γ grains along colony boundaries
[Fig. 5(b)]. The lamellar structure also had a fairly regular pattern of alternating \( \gamma \) and \( \alpha_2 \) lamellae. While the general lamellar structure was fairly uniform, the following defects could be identified within the lamellar colonies: low angle boundaries between \( \gamma \) plates [Fig. 6(a)], twin formation at the edge of a plate leading to misoriented plates [Fig. 6(b)], and kinked lamellae [Fig. 6(c)]. Aging at 800°C produced some changes within the initial lamellar structure, including: (a) dissolution of the lamellae along their common interfaces as well as by fault migration within individual \( \gamma \) lamellae, which resulted in the dissolution of the receding phase [Figs 6(d) and (e)]; (b) continuous dissolution of lamellae initiated at the interphase interface followed by step migration [Fig. 6(f)]; (c) thermal grooving at low angle boundaries between plates and the remnants of dislocation networks which formed the interfaces of the initial lamellar structure prior to aging at 800°C [Fig. 6(g)].

Aging of the Ti–47Al alloy at 1000°C produced significant continuous coarsening, as shown in Fig. 7(a), increasing the average lamellar spacing to 800–1950nm (Table 1). Discontinuous coarsening was also observed, which was not found at 800°C, as shown in Figs 8(a) and (b). At higher magnification, the interface between finer discontinuously coarsened lamellar regions indicates that coarser lamellae are growing into the finer structure [Fig. 8(b)]. A comparison of several sets of selected area diffraction patterns (SADP) of the \( \gamma \) and \( \alpha_2 \) phases in the initial and final lamellar structures showed that no simple rational crystallographic orientation relationship existed between these phases across the transformation interface between the continuously and discontinuously coarsening lamellar regions [Figs 8(c) and (d)]. However, within each of these lamellar structures the Shoji–Nishiyama orientation relationship between \( \gamma \) and \( \alpha_2 \) phases was observed: \{111\} of \( \gamma \) // \{0001\} of \( \alpha_2 \), \{110\} of \( \gamma \) // \{1120\} of \( \alpha_2 \); [Figs 8(e) and (f)]. Aging the binary alloy at 1200°C, in addition to accelerating the mechanisms of microstructural instability observed at 800 and 1000°C, also produced an unusual feature consisting of parallel coarse and fine \( \gamma \) plates growing roughly perpendicular to and through finer lamellae structure regions [Fig. 8(g)]. Both the coarser and the finer new plates appear to shear and displace the initial lamellar structure during their growth process, as illustrated in Fig. 8(h).

In the Ti–47Al+B alloy, the initial lamellar structure produced by heat-treatment for 1h at 1400°C was finer than found in the binary alloy, with an average lamellar spacing of 170nm (Table 1). Both the \( \gamma \) and \( \alpha_2 \) lamellae were refined, but many of the \( \alpha_2 \) lamellae were discontinuous or fragmented between the more continuous and defect-free \( \gamma \) lamellae. There were also some isolated equiaxed \( \gamma \) grains along colony boundaries, with an average size of 4\( \mu \)m [Fig. 9(a)]. Aging for 168h at 800°C produced very little change in microstructure, but some lamellar dissolution was observed in a few areas [Fig. 9(b)] within each colony. Aging for 168h at 1000°C produced more coarsening, but the continuous coarsening in the Ti–47Al+B alloy was

![Fig. 5. (a).](attachment:image.png)
Fig. 5. TEM micrographs at lower magnification of the typical lamellar structure within colonies of (a) Ti-47Al alloy and (c) Ti-47Al + W + B alloy, both after heat-treatment for 1h at 1400°C. Part (b) shows elongated and equiaxed γ grains that formed at colony boundaries in the Ti-47Al alloy. Parts (a) and (c) are tilted to contrast conditions that make the α₂ phase lamellae strongly-diffracting (dark).
significantly less (2–5 times less, see Table 1) than found in the binary alloy. The initial lamellar structure could be seen to be coarsening via dissolution of the lamellae by faults migrating along the interfaces and leaving dislocation debris (networks) within the new, coarser lamellae [Figs 9(c) and (d)]. Both $\gamma/\gamma_{\text{ mix}}$ and $\gamma/L$ interfaces were found to be unstable, although a study of the relative stability of such different interfaces was not done here. The Ti-47Al + B alloy also did not show the regions of discontinuous coarsening that were found in the binary alloy after aging at 1000°C. Aging at 1200°C did cause discontinuous coarsening in the Ti-47Al + B alloy, in addition to the modes of coarsening observed at lower temperatures [Fig. 9(c)].

