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On the Low Tensile Ductility at Room Temperature in High Temperature Titanium Alloys

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Abstract

Several reviews have been written on various classes of titanium alloys including the high temperature titanium alloys. However, there is a gap in comprehensively putting together the research efforts on this important topic of low tensile ductility at room temperature in this class of high temperature titanium alloys used for critical applications in aeroengines. Thus, this review is aimed at bridging this gap. These alloys are designed based on exhausting the solid solubility of the alloying elements (mainly Al, Sn, Zr, Si) in titanium to improve the high temperature properties. Microstructural changes like precipitation of silicides and/or Ti_3Al have been observed in these alloys depending upon their composition, thermomechanical processing, heat treatment and long exposures to service conditions. Several investigations have reported very low tensile ductility at room temperature in these alloys under certain conditions. This has been mainly attributed to either due to "silicides" or "silicides aided by Ti_3Al" or "Ti_3Al exacerbated by silicides" or "Ti_3Al" and surface oxidation at the upper end of service temperatures. This low ductility is very pronounced in the lamellar than in the bimodal microstructures when silicides or Ti_3Al or both occur. When Ti_3Al precipitates are small ($\leq$ ~6nm) and/or precipitation is only in the primary α in the bimodal microstructures, the impact on tensile ductility is very negligible. Alloy design and designing heat treatments are the methods found to be helping to mitigate the reduction in tensile ductility by either avoiding or controlling their size, volume fraction and location of the embrittling phases namely silicides and Ti_3Al. Protective coatings are helping in reducing oxidation and help in avoiding the drop in tensile ductility. More research is required in these areas of alloy design, thermomechanical processing and coatings and also in optimizing the various conflicting requirements in these high temperature titanium alloys.

Keywords: titanium alloys; silicides; Ti_3Al; ductility, microstructure, alpha case, embrittlement, coatings

Abbreviations: HTTA – High Temperature Titanium Alloys

1. Introduction

Different classes of titanium alloys were developed for various applications and reviews have been written on various aspects [1-14]. Some conventional High Temperature Titanium Alloys (HTTAs) developed [1-9] over more than half a century for use in advanced jet engines as compressor discs, blades, vanes, engine crank cases etc. are listed in Table 1. The evolution of HTTAs has been described in the next section.

Aging and/or long-term exposures of these HTTAs cause precipitation of Ti_3Al or silicides or both [15-72]. These precipitates mainly reduce [16, 18, 24, 32-34, 36, 55-72] the Tensile Ductility at Room Temperature (TDRT). TDRT is a very important parameter and a certain minimum ductility is desired by the jet engine designers and this value is proprietary. However, it is common knowledge that low TDRT is not acceptable for the critical aeroengine start operations from cold. TDRT is assessed and understood by the values of the ductility parameters namely % elongation (% total elongation up to fracture and/or % uniform elongation) or % reduction of area (at the fracture) or both obtained from a tensile test. These ductility parameter/s values in these alloys are a function of several factors like prior β grain size [3, 46], bimodal or lamellar structures [1, 3, 5, 44-46], precipitation of embrittling phases like Ti_3Al and/or silicides [16, 19-72] with changes in chemical composition and heat treatment/thermomechanical processing and surface oxidation [1, 3, 5, 37, 46, 73-80]. As an example, Table 2 presents the tensile results on the effects of prior β grain size, lamellar and bimodal microstructures on the room temperature tensile properties [46]. In alloys Ti-5.6Al-4.5Sn-3Zr-1Mo-0.8Nd-0.34Si-0.071O-0.03Fe [23], IMI 834 [46] and Ti-1100 [46], the ductility drop is more in the lamellar than in the bimodal microstructures. While there is a general agreement regarding the impact of these factors on the TDRT, there are differing views regarding the impact of the fine precipitation of silicides or Ti_3Al or both [16-72] in these most important class of alloys. TDRT is also reduced significantly due to surface oxide or alpha case in these alloys [1, 3, 5, 37, 46, 73-80].

It is important to point out that despite several reviews [1-14] on titanium alloys, the research efforts on this important topic of TDRT in HTTAs have not been comprehensively consolidated. Thus, the objective of this review is to summarize and rationalize the findings in the literature mainly with regard to the effect of the precipitation of silicides and/or Ti_3Al and oxidation on the TDRT during the last three decades. The author in his investigations of a near α Titanium alloy IMI 685 observed drastic reduction in TDRT and attributed it to the silicides [55]. In the recent years, investigations [62, 71, 72] have been conducted to mitigate the reduction in TDRT and thus this attempt is made here to summarize the status and also provide some insights and recommendations for further investigations.

2. Evolution of the HTTAs

The development of HTTAs has been presented earlier in some excellent review articles [1, 5, 7, 9]. The philosophy of design of these alloys is based mainly on exhausting the limits of solid solubility (for Al, Sn, Zr, Si) to improve their high temperature capability. In this regard, Rosenberg [15] has empirically defined the aluminum equivalence to be less than 9 wt.% (empirical relation is presented below) to avoid the embrittlement associated with the precipitation of the ordered Ti_3Al phase.

$Al_{eq} = Al + 1/3\ Sn + 1/6\ Zr + 10\ (O+C+2N) \leq 9$

However, coherent Ti_3Al phase has been reported in titanium alloys [16-25] under some conditions even on obeying the above aluminum equivalence of Rosenberg. Further it is important to point out that Zr is a weak β stabilizer [81] and thus it becomes necessary to modify the above empirical relation for aluminum equivalence and consequently this factor may not be favorable for Ti_3Al precipitation.

Small amounts of β stabilizers are added to provide some microstructural strengthening without causing metallurgical instability. While Molybdenum (Mo) has been used to increase heat-treatment response and short-term high-temperature strength [5], Niobium (Nb) is added to improve the surface stability and oxidation resistance at high temperature and improve the β strengthening without affecting the β transus temperature [5].

The addition of small amounts of silicon up to ~0.5 wt. pct. to improve creep resistance and other high temperature properties of HTTAs required for high temperature applications is well established [5, 6, 9, 30-47]. Silicon, in solid solution as well as in the form of silicide dispersions, has been considered important in affecting creep behavior of titanium alloys, depending upon test conditions. There are differing findings on the mechanism by which creep resistance is improved due to silicon [16, 27-32, 36-39, 43].

Creep strength is also increased in the lamellar microstructure obtained by processing in the β phase field (above the β transus) and cooling than the bimodal microstructures (equiaxed primary α and the transformed β structure) that result when these alloys are processed below the β transus (i.e. in the α / β phase field) and in both the cases it is minimum at an optimum cooling rate [1, 5-7, 9, 41, 43]. While some commercial alloys are optimized [3, 44-46] for use in the lamellar microstructures and some others in the bimodal microstructural condition. Kosaka and Fox [6] have published a very comprehensive account of various factors on creep properties.

While the (α+β) treated bimodal microstructures (Fig. 1a) exhibit higher ductility and Low Cycle Fatigue (LCF) properties [1, 3, 5, 7, 33, 46], the β heat-treated lamellar microstructures (Fig. 1b) show superior fracture toughness (FT) and fatigue-crack growth rate or FCGR [1, 3, 5, 7, 9, 32-34, 44-46]. A control of primary α volume fraction to about 15% was found to be optimum in these advanced HTTAs like IMI 834 [5, 46]. Close control of the volume fraction of the equiaxed primary α is achieved by widening the (α+β) phase field and is made possible by small additions of C [47] to help optimize the properties (creep, FT, FCGR, LCF) between the two microstructures namely bimodal and lamellar.

