

A Review of the Ti-Al-Ta (Titanium-Aluminum-Tantalum) System

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This article presents a review of the phase equilibria of the Ti-Al-Ta system. Several investigators have studied this system. The partial liquidus surface of the system and several isothermal sections at temperatures in the range of 1450 to 1000 °C have been reported in the literature. However, detailed studies have been reported for the isothermal section at 1100 °C.

1. Introduction

The alloys based on the Ti-Al-Ta system are potential candidates for high-temperature structural applications. Several investigators have studied the Ti-Al-Ta system. Homogeneous alloys of various compositions have been used for the purpose. Additionally, diffusion couple experiments have yielded information on tie lines. Isothermal sections at six different temperatures have been reported. The study of the solidification behavior of several alloys of the Ti-Al-Ta system has been the basis for the liquidus surface developed for this ternary system.

2. Binary Systems

2.1 Al-Ta

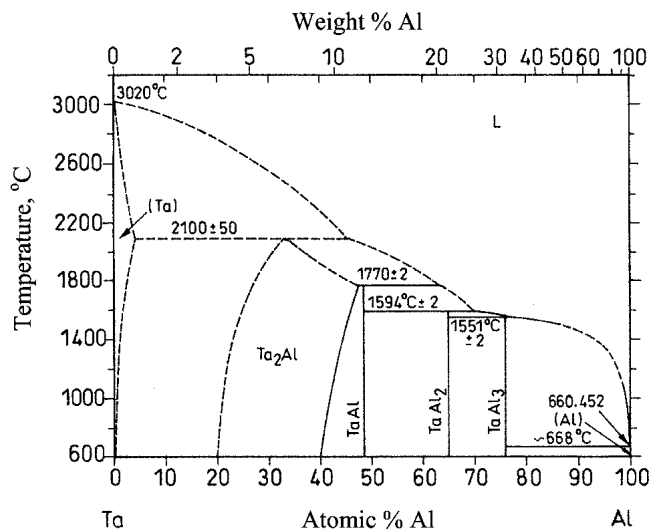
In the Al-Ta binary system, shown in Fig. 1(a) [Massalski2], the terminal solid solutions are the body-centered cubic (bcc) (Ta), with a solubility of <4 at.% Al, and the face-centered cubic (fcc) (Al), with a solubility of <0.04 at.% Ta. The other equilibrium phases present are the liquid, L ; Ta_2Al , having a wide range of homogeneity (20-40 at.% Al at 600 °C); $TaAl$; $TaAl_2$; and $TaAl_3$. Owing to the absence of experimental data for the solidus of (Ta) and Ta_2Al and their mutual solvus lines, these regions are indicated by dashed lines in the assessed diagram. In this binary system, Ta_2Al is formed by a peritectic reaction $(Ta) (4 \text{ at.}\% \text{ Al}) + L (45 \text{ at.}\% \text{ Al}) \rightarrow Ta_2Al (33 \text{ at.}\% \text{ Al})$ at $2100 \pm 50 \text{ }^\circ\text{C}$; the compositions of the (Ta) and L phases are tentative. $TaAl$ forms by a peritectic reaction, $Ta_2Al (48 \text{ at.}\% \text{ Al}) + L (63 \text{ at.}\% \text{ Al}) \rightarrow TaAl (48.5 \text{ at.}\% \text{ Al})$, at $1770 \pm 2 \text{ }^\circ\text{C}$; the first two composition values are tentative ones. The formation of $TaAl_2$ is also accomplished through a peritectic reaction, $TaAl (48.5 \text{ at.}\% \text{ Al}) + L (70 \text{ at.}\% \text{ Al}) \rightarrow TaAl_2 (65 \text{ at.}\% \text{ Al})$, at $1594 \pm 2 \text{ }^\circ\text{C}$. $TaAl_3$ forms through a peritectic reaction, $TaAl_2 (65 \text{ at.}\% \text{ Al}) + L (77 \text{ at.}\% \text{ Al}) \rightarrow TaAl_3 (76 \text{ at.}\% \text{ Al})$, at $1551 \pm 2 \text{ }^\circ\text{C}$. Another peritectic reaction, $TaAl_3 (76 \text{ at.}\% \text{ Al}) + L (99.98 \text{ at.}\% \text{ Al}) \rightarrow (Al) (99.96 \text{ at.}\% \text{ Al})$, occurs at $668 \text{ }^\circ\text{C}$.