Most of the interfaces between continuous and discontinuously coarsening lamellar regions showed only extinction fringes in dynamical contrast due to thickness, and did not show dislocations. However, occasionally, some such interfaces showed a more complex contrast structure [Fig. 9(f)].

The Ti-47Al + W + B alloys also had a relatively finer lamellar structure after heat-treatment for 1h at 1400°C that was similar to the Ti-47Al + B alloy, with an average lamellar spacing of 190nm [Fig. 5(c) and Table 1]. However, unlike the Ti-47Al + B alloy, the $\alpha_2$ lamellae in the Ti-47Al + W + B alloy were somewhat thicker and more uniformly distributed within the lamellar structure, and were as continuous as the $\gamma$ lamellae [Figs 5(c) and 10(a)]. Intercolony $\gamma$ grains were not observed. Aging at 800°C produced almost no change in the initial lamellar structure, although occasional dissolution of isolated individual lamellae could be found [Fig. 10(b)]. The Ti-47Al + W + B alloy was much more resistant to lamellar coarsening during aging for 168h at 1000°C compared to the other two alloys, with the lamellar spacing being almost the same as it was initially [Fig. 8(b) and Table 1]. It appears that there is some minor dissolution of some $\alpha_2$ lamellae and a few more twins in the $\gamma$ lamellae, but these are minor changes compared to the binary alloy during similar aging. Some new $\gamma$ grains also appear to have grown with the lamellar colonies during aging [Fig. 10(c)], and some of the larger $\gamma$ grains contained spheroidized $\alpha_2$ and $\gamma_{\text{ mix}}$ [Fig. 10(d)]. Aging at 1200°C, produced more of the larger $\gamma$ grains containing spheroidized $\alpha_2$ and $\gamma_{\text{ mix}}$ phases precipitates. Regions of very coarse lamellae produced by discontinuous coarsening were
also observed, but many more regions of finer lamellae that were coarsening continuously still persist in the Ti-47Al + W + B alloy [Figs 10(e) and (f)], while such regions were gone in the other two alloys aged at 1200°C.

Microcompositional analysis of individual γ and α₂ lamellae was also performed using AEM for the Ti-47Al + B and Ti-47Al + W + B alloys in the as-heat-treated and the heat-treated and aged (1000°C) conditions, and data are given in Table 2. The γ phase contains about 48at.%Al in both alloys, with little difference between aged and unaged material. The finer α₂ phase lamellae contain about 40% Al initially in both alloys, but then have slightly less Al (about 36%) after aging at 1000°C. The variation between individual measurements (±0.4%,