Further refinements in alloy development and thermomechanical optimization are a subject of study [23, 24, 62, 65, 66, 68 -72].

3. Analysis of TDRT from Literature

The reported TDRT reductions due to silicides or Ti_3Al or both have been presented and discussed under different headings. They are Silicides, Silicides aided by Ti_3Al, Ti_3Al aided by silicides and Ti_3Al. Finally, TDRT reduction due to Surface oxidation or alpha case formation.

Further, some observations have been made from the findings as to how this TDRT reduction is mitigated by designing special heat treatments [62] to eliminate precipitation of Ti_3Al or avoid precipitation of Ti_3Al [72] or control their size [24, 70]. Research efforts on oxidation protective coatings [79, 80] have also been examined.

3.1 Silicides

Table 3 presents data from literature that attribute the significant TDRT reduction as due to silicides in the near – α HTTAs namely IMI 685 [55-57] and IMI 829 [58-60] in the heat-treated conditions noted. In the alloy IMI 685 [50], only the incoherent S_2 silicide was present at the martensitic platelet boundaries and there was no precipitation of Ti_3Al in the alloy β solution treated followed by water quenching and aging at 800°C. The absence of Ti_3Al could be rationalized as the temperature of aging could be well above the critical ordering temperature for the formation of Ti_3Al [20]. Thus, the reduction in ductility with the martensitic matrix with no other phases existing was attributed to the role of silicides. The authors [55] showed evidence for planar slip and a heterogeneous deformation with facets in the fracture surface and intense slip bands on the surfaces near to the fracture end of the tensile sample and crystallographically oriented stringers of voids (longitudinal section close to the fracture end) in the aged (S_2 silicides) condition and a very homogeneous slip with very ductile features in the fracture surface, uniform slip traces on the surface close to the fracture end and random voids (longitudinal section close to the fracture end) in the unaged (with no silicides) condition. The mechanism (namely any direct evidence for dislocation silicide interaction) for the reduction of the TDRT was not investigated [55]. Similar observations were noted for other aging temperatures 650°C and 700°C [2, 56].

Thermomechanical processing was done [57] to refine and effect a more uniform distribution of silicides in IMI 685 by aging of the cold worked β solution treated and water quenched material with the idea of improving strength and ductility. The results are presented in Table 3. Although strength properties improved marginally (by about 40 to 80MPa), it did not improve the ductility but rather, it dropped further.

In alloy IMI 829 [58, 59], similar observations as in IMI 685 above were made. The authors [60] showed evidence for the planar dislocations in the material interacting with silicides and also the silicide particles acting as barriers to dislocation motion.

One important point that needs to be highlighted is that despite the aluminum equivalence for the alloy IMI 829 in the study [58-60] being 9.26 (as calculated and shown in Table 3) which is higher compared to 8.66 for the alloy IMI 685 (Table 3), no Ti_3Al precipitation was observed in that alloy for the heat treatments studied.

The variations in the strength parameters are not significant. The marginal overall increase or decrease in strength could be understood as due to the formation of silicides causing precipitation hardening or decrease in the solid solution strengthening owing to the depletion of the elements like Zr, Si, Sn from the matrix to form the silicides [2, 55, 58] . It may be pointed out that the β treated followed by air cooling (829AC in Table 3) strength values are comparable to those reported for the Ex Ti5331S (IMI 829) alloy by Woodfield et al [61]. However, aging (625°C-24 h) of this alloy with higher value of aluminum equivalence of 9.26 by Sridhar and Sarma [58] has showed a lesser value of 858 MPa YS with no Ti_3Al observed as compared to 881.5 MPa for the aged (625°C-2 h-AC-575°C-1000 h-AC) Ex Ti-5331S alloy having lower aluminum equivalence of 8.14 (Table 4) with the precipitation of Ti_3Al by Woodfield et al [61]. Thus, this difference is attributed to the Ti_3Al precipitation hardening due to longer aging schedule in the alloy Ex Ti-5331S by Woodfield et al [61] despite higher aluminum equivalence of 9.26 in the IMI 829 alloy in the study of Sridhar and Sarma [58]. Thus, this aspect from the strength point of view also supports that the TDRT reduction in this alloy for the 625°C-24 h aged condition is attributable to silicides [58-59].

Imbert has studied [16] the embrittlement of the near-α Titanium alloy IMI 684 (in β treated oil quenched and aged at 500°C for 24 h). The TDRT drops were noted for exposures in the temperature range up to about 550°C. The ductility parameter %Reduction of area dropped from 25% at no exposure to 8% for 520°C-300h exposure. This reduction in ductility was attributed to the precipitation of silicides aligned along the martensite plates during the heating. These precipitates at the interplatelet boundaries do not contribute to creep resistance. The very fine silicides that precipitate inside the platelets pin the dislocation motion and contribute to the creep resistance. Thus Imbert [16] observed that there is an interplay between creep resistance and the embrittlement in this alloy. It was also mentioned that though there is precipitation of very fine Ti_3Al particles, it is considered not responsible for the observed reduction in the TDRT. The other factors that are likely to cause embrittlement namely hydrogen, ω-phase, or surface contamination were considered, and no supporting evidence was noted.

Based on the studies in these HTTAs, Popova and Popov [40] state that silicide S_1 precipitated at the α/β platelet boundaries and caused reduction in TDRT and is consistent with other findings [16, 55-60] above.

3.2 Silicides aided by Ti_3Al

Woodfield et al [61] have investigated the near-α titanium alloy IMI 829 (also referred to as Ti 5331S). The tensile results and microstructural differences from two experimental alloys namely Ex Ti 5331S (equivalent to IMI 829) and Ex Ti 5331 (Equivalent to Si free IMI 829) are presented in Table 4 for identical thermomechanical processing conditions (namely "β treated" and "β treated-aged") and compared. It is seen from the Table 4 that there is a considerable reduction in the TDRT parameters (drops of 67.7% in Elongation and 76.3% in reduction of area) in the former alloy namely Ex Ti5331S after aging (showed S_2 silicide and about ~5nm size Ti_3Al precipitates in the "β treated-aged" condition) while the latter namely Si free Ex Ti5331 showed only the Ti_3Al (~5nm) in the same condition with very marginal drop in the ductility values (drops of 3.7% in Elongation and 12.2% in reduction of area in the "β treated-aged" condition). Thus, it was stated that the degree of order that was observed on exposure is not important in reducing ductility. Evidence was shown on the interaction of slip bands on (0001) which had intersected silicide particles at several places along the length of the slip bands resulting in the cracking of silicides or voided at the matrix silicide interface. Essentially, based on all the observations and ordering, Woodfield et al [61] rationalized the observations and mentioned that there is an interplay between ordering and the presence of silicides. It was stated that silicides are the main reason to enhance the evolving of intense slip bands in the presence of small amount of ordering and cause very low TDRT.

The strength parameters Yield Strength (YS) and Ultimate Tensile Strength (UTS) are 7.6% and 8.1% higher respectively in the Ex Ti 5331S (equivalent to IMI 829) alloy compared to Ex Ti-5331 (equivalent to Si free IMI829) with almost no silicon in the unaged condition and is attributed to solid solution strengthening due to Si in that alloy. The strengthening due to silicides and Ti_3Al together through precipitation hardening mechanism in the alloy Ex Ti5331S (equivalent to IMI 829) is negligible at about 2.4% in the YS but it is in the negative in the UTS (indication of embrittling and is treated as fracture strength). Similarly, there is only a 2.4% precipitation strengthening in the YS due to Ti_3Al in the almost no-silicon alloy Ex Ti 5331. UTS showed a similar behavior as in the Ex Ti 5331S. Thus, it appears that the YS strengthening obtained in the Ex Ti 5331S alloy seems to be only due to Ti_3Al while the precipitation hardening effect due to silicides is cancelled out due to the decrease in solid solution hardening (alloying elements were out of solution to form silicides). The experimental alloy showed comparable results with the commercial IMI 829 (for details, please refer to the original paper [61]).