A recent assessment of the Al-Ta binary phase diagram by [1996Du] (Fig. 1b) reported various new compounds in the Al-Ta system. The following reactions describe the formation of these compounds: $L (36.3 \text{ at.}\% \text{ Al}) + Ta (8 \text{ at.}\% \text{ Al}) \rightarrow AlTa_2 (24 \text{ at.}\% \text{ Al})$ at $2061 \text{ }^\circ\text{C}$; $L (63 \text{ at.}\% \text{ Al}) + Al_3Ta (75 \text{ at.}\% \text{ Al}) \rightarrow Al_{69}Ta_{39} (63.9 \text{ at.}\% \text{ Al})$ at $1548 \text{ }^\circ\text{C}$; $L (55 \text{ at.}\% \text{ Al}) \rightarrow Al_{69}Ta_{39} (63.9 \text{ at.}\% \text{ Al}) + AlTa_2 (35 \text{ at.}\% \text{ Al})$ at $1499 \text{ }^\circ\text{C}$; $Al_{69}Ta_{39} (63.9 \text{ at.}\% \text{ Al}) + AlTa_2 (35 \text{ at.}\% \text{ Al}) \rightarrow AlTa (50 \text{ at.}\% \text{ Al})$ at $1446 \text{ }^\circ\text{C}$; $Al_{69}Ta_{39} (63.9 \text{ at.}\% \text{ Al}) + AlTa (50 \text{ at.}\% \text{ Al}) \rightarrow Al_7Ta_5 (58.4 \text{ at.}\% \text{ Al})$ at $1345 \text{ }^\circ\text{C}$; $Al_{69}Ta_{39} (63.9 \text{ at.}\% \text{ Al}) + Al_7Ta_5 (58.4 \text{ at.}\% \text{ Al}) \rightarrow Al_3Ta_2 (60 \text{ at.}\% \text{ Al})$ at $1226 \text{ }^\circ\text{C}$; $Al_{69}Ta_{39} (63.9 \text{ at.}\% \text{ Al}) \rightarrow Al_3Ta (75 \text{ at.}\% \text{ Al}) + Al_3Ta_2 (60 \text{ at.}\% \text{ Al})$ at $1183 \text{ }^\circ\text{C}$; $L (99.975 \text{ at.}\% \text{ Al}) + Al_3Ta (75 \text{ at.}\% \text{ Al}) \rightarrow (Al) (99.55 \text{ at.}\% \text{ Al})$ at $662 \text{ }^\circ\text{C}$; and $L (75 \text{ at.}\% \text{ Al}) \rightarrow Al_3Ta (75 \text{ at.}\% \text{ Al})$ at $1608 \text{ }^\circ\text{C}$.