Fig. 6. (b and c).
Table 2) is less than the measured Al difference, and the same modest reduction in Al content is observed in two different aged alloys. Therefore, this measured change in \( \alpha_2 \) phase composition as a result of aging appears to be meaningful. Even the thinnest \( \alpha_2 \) phase lamellae are still much larger than the electron probe size in the FEG/AEM (\( >50 \text{nm} \) \( \alpha_2 \) width compared to \(<10 \text{nm} \) beam size) so that spatial resolution is not a problem, and any overlap with the \( \gamma \) matrix would raise rather than lower the Al. Therefore, there does appear to be a subtle effect of aging at 1000°C lowering the Al content of the \( \alpha_2 \) phase lamellae,
Fig. 6. TEM micrographs (bright field = BF, centered dark field = CDF) of Ti-47Al alloy heat treated at 1400°C for 1h followed by furnace cooling (a) low angle boundaries between plates (BF); (b) misoriented lamellae (BF); and (c) kinked lamellae with kinks marked with arrows (BF). Aged at 800°C for 1 week: (d) migration of plate edge (marked with an arrow) leading to plate dissolution (CDF); (e) dissolution along lamellar interfaces and fault migration (marked with an arrow); (f) dissolution at interphase interface accompanied by step migration; (g) thermal grooving at faulted region (marked with an arrow) and dislocation network debris are also observed.
Fig. 7. TEM micrographs at lower magnification showing typical lamellar structures within colonies of (a) Ti–47Al alloy and (b) Ti–47Al + W + B alloy, both after heat-treatment for 1h at 1400°C, followed by aging for 168h at 1000°C. Note the significantly coarser lamellae in the Ti–47Al alloy in (a).
which is also consistent with the observation that they are dissolving and/or coarsening at this temperature. In the Ti–47Al + W + B alloy, there also appears to be a slight difference in W content between the α and γ phases in the initial lamellar structure, with the initial fine α phase having slightly more W. Aging, however, at 1000°C eliminates that W partitioning difference between the two lamellar phases.

Additional AEM measurements of non-lamellar microstructural features in the Ti–47Al + W + B alloy aged at 1000°C showed equiaxed γ grains to have about the same composition found in finer γ lamellae. Analysis of the microstructural features
observed in the Ti-47Al + W + B alloy aged at 1200°C showed that the average composition of coarse 𝛼₂ lamellae was Ti-43.3Al-0.7W, which was measurably different than that of spheroidized 𝛼₂ precipitates (Ti-39.6Al-0.3W). The 𝛾 matrix surrounding such spheroidized 𝛼₂ phase however, still had about the same composition found in lamellae of that phase at 1000°C (Ti-50.6Al-0.6W).

**DISCUSSION**

The initial microstructures of the Ti-47Al, Ti-47Al + B and Ti-47Al + W + B alloys after heat treatment at 1400°C for 1 h are similar in the general sense that they consist of large fully-lamellar colonies with relatively fine lamellar structures. However, there are some subtle differences in the finer details of
those lamellar structures produced by the B and W alloying additions that are important. Both the Ti-47Al and Ti-47Al + B alloys contain a small volume fraction of equiaxed γ grains along colony boundaries, but the Ti-47Al + W + B alloy does not. The initial lamellar spacing of both the B and B + W modified alloys are about a factor of two finer than that of the Ti-47Al binary alloy. However, the α2 phase lamellae are different, with those lamellae being quite thin, and fragmented or discontinuous in the Ti-47Al + B alloy, and being slightly thicker, continuous and more defect-free in the Ti-47Al + W + B alloy.

The data on the effects of aging for 168h on these
Fig. 8. TEM microstructures of Ti-47Al alloy aged at 1000°C showing (a) BF of discontinuous coarsening; (b) BF of interface between final (coarse) and initial (fine) lamellae. The initial lamellar structure is on the lower left of the figure; (c) SADP of coarse γ lamellae, Zone Axis (Z.A.) = (1010) α₂; (d) SADP of γ and α₂ phases in the fine lamellae, Z.A. = (110)γ + (1120)α₂. Note that common directions (e.g. (0002)γ) are not parallel in the coarse and fine lamellae; (e) SADP of coarse lamellae showing standard Shoji-Nishiyama orientation relationship, Z.A. = (101)γ + (1120)α₂; (f) SADP of fine lamellae showing standard Shoji-Nishiyama orientation relationship, Z.A. = (101)γ + (110)γ + (1120)α₂. Ti-47Al aged at 1200°C showing (g) coarse lamellae growing approximately perpendicular to fine lamellae; (h) sheared fine lamellae as a result of coarse lamellae.
fully-lamellar γ-based TiAl alloys show that at 800°C, there is very little change in the initially fine structure, at 1000°C the changes are more significant, and at 1200°C dramatic coarsening and spheroidizing occurs. Since the γ and α2 phases in the lamellar structure have a thin, long plate morphology, there appear to be no curvature effects, which implies that the chemical potentials of the atoms are the same everywhere except for growth faults and edges. Small perturbations in the shape of an infinite perfect plate are expected to decay as in the case of a flat surface, so that reducing imperfections would enhance their stability. Conversely, these present experimental observations also indicate that imperfections such as faults, edges and grain boundaries play a key role in initiating instability of the lamellar structure during aging.