Neal and Fox [36] made similar observations as Woodfield et al [61] in alloy IMI 834. Their observations are (a) less dislocation mobility because of fine heterogeneous silicide precipitation and (b) void formation induced when planar slip bands intersect silicide particles.

3.3 Ti_3Al aided by Silicides

Donlon et al [20] have investigated three different HTTAs namely Ti-1100, IMI 834, and Ti6242S with decreasing aluminum equivalence and Si at a high value in Ti-1100 and lowest in Ti6242S. Table 5 shows TDRT values for the three alloys in different conditions. TDRT values observed were significantly low (1 to 2% Elongation – down from around 10%) for exposures (300, 1000 and 2000h) for all the three alloys at around 600°C in the β processed microstructures. In contrast, the TDRT values for exposures at the same temperature are around 6 to 8% compared to the 12% elongation for the (α+β) solution treated condition. This smaller decrease in TDRT in the (α+β) microstructures as compared to the β microstructures is mainly due to the refined microstructures of the former and thus limiting the slip length leading to increased TDRT. Exposures at temperatures above the Ti_3Al solvus resulted in no decrease of ductility and in some cases, it was noted to increase the ductility. Based on the tensile results and the microstructural analysis, the decrease in TDRT is primarily attributed to the increase in strain localization due to the formation of the coherent Ti_3Al precipitates. However, it was mentioned that the presence of silicides exacerbates the effect.

In alloy Ti-1100, Madsen et al [62, 63] have made similar observations as Donlon et al [20]. The results are presented in Table 5. It may be noted that in the **UNAGED** condition (βST1093°C-30min-AC-593°C-8 h-AC), the TDRT was at 14.9% (reduction of area) where no precipitation (No Ti_3Al and silicides) was observed. This is understandable as the solution treatment temperature is well above the solvus temperatures 1030°C and 740°C for silicides and Ti_3Al respectively and the cooling rate is reasonable to avoid the precipitation of these phases [20]. This value of 14.9% dropped to 8% (reduction of area) in the Post Aging Heat Treatment (**PAHT)** condition (UNAGED-593°C aged 60000mins/750°C-4 h-AC) where only silicides are present (as Ti_3Al has been dissolved at 750°C which is above the critical temperature for dissolution). This is calculated as **46.3%** drop in ductility (from no precipitates condition to a condition with silicides). Similarly, a TDRT reduction to 4.58% (% reduction of area) was observed in the **OVERAGED** condition (UNAGED+593°C-180K minutes-AC). This drop was calculated to be 69.2**%** (for silicides & Ti_3Al condition from no precipitates condition) from the **UNAGED** condition (14.9%). Thus one could calculate the drop due to Ti_3Al by subtracting the "drop due to only silicides (46.3%)" from "drop due to silicides + Ti_3Al (69.2%)". This works out as 22.9% (69.2 minus 46.3). Thus the impact due to Ti_3Al could be considered as only 22.9% as against the 46.3% from the precipitation of silicides. Although the above simple arithmetic calculation may not be very appropriate to look at this complex interplay of different mechanisms, at the same time this may be a reasonable way to bring a perspective to the contributions of each of the microstructural components. If this is the way it can be examined, then the drop in TDRT due to silicides is considered higher than the Ti_3Al in this situation. So, a question may arise if the silicides are causing a more impact in reducing the TDRT than Ti_3Al. If this logic is acceptable, then these observations may fit more into the category of silicides as a major contributor and Ti_3Al is helping out in initiating planar slip as per the mechanism proposed by Woodfield et al [61]. Madsen et al [63] mention that their results do not contradict the mechanism of Woodfield et al [61], however, there is no supporting evidence. The results however show [62] that the formation of Ti_3Al can be suppressed in alloys that contain silicides and thus the TDRT could be improved. The silicide and Ti_3Al precipitation and their sizes are noted in reference [64]. It may be observed from the other findings [24, 70] that below a certain size of Ti_3Al, it is not deleterious to TDRT and is discussed under the heading Ti_3Al.

Cui et al [65] have investigated alloy IMI 834 and observed reduction in strength and TDRT for thermal exposure times of 100 h at temperatures in the range 600 to 750°C. They attributed this reduction as due to planarity of slip caused by Ti_3Al and mention that silicides enhanced the tendency to planar slip. Both coarser silicides and Ti_3Al precipitates decrease strength and ductility at higher temperature exposures.

In a near α Titanium alloy TG6, Cai et al [66] observed that the ductility values reduced by more than 50% due to long term exposure of (α+β) processed material (bimodal microstructures) at 600°C. They attributed this reduction in ductility as mainly due to the precipitation of Ti_3Al during long term exposures and thus promoting planar slip and inhomogeneous deformation. It was observed that the precipitation of silicides may promote the intensity of slip and is a minor reason for the ductility loss.

3.4 Ti_3Al

In this section, "loss of TDRT due to Ti_3Al" and the "effect of size of Ti_3Al on the TDRT" are presented under separate headings.

3.4.1. Loss of TDRT due to Ti_3Al

In one investigation in alloy IMI 834 [67], the alloy was solution treated in the β phase field at 1080°C for 30 minutes and subsequently cooled to 1010°C for 1 h followed by water quenching (**Unaged**). One set of samples were aged at 700°C for 2 h followed by air cooling (**Peak-aged**) and another set was aged at 825°C for 2 h followed by water quenching (**Over-aged**). The matrix microstructure showed grain boundary α and the transformed β. Tensile results are presented in Table 6. The authors attribute the increase in strength parameters in the Peak-aged condition as due to the precipitation of Ti_3Al. The precipitation of the Ti_3Al observed was in the transformed β platelets. It is suggested that the platelets need to be understood as martensitic α' although it is referred to as α platelets. It was mentioned that silicides played only a minor role in the strengthening in the peak-aged condition. In the over-aged condition, the decrease in strength could be mainly due to the elimination of the strengthening phase Ti_3Al. Further, the minor precipitation strengthening effect due to silicides is cancelled out due to the decrease in solid solution strengthening (because of depletion of solid solution strengthening elements Zr, Si in the matrix due to the precipitation of silicides).

The investigators Srinadh et al [67] state that the ductility in the Peak aged material (6.5% Elongation, 8% Reduction of area) with coherent Ti_3Al precipitates and incoherent S_2 silicide precipitates is lower than the Overaged one (7.5% Elongation, 7.5% Reduction of area) with only silicides and attribute this difference to the detrimental role of the coherent Ti_3Al precipitates in the Peak aged condition and refer to the work of Madsen et al [63, 64] in Ti-1100 where the difference in ductility was significant. However, the ductility values between the two conditions in the work of Srinadh et al [67] above are considered equivalent with in the experimental errors in the opinion of the author of this review. Further, if the size of the Ti_3Al is considered as 5nm, it was shown [24, 70] that at this size, the TDRT is not impaired and is discussed in the next section.