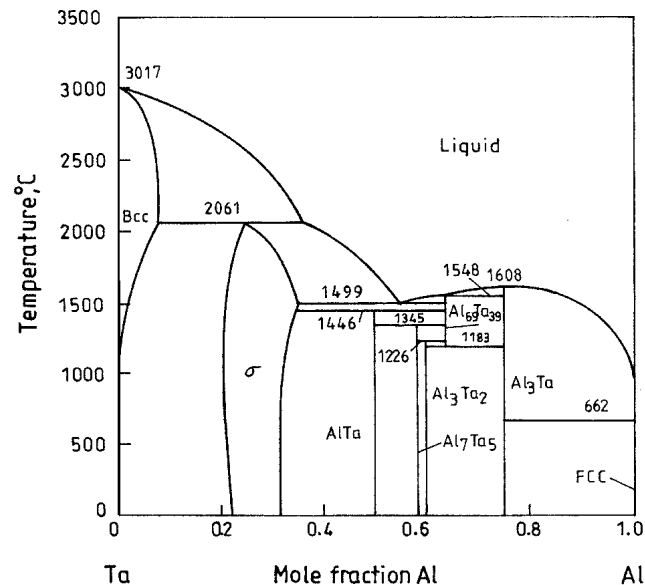
2.2 Al-Ti

The diagram of [1992Kat] is basically the diagram of [Massalski2] with the dashed lines of the latter changed to solid lines by combined experiment and calculation. The equilibrium solid phases present in the Ti-Al binary system (Fig. 2a) [1992Kat] are: (a) the terminal solid solutions of β Ti (bcc) and α -Ti (close-packed hexagonal), with the maximum solubilities of aluminum being 44.8 and 47.3 at.%, respectively; (b) $Ti_3Al (\alpha_2)$, an ordered $D0_{19}$ structure with a homogeneity range of 20 to 38.2 at.% Al; (c) $TiAl (\gamma)$, an ordered $L1_0$ phase with a homogeneity range of approximately 46.7 to 66.5% Al; (d) $TiAl_2$ and Ti_2Al_5 , two stoichiometric phases; (e) $TiAl_3$, an ordered $D0_{22}$ structure with a homogeneity range of 74.2 to 74.6% Al; and (f) the Al (fcc) terminal solid solution with a maximum Ti solubility of about 0.6 at.%. The various invariant equilibria of the Ti-Al system, as described by [1992Kat], are expressed by the following reactions: $L (49.4 \text{ at.}\% \text{ Al}) + (\beta Ti) (44.8 \text{ at.}\% \text{ Al}) \rightarrow (\alpha Ti) (47.3 \text{ at.}\% \text{ Al})$ at $1490 \text{ }^\circ\text{C}$; $L (55.1 \text{ at.}\% \text{ Al}) + (\alpha Ti) (51.4 \text{ at.}\% \text{ Al}) \rightarrow TiAl (55.0 \text{ at.}\% \text{ Al})$ at $1462.8 \text{ }^\circ\text{C}$; $L (72.5 \text{ at.}\% \text{ Al}) + TiAl (66.5 \text{ at.}\% \text{ Al}) \rightarrow Ti_2Al_5 (71.4 \text{ at.}\% \text{ Al})$ at $1415.9 \text{ }^\circ\text{C}$; $L (79.1 \text{ at.}\% \text{ Al}) + Ti_2Al_5 (71.4 \text{ at.}\% \text{ Al}) \rightarrow TiAl_3 (74.6 \text{ at.}\% \text{ Al})$ at $1392.9 \text{ }^\circ\text{C}$; $L (99.9 \text{ at.}\% \text{ Al}) + TiAl_3 (75.0 \text{ at.}\% \text{ Al}) \rightarrow Al (99.4 \text{ at.}\% \text{ Al})$ at $664.2 \text{ }^\circ\text{C}$; $(\alpha Ti) (30.9 \text{ at.}\% \text{ Al}) \rightarrow Ti_3Al (30.9 \text{ at.}\% \text{ Al})$ at $1164 \text{ }^\circ\text{C}$; $(\alpha Ti) (39.6 \text{ at.}\% \text{ Al}) \rightarrow Ti_3Al (38.2 \text{ at.}\% \text{ Al}) + TiAl (46.7 \text{ at.}\% \text{ Al})$ at $1118.5 \text{ }^\circ\text{C}$; $TiAl (64.5 \text{ at.}\% \text{ Al}) + Ti_2Al_5 (71.4 \text{ at.}\% \text{ Al}) \rightarrow TiAl_2 (66.7 \text{ at.}\% \text{ Al})$ at $1199.4 \text{ }^\circ\text{C}$; and Ti_2Al_5

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(a)



(b)

Fig. 1 (a) Al-Ta binary phase diagram after [Massalski2] and (b) Al-Ta binary phase diagram after [1996Du]

(71.4 at.% Al) \rightarrow TiAl_2 (66.7 at.% Al) + TiAl_3 (74.2 at.% Al) at 990.0 °C. Recent assessments of the Al-Ti binary phase diagrams by [2000Ohn] and [2002Pal] are shown in Fig. 2(b) and (c), respectively. It is obvious that the system is complicated and there are conflicting data. Oxide contamination is suspected as being the root cause of the differences. For the present, the Kattner diagram [1992Kat] is preferred.