The general characteristics of microstructural instabilities in the matrix of all three alloys were similar. This can be understood from the following considerations: (1) when the as-cast alloys were heat-treated at 1400°C in the α phase field, the γ phase dissolves leaving large α grains; (2) furnace cooling causes the formation of the lamellar structure by the precipitation of closely-spaced and parallel γ plates within each α grain. These plates are parallel to each other due to the constraint of the crystallographic orientation relationship described earlier. However, one clear difference in the initial structure among these alloys is that equiaxed γ grains were found along lamellar colony boundaries in the binary and B-modified alloys, but not in the W + B containing alloy. TEM analysis showed that the initial lamellae contain varying degrees of imperfections such as misoriented and kinked lamellae due to low angle boundaries and twinned regions between plates, which in turn are affected by alloying. The interfaces at the broad faces of plates also appear to have irregularities in their dislocation structures.

When these alloys are aged at 800°C, both optical microscopy and TEM show little degradation of the structure. Detailed higher magnification TEM analysis did show that some thermal grooving does occur at the interface between the faults and the broad interphase interfaces, particularly in the Ti-47Al alloy. This grooving appears to lead to fault migration and dissolution of the interface ahead of the fault [Fig. 6(g)]. Dislocation network debris suggests that these networks comprise the interphase interface prior to fault migration. Such fault migration during the instability of lamellar structures has been reported earlier in several other alloy systems as well as in Ti-48Al-2Mo-2Nb alloys [6–12,28]. Continuous lamellar dissolution also began at the broad faces of the plates, and propagated by step migration, whose kinetics would be determined by bulk diffusion and the rates of nucleation and growth of each new step. During aging for 168h at 1000°C, new equiaxed γ grains also formed, often containing spheroidized α2 precipitates, and discontinuous lamellar coarsening observed. These new coarse lamellae were roughly normal to the local transformation interface, and not parallel to the finer lamellar structure that is being completely dissolved as that interface propagates [22]. This discontinuous transformation boundary showed no dislocation network structure or other periodic structure at the

Fig. 9. (a).
interface, suggesting (together with the diffraction analysis) that these boundaries were high angle grain boundaries.

The ratio of grain boundary migration velocity to fault migration velocity ($v$) has been shown by Livingston and Cahn [22] to be proportional to $(D_b \delta \lambda_i)/(D \phi)$, where $D_b$ and $D$ are the grain boundary and bulk diffusivity, respectively, $\delta$ is the grain boundary thickness, and $\lambda_i$ and $\phi$ are the interlamellar spacing before and after coarsening, respectively. If $\phi$ is proportional to $\lambda_i$, then $v$ increases with decreasing temperature since $D_b \delta / D$ increases as the temperature is reduced. The value of $v$ also increases with decreasing $\lambda_i$. Therefore, the
importance of discontinuous coarsening relative to continuous coarsening (by the fault migration mechanism) increases with both decreasing temperature and finer interlamellar spacing (all things being equal and alloying effects not causing or stabilizing the finer structure).