The tensile results from Cope and Hill [32] in alloy IMI 834 in (α+β) treated (bimodal microstructures) and aged material are presented in Table 6. When Ti_3Al is precipitated only inside the primary α as in 600°C-4 h aged material, the ductility values are good. At higher aging temperatures, Ti_3Al precipitated in transformed β too. They also became coarser with higher aging temperatures. Then the ductility values were dropped by more than 50%. Cope and Hill [32] stated that the larger and more precipitation of the Ti_3Al in α would increase the propensity to planar slip with in the α phase thus leading to reduced ductility. β spheroidization and silicide precipitation could have also contributed.

It is important to point out that when the ductility values of the 700°C-2 h aged alloy IMI 834 in Srinadh et al's investigation [67] are much lower compared with those of Cope and Hill [32] for the same aging treatment despite the aluminum equivalence being close. This could be because Srinadh et al's [67] material is essentially in the lamellar structure ((with martensitic α' platelets and thin broken primary α along the prior β grain boundaries formed during cooling from β phase field to (α+β) phase field as is evident in the optical microphotographs shown in their paper)) as compared to the bimodal structures in the Cope and Hill's investigation [32].

In alloy VT9, Banerjee et al. [18] attributed the drastic reduction in TDRT in the beta solution treated followed by furnace cooling and annealing at 600°C for 64 h as due to short range order. It was reasoned out that the short-range order is the result of enriching of alpha phase in aluminum due to the partitioning of alloying elements. In the same alloy, for a similar thermal treatment (beta solution treated followed by furnace cooling and annealing at 550°C for 24 h) evidence was noted for both S_2 silicides and Ti_3Al [19].

Jia et al [68] have investigated the effect of the fine precipitation of silicides and Ti_3Al in the alloy Ti60 (very close to IMI 834) in the (α+β) treated /bimodal microstructure on tensile properties. Three different microstructural conditions with regard to the fine precipitation in the bimodal matrix microstructure have been defined. They are (i) fine incoherent silicides along the interplatelet boundaries with no Ti_3Al, (ii) Larger incoherent silicides at the interplatelet boundaries with small coherent Ti_3Al inside the platelets and (iii) larger silicides along with large incoherent Ti_3Al. It was observed that the ductility was not impacted in the microstructural condition (i). However, for the microstructural condition (ii), strength increased, and ductility decreased. The mechanism of increase in strength was attributed to cutting of the coherent Ti_3Al and the planarity of slip caused the reduction in ductility. In the condition (iii), it was proposed that Orowan looping mechanism around Ti_3Al could be operating and thus there are debits in strength and ductility. Essentially, the authors conclude that Ti_3Al is the reason for reduction in TDRT. These results are in line with those of Cope and Hill [32] in the bimodal microstructure in alloy IMI 834. In another investigation of Ti60 alloy, Jia et al [69] find that the drop in ductility in alloy with lamellar structures is higher than the bimodal structures and attribute it to the precipitation of Ti_3Al.

3.4.2. Effect of Size of Ti_3Al on TDRT

Very recently, studies were done to examine the ductility with size of the Ti_3Al in alloys Ti-6.69Al-1.93Sn-3.89Zr-4.6Mo-0.96W-0.23Si [70] and TA29 [24].

The Ti_3Al size and tensile results [70] of Ti-6.69Al-1.93Sn-3.89Zr-4.6Mo-0.96W-0.23Si alloy in the bimodal microstructure with different aging treatments are presented in Table 7. Ti_3Al precipitation in the $\alpha+\beta$ solution treated alloy increased in size from 3nm to 15nm with aging temperature and time. This increase is faster with temperature than with time. The results show that at less than 7nm, the TDRT is not reduced. However, it is noted that the TDRT drastically decreases from 7nm to 15nm. No other precipitate was reported. The investigation did not comment on silicides although silicide precipitation is expected in some of the aging treatments that were studied.

More reduction in TDRT in the β processed microstructure as compared to the above ($\alpha+\beta$) processed microstructure was observed with the size of Ti_3Al as seen in another alloy TA29 [24]. It is important to point out that there was precipitation of only silicide at 750°C-2 h for the lamellar microstructure. However, Ti_3Al also precipitated on further aging at 650°C from 8 h to 1000 h while both silicides and Ti_3Al increased in size with time. Up to a size of 5nm Ti_3Al, the ductility did not reduce when compared with the condition with only the silicide (Table 7).

Woodfield et al (61) also observed negligible impact of 5nm size Ti_3Al in alloy IMI 829 on TDRT.

3.5 Oxidation

Several investigations [3, 33, 36, 46, 73-80] have reported surface oxidation in these HTTAs up on exposure during service or on aging at the higher end of the useful temperature range they are designed for.

It was observed that while UTS and YS were marginally increased due to 600°C exposures for the lamellar and bimodal microstructures, the ductility is mainly affected by the oxygen penetration into the material [73]. Oxygen penetration is faster in bimodal microstructures as compared to lamellar structures [73, 75]. Guan et al [74] have made similar observations of increase in strength and decrease in ductility parameters up on long term exposures close to the service temperatures in a HTTA Ti-55. The reduction in ductility has been attributed mainly to the surface oxidation.

Oxidation studies [46, 75] were done on current commercial HTTAs IMI 834 and Ti-1100. IMI 834 was found to be better at 700°C and higher temperatures [46]. Oxidation studies have been done on Ti64 and other grades [77-79].

4. Mitigation of reduction in TDRT

It is seen from most of the work that silicides and/or Ti_3Al (or α_2) are the reasons for reducing the TDRT. Essentially, four approaches for the factors of silicides and Ti_3Al were noted from the various investigations. Finally, mitigating the surface oxidation is also needed. Thus, they are FIVE approaches:

A. Elimination of Silicides: It is evident that in some alloys silicides are causing the reduction in TDRT. Thus, it is necessary to eliminate the precipitation of the silicides by proper modification of the alloys and also the thermomechanical treatments.

B. Elimination of Ti_3Al: Performing aging treatment/s just above the ordering temperature [62] eliminates the formation of Ti_3Al and thus improves the ductility of the alloy. However, one needs to adjust the composition in a way not to affect the strength and other properties somewhat like in alloy KIMS (72). In this KIMS alloy Ti-6.5Al-3Sn-4Hf-0.4Mo-0.2Nb-0.4Si-0.1B, Hf replaced Zr when compared to the most current advanced commercial HTTA alloys IMI834 and Ti1100. Hf participated in a small measure into the silicides and thus retained the solid solution strengthening effect well. The effects of the Ti_3Al elimination and the Hf modification of silicide and the solid solution effects are presented in Figure 2(a). It is clear that the ductility and strength parameters are higher than the current commercial alloys IMI 834 and Ti-1100 in their respective standard heat treatments compared to KIMS-II (with no Ti_3Al). Figure 2(b) presents a comparison of the elevated temperature tensile (ETT) properties as well. It is seen that the ETT properties are good up to a higher temperature of 650C compared to 600C (approximately) for the current commercial alloys IMI 834 and Ti-1100.

C. Size of Ti_3Al: Designing and controlling the aging treatment could keep the size of the Ti_3Al precipitates to be 6nm or less [24, 70] and thus the ductility is not affected. This aspect was discussed and presented above in the section on "Effect of size of Ti_3Al on TDRT".

D. Limit Ti_3Al precipitation to primary alpha by controlling aluminum and aging [71]: Zhang et al [71] used two alloys (Ti-5.6Al-4.8Sn-2Zr-1Mo-0.35Si and Ti-6Al-4.8Sn-2Zr-1Mo-0.35Si) and demonstrated that in the bimodal microstructural condition, the alloy with lower aluminum aged at 700°C (for times in the range 2 h to 15 h) and exposed to 600°C for up to 100 h did not cause precipitation of Ti_3Al in the transformed β and thus there was negligible effect on the TDRT. However, there was precipitation of Ti_3Al in the transformed β in the alloy with higher aluminum content in the bimodal microstructure aged at 760°C (for times in the range 2 h to 10 h) and exposed to 600°C for 100 h and this resulted in a drop of 30 to 50% in TDRT.