2.3 Ta-Ti

The Ta-Ti binary phase diagram is that of a simple isomorphous type. Data on the liquidus are lacking, and the solid-phase boundaries are both mutually contradictory and inconsistent with the thermodynamic properties of Ti. Ther-

modynamic calculation has played a major role in the construction of the assessed phase diagram, shown in Fig. 3 [Massalski2]. There are large discrepancies in the melting temperatures, which are mainly due to the effect of alloy contamination. The maximum solubility of Ta in (α Ti) is about 3 ± 2 at.% at 600 °C. The excess Gibbs energies being positive and the location of the β Ti transus being above a metastable β Ti miscibility have resulted in the absence of compounds. The assessed phase diagram has been drawn by utilizing the calculations used for the construction of the Gibbs energy functions that reproduce the α Ti solvus and the approximate metastable β Ti miscibility gap. In the Ti-Ta alloys, the ω phase forms only during tempering of the bcc phase near 400 °C. The critical (metastable) reaction, (β Ti,Ta) \rightarrow (β Ti) + (Ta), occurs at 450 °C, where the (β Ti,Ta) phase has 50 at.% Ta.

3. Solid Phases

The crystal structure data of the equilibrium phases in the Ti-Al-Ta system are shown in Table 1 [Pearson3].

4. Ternary System

The Ti-Al-Ta system has been studied by several investigators: [1966Ram], [1983Sri], [1991Boe], [1991Das1], [1991Das2], [1991Das3], [1992Jew], [1992McC], [1993Das], [1993Jew], and [1995Wea]. Isothermal sections have been proposed by [1966Ram], [1983Sri], [1991Boe], [1992McC], [1993Das], and [1995Wea]. [1992McC] has studied the solidification behavior of the alloys in this system and evaluated the partial liquidus surface. [1991Das1], [1991Das2], [1991Das3], [1993Das], and [1993Jew] have used diffusion couples of Ta/ γ -TiAl along with homogenized bulk alloys for the study of phase equilibria.

5. Liquidus Surface and Solidification

[1992McC] used several alloys of the Ti-Al-Ta system to study the solidification behavior. The composition and primary phases of the various alloys are given in Table 2. Figure 4 is the partial liquidus surface of the Ti-Al-Ta system on which the alloy compositions listed in Table 2 are plotted. Alloys of group B are found to consist of β as the primary phase, and thus are shown on the β liquidus surface in Fig. 4. Group A alloys consisting of primary α phase fall on the α liquidus surface. Alloys G1 to G3, consisting of primary γ phase, are shown on the γ liquidus surface. The arrows indicate the direction of decreasing slope with temperature. The direction of the arrows indicates that the liquid compositions move toward the Ti-Al binary side during solidification. Sudden changes in slope on the ternary liquidus surface are representative of the binary peritectic reactions $L + \beta \rightarrow \alpha$ and $L + \alpha \rightarrow \gamma$. The H group of alloys consists of η (TiAl_3) primary phase and hence is shown on the η liquidus surface in Fig. 4. The γ and η surfaces have opposite slopes, and they meet at a "valley," indicating that solidification along this line leads to a mixture of γ and η