After aging at 1200°C, extensive microstructural coarsening had occurred in all of the alloys, so that the lamellae were now very coarse, branched and irregular relative to the initial structures. Lamellar branching changes the interlamellar spacing without nucleating a new plate [7–9]. Aging at 1200°C also
Fig. 9. BF TEM micrographs of Ti-47Al-40wppm B alloy heat treated at 1400°C for 1 h followed by furnace cooling with (a) lamellar colonies and isolated γ grains within lamellar colonies. Aged at 800°C for 1 week showing (b) lamellar dissolution; (c) dissolution of lamellae as a result of fault and edge migration; (d) higher magnification micrograph of central area of Fig. 9(c) showing dislocation structure at receding lamellae and dislocation network debris. Aged at 1000°C for 1 week; (e) interface between coarse and fine lamellae with thickness fringe contrast, fine lamellae is to the right; (f) different area of interface of the discontinuously coarsened structure showing complex structural contrast in isolated areas (marked with an arrow). Aged at 1200°C for 1 week: (g) spheroidized α₂ and γ phases and branched lamellae.
produces the coarse, spheroidized $\alpha_2$ precipitates within $\gamma$ grains, as well as large equiaxed $\gamma$ grains containing plate-shaped precipitates of the $\alpha_2$ phase. Details of such precipitation have been reported earlier [28]. The ratio ($r$) of the interlamellar spacing before and after discontinuous coarsening in this work varies from 2.5 to 60. It has been pointed out earlier [22,25] that the commonly used extremum principles of maximum growth rate and maximum rate of entropy production used to model thermodynamically irreversible processes are inconsistent with values of $r$ greater than 3. However, in this work, grain boundary migration was accompanied by complex changes in the microstructural morphology.
and the volume fraction and of the phases, and such factors (in addition to alloying effects) may contribute to the difference between the theoretical predictions and the experimental results. The present results also show that the colony boundary migration kinetics varies significantly from one adjacent colony to another in the binary alloy, and alloying, particularly the addition of B + W, significantly affects the kinetics and modes of microstructural change. Earlier work [17,18,25] has demonstrated that this effect is related to the relative misorientation between the colonies forming the boundary.

A summary of the observed modes of microstructural instability is presented in Fig. 11. Figures 11(a)–(c) and 11(g)–(i) represent colony boundary related instabilities, while Figs 11(d)–(f) represent the
Fig. 10. BF TEM micrographs of Ti-47Al-40wppm B-0.5W alloy heat treated at 1400°C for 1h followed by furnace cooling: (a) lamellar colonies. Aged at 800°C showing (b) continuous coarsening by edge migration, dislocation network debris is observed. Aged at 1000°C for 1 week showing (c) growth of a γ grain into the lamellar structure; (d) γ grains with spheroidized α2 and γ precipitates as well as continuous coarsening of lamellae. Aged at 1200°C for 1 week with (e) spheroidized precipitates and (f) remnants of undissolved lamellae.
formation and growth of new equiaxed \(\gamma\) grains. Figures 11(j)–(s) refer to the case of continuous coarsening and dissolution. For the case of grain or colony boundary driven coarsening, the initial lamellar microstructure [Fig. 11(a)] is altered by the growth of pre-existing or new \(\gamma\) grains or by the displacement of the colony boundary into the initial lamellar structure [Fig. 11(b)]. Further boundary growth is accompanied by the formation of new coarse lamellae behind the advancing boundary [Fig. 11(c)]. Alternatively, following the nucleation of \(\gamma\) grains within a lamellar colony [Figs 11(d) and (e)] \(\gamma\) grain growth as well as precipitation of \(\alpha\) within the \(\gamma\) grains occurs [Fig. 11(f)]. One lamellar colony can also grow into an adjacent colony followed by the formation of new secondary lamellae or branching of existing lamellae [Figs 11(g)–(i)]. The relative rates of the migration of the colony boundary compared to the interpenetration of the lamellar colonies may depend on the relative misorientation of the colonies as well as the aging temperature. Lower aging temperatures will favor colony boundary migration over interpenetration since, at low temperatures, grain boundary diffusivity which governs colony boundary migration plays a greater role than lattice diffusivity which controls the rate of interpenetration of colonies.