E. Protective coatings: Some efforts (using platinum aluminides coatings) have been noted [79, 80] to mitigate the surface oxidation of HTTAs to considerably reduce the impact in the reduction in TDRT.

Figure 3 is a sketch that presents the entire review in a very comprehensive way for easy understanding. It is self explanatory.

5. Conclusions

A. There is a general agreement that the reduction in TDRT in lamellar microstructures is higher than that in bimodal microstructures.

B. The precipitation of silicides and/or Ti_3Al is deleterious and cause drop in TDRT. It appears that there is general agreement amongst various investigations that Ti_3Al formation causes planar heterogeneous slip and results in the reduction in TDRT.

C. The mechanism of planar slip and heterogeneous deformation when only silicides are precipitating in some of these HTTA alloys is not well understood.

D. One mechanism that was proposed is planar slip bands cutting through the silicides and thus cracking of the silicides causing low TDRT in alloy IMI 829. Small size Ti_3Al that existed has negligible impact on TDRT but helped initiate planar slip.

E. When Ti_3Al is of size about 6nm or less and the precipitation location is primary alpha the effect on TRDT is negligible.

F. By controlling aluminum and aging, Ti_3Al can be controlled or eliminated and the drop in TDRT could be significantly mitigated.

G. As the Ti_3Al dissolution temperature and the silicide solvus temperature seem to depend on the alloy grade chemistry, these need to be evaluated prior to designing the stabilization treatments for both the bimodal and lamellar microstructural conditions.

H. Substitution of Hf for Zr in these alloys has beneficial effect on strength properties due to solid solution strengthening as only very small %Hf is entering silicides.

I. Oxidation has significant effect in reducing TDRT in these HTTAs. Oxidation resistance coatings are mitigating the reduction in ductility.

6. Recommendations

It is necessary to establish the reason for planar slip in those situations where only silicides are attributed to be reducing the TDRT. Is it short rage order is a question and if so that requires evidence.

Certainly, much more data and systematic studies are needed on the effect of size, volume fraction and location of the Ti_3Al precipitation on the TDRT and similarly with regard to silicides too. More studies on coatings to protect from oxidation are needed.

Author Contributions: Ramachandra Canumalla had examined different findings in the literature on this very important subject of low tensile ductility at room temperature in these high temperature titanium alloys of great commercial and critical importance for use in aeroengines, then critically reviewed and written the manuscript with his insights and indicated areas to focus for further work.

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Figure Captions

Figure 1: (a) Bimodal microstructure (b) Lamellar microstructure

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Figure 2: KIMS-I [72] and KIMS-II [72] tensile properties are compared with IMI834 [38] and Ti-1100 [63]. (a) Room temperature (b) Elevated temperatures

The heat treatments for the alloys IMI 834 [38], Ti-1100 [63], KIMS-I [72] and KIMS-II [72] respectively are (α+β)ST1022°C-2 h-OQ/700°C-2 h; βST1093°C-30min.-AC/593°C-8 h-593°C-10000mins; (α+β)ST-1 h-WQ/650°C-5 h; (α+β)ST-1 h-WQ/700°C-5 h

Figure 3: Low Tensile Ductility at Room Temperature (TRDT) has been attributed to the reasons indicated at the five corners of a pentagon in the schematic. The arrow points to the possible FOUR mitigation efforts to improve the TDRT that are being tried. Note: α_2 is Ti3Al

Tables

Table 1: Development of some Conventional High Temperature Ti Alloys [5, 7]

Alloy Designation	Year	Max. Temperature, °C (°F)	Chemical Composition								
			Al	Sn	Zr	Mo	Nb	V	Si	Others	Al.Eq
IMI 318	1954	305 (580)	6	-	-	-	-	4	-	-	6
IMI 550	1956	400 (750)	4	2	-	4	-	-	0.5	-	4.66
IMI 679	1961	450 (840)	2	11	5	1.0	-	-	0.2		6.5
IMI 685	1969	520 (970)	6	-	5	0.5	-	-	0.25	-	6.8
Ti-6242S	1974	480 (900)	6	2	4	2	-	-	0.1	-	7.3
IMI 829	1976	580 (1080)	5.5	3.5	3	0.3	1	-	0.3	-	7.3
IMI 834	1984	590 (1100)	5.5	4	4	0.3	1	-	0.5	0.06C	8.1
Ti-1100	1987	600 (1115)	6	2.8	4	0.4	-	-	0.4	-	7.6
NOTE: Aluminum equivalence (Al.Eq) is calculated with out considering O and N using Al-Eq= Al + 1/3 Sn + 1/6 Zr + 10 (O+C+2N) ≤ 9 [15]											

Table 2: Effect of Prior beta size, Lamellar and Bimodal structures on tensile parameters at Room Temperature

Alloy	Chemistry/beta transus	Al Eq.	Pre-Aging Thermal Process	Aging	Micro-structure	0.2%YS, MPa	UTS, MPa	%El	%RA	Ref.
IMI 834	Ti-5.8Al-4Sn-3.5Zr-0.5Mo- 0.35Si-0.7Nb-0.06C/**1040°C**	8.89	(α+β)ST1025OQ	700°C	Micro 1	1040	1125	9	29	46
			TMT-(α+β)ST1000WQ	600°C-4 h	Micro 2	1200	1255	13	37	
			TMT-βST1080WQ	600°C-4 h	Micro 3	1110	1220	4	9	
NOTE: **ST:** Solution Treatment; **OQ:** Oil Quenching; **WQ:** Water Quenching; **AC:** Air cooling; **(α+β)ST1025OQ:**(α+β)1025°C-20minutes-OQ; **TMT-(α+β)ST1000WQ:** Swaged at 950°C -74% AC (15Vol.% αp)-(α+β)ST1000°C-20min-WQ; **TMT-βST1080WQ:** Swaged at 950°C -74% AC (15Vol.% αp)-βST1080°C-20min-WQ; **Micro 1:** Bimodal: Primary α (15Vol.%/15 to 20μm) & transformed β matrix; **Micro 2:** Bimodal (higher amount of Primary α); **Micro 3:** Lamellar Microstructure										
Ti-1100	Ti-6Al-2.7Sn-4Zr-0.4Mo-0.45Si/**1020°C**	7.6	Forged at 980C-AC	NA	Micro A	900	965	13.5	25	46
			(α+β)ST940WQ	600°C-4 h	Micro B	965	1050	14	28	
			(α+β)ST980WQ	600°C-4 h	Micro C	995	1090	8	18.5	
			βST1020WQ	600°C-4 h	Micro D	1050	1160	7.5	19	
			βST1060WQ	600°C-4 h	Micro E	1080	1190	5	14	
			TMT-(α+β)ST980WQ	600°C-4 h	Micro F	1100	1200	10	34	
			TMT-βST1060WQ	600°C-4 h	Micro G	1150	1250	2.5	6	