Section I: Basic and Applied Research

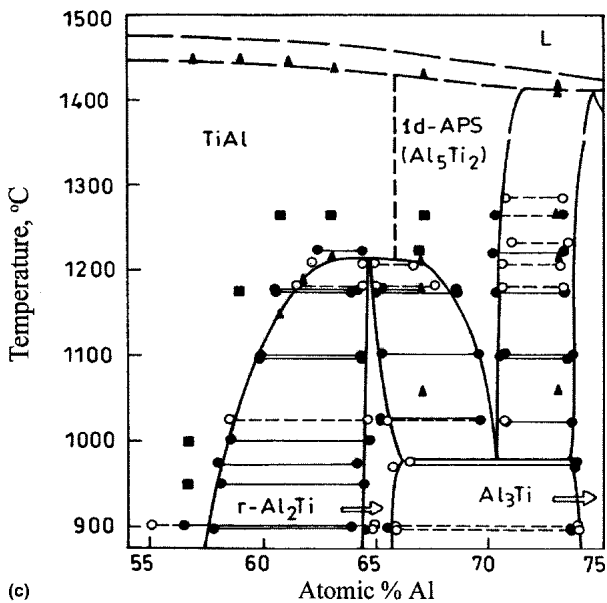
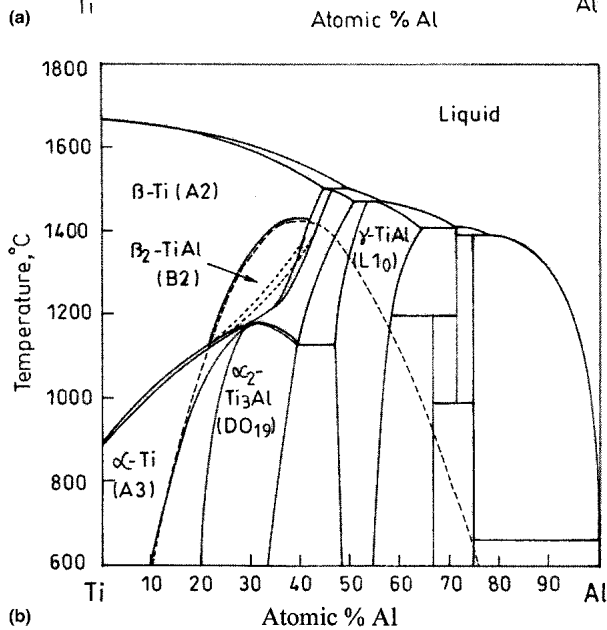
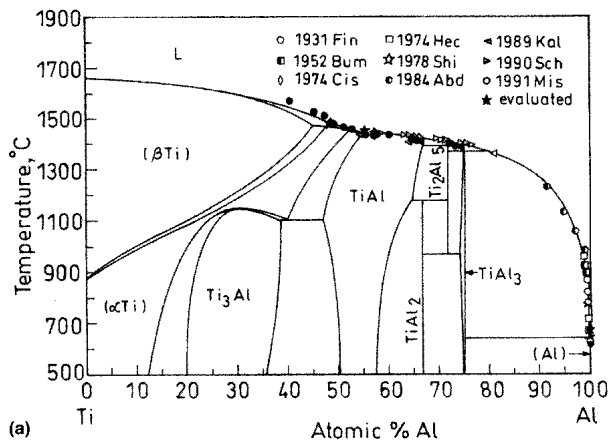


Fig. 2 (a) Al-Ti binary phase diagram from [1992Kat], (b) Al-Ti binary phase diagram from [2000Ohn], and (c) Al-Ti binary phase diagram from [2002Pal]

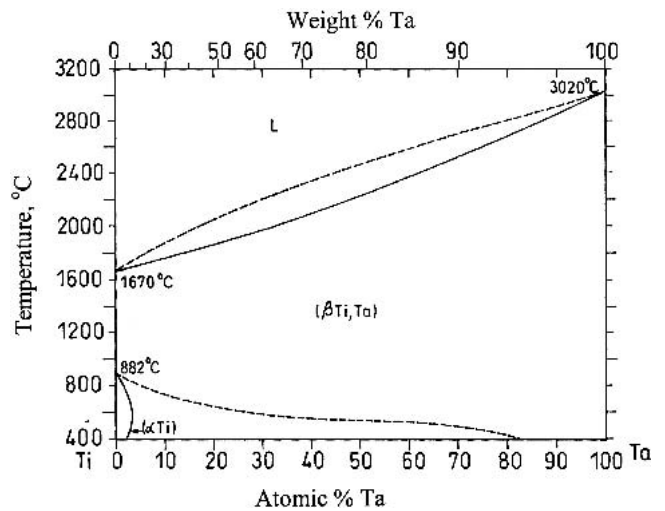


Fig. 3 Ta-Ti binary phase diagram from [Massalski2]