In the case of continuous coarsening or dissolution of individual lamellae, growing at the fault–interphase interface, followed by faulting is one way in which lamellae can become unstable [Figs 11(j)–(l)]. These faults can occur between plates as a result of twinning at the edge of the plate, growth of a plate through an antiphase boundary or as a result of nucleation and growth of new plates at the edges of old plates. Edges of plates [Figs 11(m) and (n)], steps or ledges [Figs 11(o) and (p)] and interphase interfaces [Figs 11(q) and (r)] are preferential dissolution sites for the plates in the lamellar structure. These processes eventually lead to the formation of spheroidized \(\alpha\) and \(\gamma\) phases.

The AEM results (see Table 2) of the \(\alpha\) and \(\gamma\) phases in Ti–47Al + B and Ti–47Al + W + B alloys show that the compositions of the \(\alpha\) and \(\gamma\) phases in the as-heat treated materials prior to aging are close to those expected in the binary phase diagram slightly above the eutectoid temperature (Fig. 1). This is consistent with the \(\gamma\) phase precipitating from the \(\alpha\) phase in the two-phase region, and the lamellar structure being formed somewhat above the eutectoid temperature. The Ti–47Al + W + B alloys show only a slight difference in W partitioning between the phases, with a little more W in the \(\alpha\) relative to the \(\gamma\) phase (Table 2). Aging for 168 h at 1000°C lowers the Al composition of the \(\alpha\) phase lamellae by about 4at. % (which is an easily measureable difference using this particular AEM equipment), roughly consistent with the relative difference in Al content between the \(\alpha\) and \(\alpha\) phases above and below the eutectoid temperature. This decrease in Al content of the \(\alpha\) phase during aging is observed in both the B and W + B modified alloys, and it is not unreasonable for phase composition changes to coincide with changes in the microstructure during aging. The small W difference measured between the phases initially diminishes during aging in the W + B modified alloy. This microcompositional data indicates that diffusion is changing the phase compositions as the microstructure evolves, and in fact interdiffusion of certain elements like Al and W may be prerequisites to lamellar coarsening and instability. The composition of the spheroidized \(\alpha\) phase particles (these were \(\alpha\) phase at the aging temperature and converted to \(\alpha\) phase upon cooling) developed during aging of the Ti–47Al + W + B alloy at 1200°C is significant in that the Al content is above the initial fine \(\alpha\) phase lamellae in both alloys, but the W content is significantly lower. This suggests that the W or equilibrium composition of the \(\alpha\) phase is actually lower, and that W gets trapped in that phase as the \(\gamma\) phase forms during cooling to produce the initial lamellar structure. The non-equilibrium, higher W content of the initial fine \(\alpha\) phase lamellae may contribute to their observed stability and resistance to coarsening during aging at 1000°C.

The same factors that explain microstructural changes and alloying effects in the bulk material are also applicable to the behavior observed in the "damaged layers" beneath the cut surfaces of the aged specimens. The recrystallized grain portion of the sub-surface damage layer observed after aging at 800 and 1000°C was greatest in the Ti–47Al alloy (up to 100 μm deep at 1000°C) and negligible in the Ti–47Al + W + B alloy. Part of the explanation for this effect of W + B on minimizing the "damage layer" is that those elements stabilize the underlying
Fine lamellar microstructure against coarsening during aging. Another part of that effect may be related to the well-known effect that W has on retarding the recrystallization of cold-worked or deformed structures in steels and in other alloy systems [31]. The effect of W stabilizing the lamellar structure is analogous to such alloying effects in low alloy steels being used to improve the hardenability of steels [32] by slowing down pearlite growth. Our data suggesting that W retards recrystallization in γ-TiAl alloys is also consistent with the experimental observations of Huang [3] and Kim [1] that W and other heavy elements such as Nb and Ta stabilize lamellar structures to improve the creep resistance. Finally, we note that in Ti–48Al–2Mn–2Nb alloys, discontinuous coarsening of the lamellar structure was also often initiated at cut sample surfaces [28]. In several nickel-base alloys it has been shown [18, 19] that recrystallized γ grains play a significant role in discontinuous coarsening and precipitation reactions either through enhanced grain boundary diffusivity or the lack of an orientation relationship between the grains and the adjacent lamellar structure.