NOTE: **ST:** Solution Treatment; **AC:** Air Cooling; **WQ:** Water Quenching; **(α+β)ST940WQ:** (α+β)940°C-20minutes-WQ; **(α+β)ST980WQ:** (α+β)ST980°C-20min-WQ; **βST1020WQ:** βST1020°C-20min-WQ; **βST1060WQ:** βST1060°C-20min-WQ; **TMT-(α+β)ST980WQ:** Swaged at 950°C -83% AC-(α+β)ST980°C-20min-WQ; **TMT-βST1060WQ:** Swaged at 950°C -83% AC-βST1060°C-20min-WQ; **Micro A:** Bimodal: Primary α (15Vol.%/15 to 20μm) & transformed β matrix; **Micro B:** Bimodal (finer than A); **Micro C:** Bimodal (Coarse compared to B but comparable to A); **Micro D:** Lamellar Microstructure (Prior β grain size 200μm); **Micro E:**Lamellar Microstructure (Prior β grain size 500 to 600μm); **Micro F:** Bimodal finer compared to C; **Micro G:** Lamellar microstructure (finer compared to Micro E)
Al. Eq.: Aluminum Equivalent; **YS:** Yield Strength; **UTS**: Ultimate Tensile Strength; **%El:** % Elongation; **%RA:** % Reduction of Area

Table 3: Tensile parameters data at Room Temperature showing "Silicides" as the reason for low TDRT

Alloy	Chemistry/beta transus	Al Eq.	ST	Aging	Precipitation	0.2%YS, MPa	UTS, MPa	%El	%RA	Ref.
IMI 685	Ti-6.18Al-5.27Zr-0.5Mo-0.28Si-0.14O-0.01N-0.024Fe-30ppm H/**1025°C**	8.66	685WQ	NA	α' (martensitic)	919	1058	7.2	21.3	55
			685WQ	800°C-24 h	S_2 (~0.1µm)	954	1038	3.75	4.4	
			685WQ	700°C-24 h	S_2/41.2nm	917.5	1021	5.55	NA	57
			685WQ6CR	700°C-24 h	S_2/38.6nm	980	1064	4.9	NA	
			685WQ12CR	700°C-24 h	S_2/33.4nm	978	1060	2.5	NA	
			685WQ15CR	700°C-24 h	S_2/28.5nm	1025.5	1067	2.65	NA	
IMI 829	Ti-6.1Al-3.3Sn-3.2Zr-1Nb-0.5Mo-0.32Si-0.02Fe-0.0013S-0.026C-0.127O-0.029H/**1030°C**	9.26	829WQ	NA	α' (martensitic)	886	970	10	19	58
			829WQ	625°C-24 h	S_2 only	1005	1023	2	1	
			829AC	NA	Widmanstatten	867	942	10	22	
			829AC	625°C-24 h	S_2 only	858	975	7	10	

NOTE: **ST:** Solution Treatment; **WQ:** Water Quenching; **AC:** Air Cooling; **685WQ:** βST1050°C-30min-WQ; **685WQ6CR:** 6%Cold Reduced; **685WQ12CR:** 12% Cold Reduced; **685WQ15CR**:15% Cold Reduced; **829WQ:**βST1050°C-30min-WQ; **829AC:**βST1050°C-30min-AC;

Al. Eq.: Aluminum Equivalent; **YS:** Yield Strength; **UTS**: Ultimate Tensile Strength; **%El:** % Elongation; **%RA:** % Reduction of Area

Table 4: Tensile parameters data at Room Temperature showing "Silicides aided by Ti_3Al" as the reason for low TDRT

Alloy	Chemistry/beta transus	Al Eq.	ST	Aging	Precipitation	0.2%YS*, MPa	UTS*, MPa	%El*	%RA*	Ref.
Ex Ti-5331S	Ti-5.54Al-3.48Sn-2.95Zr-0.97Nb-0.34Mo-0.28Si-0.095O/**1030°C**	8.14	βST1050°C-40min-AC	NA	NA	861 (a)	977.5 (b)	9.6	24.9	61
			βST1050°C-40min-AC	625°C-2 h -AC-575°C-1000 h -AC	S_2 (~0.1µm)+ Ti_3Al (5nm)	881.5 (p)	953	3.1	5.9	
Ex Ti-5331	Ti-5.51Al-3.48Sn-3.04Zr-0.99Nb-0.33Mo-<0.02Si-0.12O/~**1030°C**	8.37	βST1050°C-40min-AC	NA	NA	800 (c)	904 (d)	9.4	24.5	
			βST1050°C-40min-AC	625°C-2 h -AC-575°C-1000 h -AC	Ti_3Al (5nm)	819.5 (q)	891.5	9.05	21.5	

NOTE: **Ex Ti-5331S** is equivalent to IMI 829; **Ex Ti-5331** (equivalent to IMI 829 with almost no Si); **ST:** Solution Treatment; **AC:** Air Cooling; The values (a) and (b) are 7.6% and 8.1% higher than (c) and (d) respectively; The values (p) and (q) are higher by 2.4% over their respective unaged values. *average of two readings

Al. Eq.: Aluminum Equivalent; **YS:** Yield Strength; **UTS**: Ultimate Tensile Strength; **%El:** % Elongation; **%RA:** % Reduction in Area

Table 5: Tensile parameters data at Room Temperature showing "Ti_3Al aided by silicides" as the reason for low TDRT

Alloy	Chemistry/beta transus	Al Eq.	ST	Aging	Precipitation	0.2%YS, MPa	UTS, MPa	%El	%RA	Ref.
IMI 834	Ti-5.8Al-4Sn-3.5Zr-0.5Mo-0.34Si-0.71Nb-0.06C-0.02Fe-0.12O/**1045°C**	9.51	βST1070°C-1 h -0.5-0.7°C/s	700°C-2 h-AC (10°C/s)	Ti_3Al, Silicides	Not indicated	Not indicated	5	NA	20
			(α+β)ST1020°C-2 h -OQ	700°C-2 h-AC (10°C/s)	Ti_3Al, Silicides	Not indicated	Not indicated	12	NA	
Ti-1100	Ti-5.9Al-2.6Sn-4Zr-0.4Mo-0.42Si-0.02C-0.02Fe-0.08O/**1014°C**	8.42	βST1070°C-1 h -0.5-0.7°C/s	None	Ti_3Al, Silicides	Not indicated	Not indicated	9.4	NA	
Ti6242S	Ti-5.7Al-1.9Sn-3.7Zr-1.9Mo-0.09Si-0.02C-0.03Fe-0.11O/**996°C**	8.25	βST1070°C-1 h-0.1-0.7C/s	None	Ti_3Al, No silicides	Not indicated	Not indicated	10	NA	
NOTE: **ST:** Solution Treatment; **OQ**: Oil Quenching; **AC:** Air Cooling; Samples for tensile tests were aged for 300, 1000 and 2000 h at temperatures in the range 450°C≤ T≤760°C and air cooled for the three alloys IMI 834, Ti-1100 and Ti-6242S. Minimum in ductility (up to ~ 2%) was observed for the exposures at 600°C for the β processed/lamellar microstructures. This behavior is not seen for the (α+β) processed or bimodal microstructural condition (the drop was from 12% to "6 to 8%"). **IMI 834:** The solvus temperatures for Silicides and Ti_3Al are 990°C±10°C and 745°C±5°C for the lamellar microstructures; For the bimodal microstructure, they are 990°C±10°C for silicides and for Ti_3Al it is 795°C±15°C (Primary α) & 735°C±15°C (trans - α); **Ti-1100:** Silicide solvus is 1030°C±10°C; Ti_3Al solvus is 740°C±10°C; **Ti6242S**: Silicide solvus is 945°C±15°C; Ti_3Al solvus is 715°C±15°C;										
Alloy	**Chemistry/beta transus**	**Al Eq.**	**ST**	**Aging**	**Precipitation**	**0.2%YS, MPa**	**UTS, MPa**	**%El**	**%RA**	**Ref.**