phases in the final aggregate. In the construction of Fig. 4, [1992McC] indicated that they used the phase diagram of [1990Cha]; in that case, there is an inconsistency in [1992McC]. [1990Cha] reported no eutectic reaction in the system but rather an intermediate phase between TiAl and TiAl₃ for which there is no liquidus in Fig. 4.

6. Isothermal Sections

Several investigators have evaluated the isothermal section of Ti-Al-Ta system at 1100 °C. The isothermal sections at 1450, 1440, 1350, 1330, and 1000 °C have also been reported.

6.1 At 1450 °C

[1995Wea] prepared 27 ternary alloys using 99.999% Al, 99.985% Ti, and 99.985% Ta as the starting materials. The alloys were made by arc melting under an argon atmosphere. Oxygen and nitrogen contents in the as-cast samples were in the range of 100 to 331 parts per million (ppm) and 100 to 380 ppm by weight, respectively. Samples were heat treated at 1550, 1450, and 1350 °C in flowing argon atmosphere for various lengths of time in the range of 1 to 72 h followed by water quenching. The various analytical methods that have been used for phase identification and analysis were x-ray diffraction (XRD), scanning electron microscopy (SEM), electron probe microanalysis, and transmission electron microscopy (TEM). Phase equilibria at 1450 and 1350 °C were evaluated. Figure 5 is the 1450 °C isothermal section, which was evaluated on the basis of the experimental analysis of several alloys used in the study of [1995Wea]. At this temperature, the β phase region is quite extensive. Analyses of some of the alloys indicated the presence of an ordered B2 phase after quenching. Thermal anti-phase boundaries were present in the microstructure of these alloys, indicating that these alloys are disordered at 1450 °C and become ordered on quenching. Most of the two-phase regions shown in Fig. 5 were experimentally observed. The remaining two-phase regions of α + η, α + γ,

Table 1 Crystallographic data of phases in the Ti-Al-Ta system [Pearson3]

Composition	Pearson symbol	Space group	Prototype	Lattice parameters, nm		
				<i>a</i>	<i>b</i>	<i>c</i>
αTi	<i>hP</i> 2	<i>P</i> 6 ₃ / <i>mmc</i>	Mg	0.29508	...	0.46855
βTi	<i>cI</i> 2	<i>Im</i> 3̄ <i>m</i>	W	0.33065
Al	<i>cF</i> 4	<i>Fm</i> 3̄ <i>m</i>	Cu	0.40488
Ta	<i>cI</i> 2	<i>Im</i> 3̄ <i>m</i>	W	0.33025
TiAl	<i>tP</i> 4	<i>P</i> 4/ <i>mmm</i>	AuCu	0.4005	...	0.4070
TiAl ₂	<i>tI</i> 24	<i>I</i> 4 ₁ / <i>amd</i>	Ga ₂ Hf	0.3976	...	2.4360
TiAl ₃	<i>tI</i> 8	<i>I</i> 4/ <i>mmm</i>	Al ₃ Ti	0.3848	...	0.8596
Ti ₃ Al	<i>hP</i> 8	<i>P</i> 6 ₃ / <i>mmc</i>	Ni ₃ Sn	0.5782	...	0.4629
Ti ₅ Al ₁₁	<i>t</i> * [*]	0.3843	...	3.3465
Ti ₉ Al ₂₃	<i>tI</i> * [*]	0.3917	...	1.6524
Ta ₁₂ Al ₁₇	<i>cF</i> * [*]	1.9315
Ta ₂ Al	<i>tP</i> 30	<i>P</i> 4 ₂ / <i>mnm</i>	CrFe	0.9864	...	0.5215
TaAl ₃	<i>tI</i> 8	<i>I</i> 4/ <i>mmm</i>	Al ₃ Ti	0.3839	...	0.8535
Al-Ti-Ta	<i>tI</i> 8	<i>I</i> 4/ <i>mmm</i>	Al ₃ Ti	0.3844	...	0.8576