Previous work on B effects in γ-TiAl has suggested...
that B retards the kinetics of the $\gamma$ to $\alpha$ transformation [28] during formation of the initial lamellar structure, and then also retards the reversion of finer lamellar back into equiaxed $\gamma$ during aging [29]. Such effects could be related to B affecting interdiffusion of Ti and Al between the phases, as well as the segregation of boron to the $\gamma_2/\gamma$ interface.

This work clearly shows that W + B cause a finer lamellar microstructure to form initially during heat-treatment, and then stabilize that microstructure so that it resists coarsening during subsequent aging. With regard to the modes of microstructure described in this work, these alloying elements retard both continuous and discontinuous lamellar coarsening, prevent the formation of intercolony $\gamma$ during aging at 800 and 1000$^\circ$C, and increase the resistance of the alloy to recrystallization during aging in the “damage layer” below cut surfaces. Such effects on lamellar structure stability during aging could be related to effects of W and B on interfacial energy or thermodynamically stabilizing the $\gamma$ phase in the lamellar structure, thereby retarding an initial critical process that leads to the formation of coarser $\gamma$ phase lamellae or new equiaxed grains. Other recent work on more complex fully-lamellar Ti–47Al–2Cr–2Nb alloys has shown that processing that produces ultrafine and stable lamellar microstructures also produces the best combination of room-temperature strength, ductility and toughness, and high-temperature tensile and creep-strength observed in such $\gamma$-TiAl alloys [33,34], emphasizing the importance of alloying element effects on lamellar refinement.

CONCLUSIONS

The results of the investigation of B and W on the formation of fully-lamellar microstructures in Ti–47Al, Ti–47Al + B and Ti–47Al + W + B alloys, and then the effects of subsequent aging of those structures for 168 h at 800, 1000 and 1200$^\circ$C show that:

(1) Prior to aging, the lamellar structure (consisting of plates of $\alpha_2$ and $\gamma$ phases) in these alloys produced by annealing the as-cast materials for 1 h at 1400$^\circ$C, contained defects such as low angle grain boundaries, misoriented and kinked lamellae. Alloying with B refines the lamellar structure, but with many defects in the $\alpha_2$ phase component of the lamellar structure. Alloying with W + B refines the microstructure more uniformly and completely, with significantly less of the defects observed in the other two alloys.

(2) Aging for 168 h at 800$^\circ$C produced little or no change in the as-heat-treated material of any of the alloys. Aging at 1000$^\circ$C produced significant continuous coarsening in the binary alloy, less in the Ti–47Al + B alloy and almost none in the Ti–47Al + W + B alloy. Aging at 1200$^\circ$C completely coarsened the initial microstructures of all three alloys, but there was the least discontinuous coarsening and large $\gamma$ regions with reprecipitated $\alpha_2$ phase in the Ti–47Al + W + B alloy.

(3) At all aging temperatures, a sub-surface “damage layer” was found, produced in part by deformation during cutting. At 800$^\circ$C, it consisted of mainly fine recrystallized grains, and at 1200$^\circ$C, consisted of primarily $\alpha$-case. After aging at 1000$^\circ$C, the recrystallized grain layer was 60–100$\mu$m thick in the binary alloy, but was negligible in the Ti–47Al + W + B alloy.

(4) The mechanisms of thermal instability were found by TEM and AEM to be related to the defects in the lamellar structure prior to aging. Continuous coarsening and dissolution processes through fault and edge migration, dissolution of lamellae at lamellar interfaces and formation of $\gamma$ grains within the lamellar structure led to microstructural instability at all aging temperatures. After aging at 1200$^\circ$C, spheroidization of the plates was also observed.

(5) Some of the modes of microstructural instability observed included: (a) continuous lamellar coarsening; (b) discontinuous coarsening of the lamellae; (c) migration of the boundaries between lamellar colonies; (d) the growth of equiaxed $\gamma$ grains at the colony boundaries; and (e) advancement of one lamellar colony into another, forming secondary lamellae.

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