Ti-1100	Ti-6Al-2.8Sn-4Zr-0.4Mo-0.45Si-0.078O-0.022C	8.29	βST1093°C-30min-AC	**UNAGED**	No Ti_3Al or Silicides	915	1000	5.5 (εtrue)	14.9	62
			βST1093°C-30min-AC	**OVERAGED**	TI_3Al and Silicides	955	982	0.18 (εtrue)	4.58	
			βST1093°C-30min-AC	**PAHT**	No Ti_3Al+ Only Silicides	895	980	4.15 (εtrue)	8	

NOTE: **ST:** Solution Treatment; **AC:** Air Cooling; **UNAGED:** βST1093°C-30min-AC-593°C-8 h-AC; **OVERAGED:** UNAGED+593°C-180K minutes-AC (Ref: 69: Ellipsoidal S_2 type silicides at about 0.175µm major axis 0.035µm minor axis) Ti_3Al (at 360K mins ~20nm Spherical & so at 180K mins, it could be estimated to be bet. 10 and 15nm -- ref. 69); **PAHT** (*Post Aging Heat Treatment*)**:** UNAGED+593°C-60K minutes-AC-750°C-4 h-AC

Table 6: Tensile parameters data at Room Temperature showing "Ti_3Al" as the reason for low TDRT

Alloy	Chemistry/beta transus	Al Equi.	ST	Aging	Precipitation	0.2%YS, MPa	UTS, MPa	% El	%RA	Ref.
IMI 834	Ti-5.07Al-3.08Sn-3.45Zr-0.31Mo-**0.2**Si-0.66Nb-0.04C-0.02Fe-0.105O-0.0025N-0.004H/**1045°C**	8.17	(α+β)ST#	No Aging	Nil	987	1128	7.5	7.5	67
			(α+β)ST#	700°C-2 h-AC	Ti_3Al, Silicides	1028	1134	6.5	8	
			(α+β)ST#	825°C-2 h-WQ	Only Silicides	980	1098	7.5	7.5	
NOTE: **ST:** Solution Treatment; **AC:** Air Cooling; **WQ:** Water Quenching; **(α+β)ST#:** βST1080°C-30min-cooled to (α+β)ST1010°C-1 h-WQ (essentially this is lamellar microstructure)										
IMI 834	Ti-5.78Al-4.54Sn-4.05Zr-0.70Nb-0.52Mo-0.44Si-0.055C-0.02Fe-0.10O-0.002N/**1045°C**	8.124	(α+β)ST1020°C-2 h-OQ## (12-15%αp)	600-4 h	Ti_3Al**	905	1037	13	21	32
				650-4 h	S_2, Ti_3Al^	953	1075	9.5	13	
				700-4 h	S_2, Ti_3Al”^	933	1060	8.7	12	
NOTE: **ST:** Solution Treatment; **OQ:** Oil Quenching; ## :Bimodal microstructures; **only in αp (primary alpha); ^Ti_3Al in αp and in platelets of transformed β; “size of Ti_3Al precipitates become larger than those at 650°C-4 h.										

Table 7: Effect of Ti_3Al size on tensile parameters at Room Temperature

Alloy	Chemistry/beta transus	Al Equi.	ST	Aging	Precipitation	0.2%YS, MPa	UTS, MPa	%El	%RA	Ref.
WJZ-Ti	Ti-6.7Al-1.93Sn-3.9Zr-4.6Mo-0.96W-0.23Si/**965°C**	7.99	(α+β)ST940°C-2 h-AC	NA	3nm Ti_3Al	1195	1300	18	NA	70
			(α+β)ST940°C-1 h-AC	600°C-2 h	6nm Ti_3Al	1325	1450	18	NA	
				750°C-2 h	7nm Ti_3Al	1230	1250	12	NA	
				750°C-12 h	15nm Ti_3Al	1100	1120	5	NA	
TA29	Ti-5.9Al-3.9Sn-4Zr-0.7Nb-1.4Ta-0.5Si-0.06C/**1050°C**	8.46	βST (at >1050°C)	750°C-2 h	1	955	1062	7.5	14	24
			AR*	650°C-8 h	2	990	1075	6.5	14	
				650°C-100 h	3	975	1060	3	8	
				650°C-500 h	4	1018	1085	2.5	7.5	
				650°C-1000 h	5	975	1060	3.5	6.5	

NOTE: **WJZ-Ti:** the alloy in ref. 70 has been designated; **ST:** Solution Treatment; **AC:** Air Cooling; **AR*:** βST (at >1050°C)-750°C-2 h-AC; **1:** S_2(<0.1μm); **2:** S_2(~0.1μm)+ Ti_3Al 5nm; **3:** S_2(~0.1μm)+ Ti_3Al 8nm; **4:** S_2(~0.1μm)+ Ti_3Al 26nm-longx13nm-short; **5:** S_2(~0.1μm)+ Ti_3Al 20nm

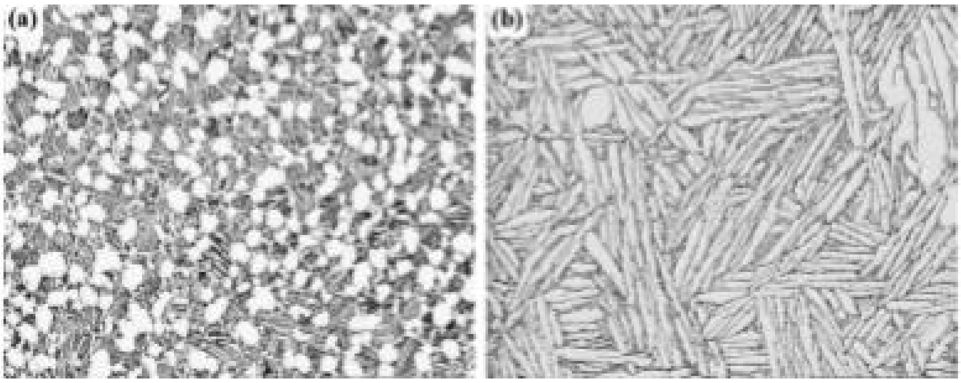

Figure 1: (a) Bimodal microstructure (b) Lamellar microstructure [75]

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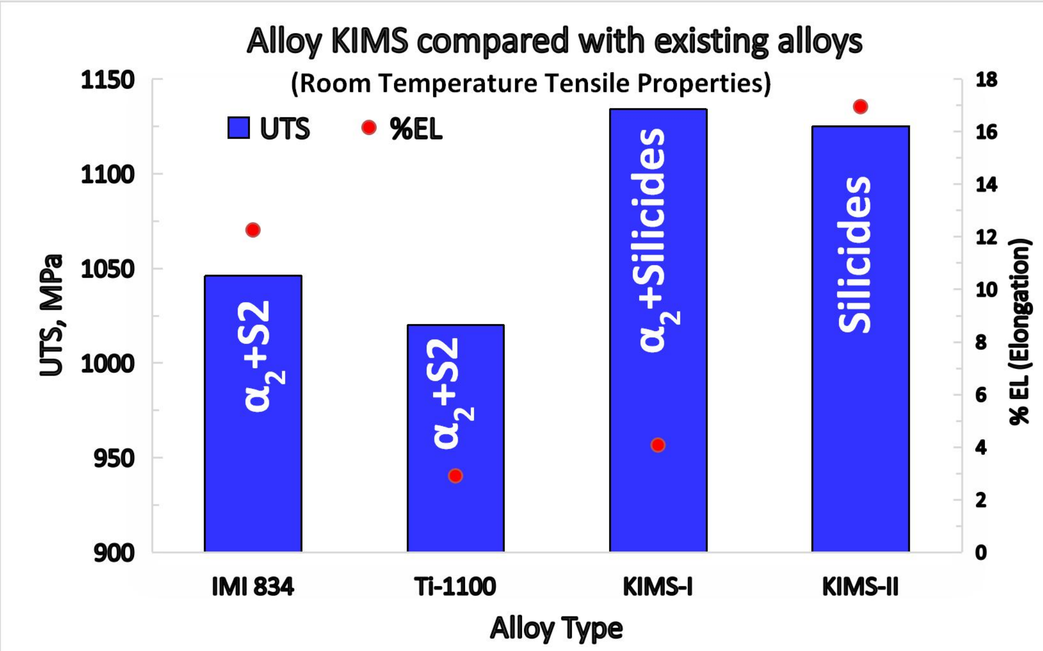
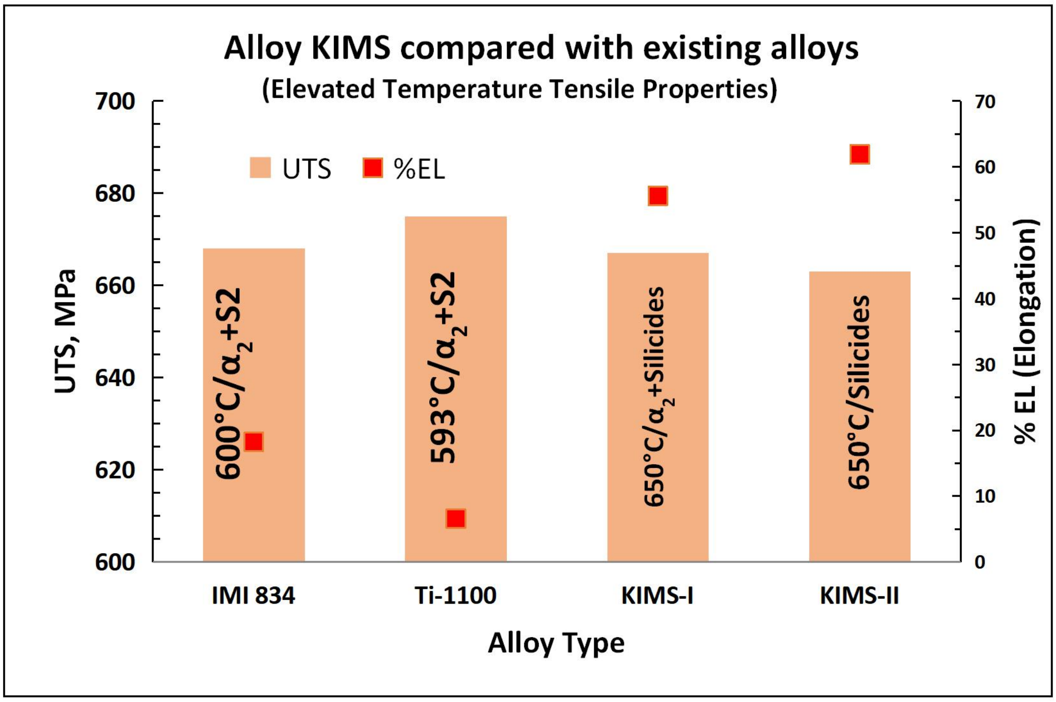


(a)



(b)

Figure 2: KIMS-I [72] and KIMS-II [72] tensile properties are compared with IMI834 [38] and Ti-1100 [63]. (a) Room temperature (b) Elevated temperatures

The heat treatments for the alloys IMI 834 [38], Ti-1100 [63], KIMS-I [72] and KIMS-II [72] respectively are (α+β)ST1022°C-2 h-OQ/700°C-2 h; βST1093°C-30min.-AC/593°C-8 h-593°C-10000mins; (α+β)ST-1 h-WQ/650°C-5 h; (α+β)ST-1 h-WQ/700°C-5 h

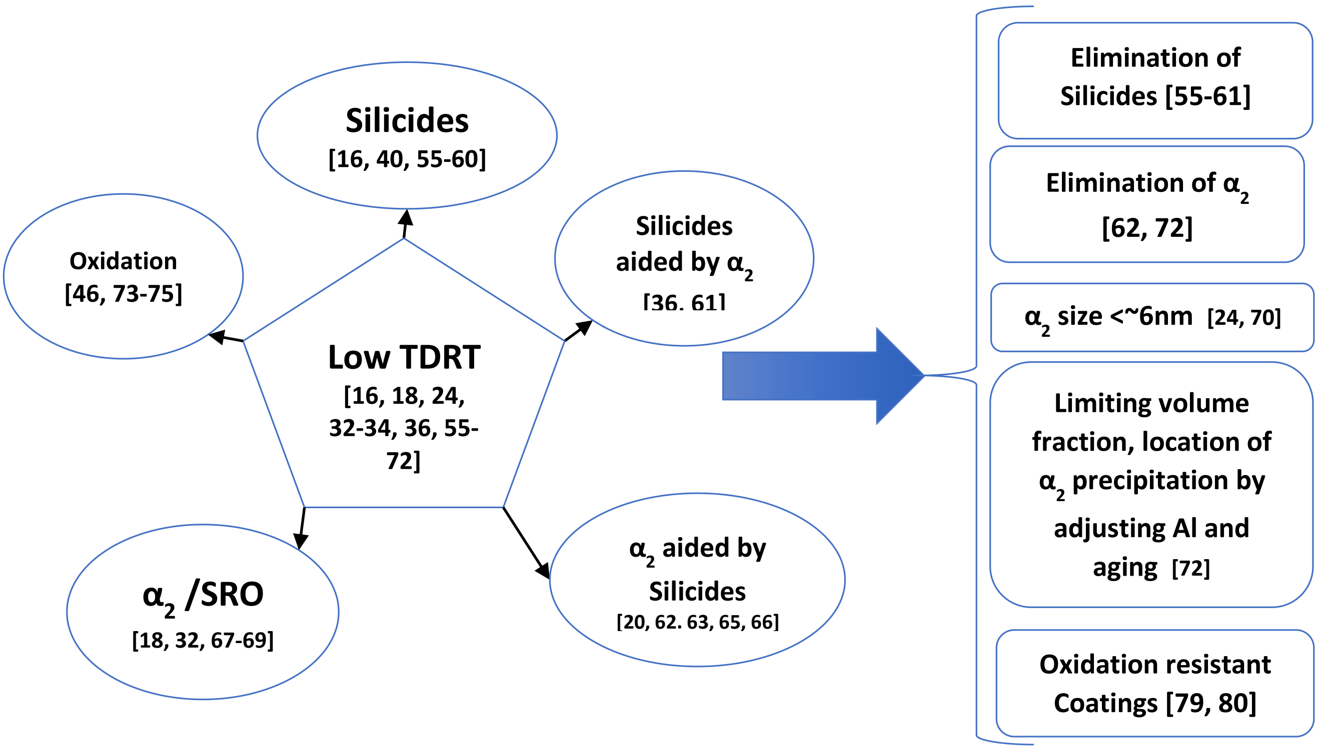


Figure 3: Low Tensile Ductility at Room Temperature (TRDT) has been attributed to the reasons indicated at the five corners of a pentagon in the schematic. The arrow points to the possible FIVE mitigation efforts to improve the TDRT that are being tried. Note: α2 is Ti3Al ; SRO: Short Range Order

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