$\gamma + L$, and $\eta + \gamma$ were approximated on the basis of the low-temperature observations of earlier investigators [1991McC], [1992McC], [1991Boe], [1991Das2], [1993Das], [1991Das1], and [1991Das3], and also using the results of [1995Wea]. The three-phase regions, $\alpha + \eta + \gamma$ and $\alpha + \eta + \delta$, were experimentally established. However, the three-phase regions, $\alpha + \gamma + L$, $\alpha + \beta + \sigma$, $\alpha + \delta + \sigma$, and $\gamma + \eta + L$, were estimated on the basis of earlier investigations by [1991Das2], [1993Das], [1991Das1], [1991Das3], [1991McC], [1992McC], [1992Jew], [1991Wea], and [1991Boe]. The isothermal section (Fig. 5) does not include the extension of the TaAl phase field into the ternary system. According to a binary Al-Ti phase diagram [1992Kat], the γ phase field should exist at this temperature for the binary Al-Ti system, thus the γ phase field should be extended all the way to the Ti-Al binary, not as an island in the phase diagram, as is shown in Fig. 5. The phases that should be present in the binary Al-Ta side, according to recent assessment [1996Du], are β (Ta), Ta₂Al (σ), Al₆₉Ta₃₉, Al₃Ta (η), and L (Al).

6.2 At 1440 °C

Figure 6 shows the Ti-Al-Ta isothermal section at 1440 °C, as determined by [1993Das]. This section is based on the experimental results of [1993Das] combined with the data of other research groups [1991Boe, 1991Wea]. The α , β , γ , σ , δ , η , and L constitute the single-phase fields at 1440 °C. It is most likely that the γ phase extends into the ternary phase >12 at.% Ta, as currently shown. The η phase extends from the Ta-Al binary at the composition of Al-25at.%Ta to about Al-15at.%Ti-10at.%Ta and has an average width of about 4 to 5 at.% Al. The α phase has an extensive solubility into the ternary system. As the α phase extends from the Ti-Al binary to the $\sigma + \delta + \alpha$ three-phase field, the width remains almost constant at about 5 at.% Al. The three-phase fields identified at this temperature are $\alpha + \delta + \sigma$, $\gamma + \eta + L$, $\alpha + \beta + \sigma$, $\alpha + \eta + \delta$, and $\gamma + \eta + \alpha$, of which only the first two are shown in Fig. 6. The $\alpha + \beta + \sigma$ and $\alpha + \eta + \delta$ three-phase fields can be extrapolated from the given two-

Table 2 Bulk compositions and primary phases present in the as-cast alloys [1992McC]

Alloy	Bulk composition(a), at. %		Primary phase
	Al	Ta	
B1	48.0	10.2	β
B2	47.9	15.2	β
B3	45.5	27.0	β
A1	53.8	10.2	α
A2	55.8	16.2	α
A3	58.9	20.9	α
G1	67.4	...	γ
G2	63.1	10.0	γ
G3	64.6	15.2	γ
H1	70.6	5.0	η
H2	70.1	10.2	η

(a) Determined by EDS: balance, titanium

phase tie lines. The location of the $\alpha + \gamma + \eta$ three-phase field cannot be estimated due to the lack of samples in this region.

The Ti-Al-Ta isothermal section at 1444 °C by [1993Das], as described in the preceding paragraph, has been modified in the present article by adding some three-phase triangles (superimposed on Fig. 6 by dash-dot lines) for the readers' easy visualization of the phase relations. It is also to be noted that the Al-Ta binary results used in Fig. 6 are inconsistent with the newest Al-Ta binary phase diagram, because the 1440 °C isothermal section [1993Das] was published in 1993, whereas the newest binary Al-Ta by [1995Du] was published in 1996. The Al₆₉Ta₃₉ and AlTa phases are missing from the Al-Ta binary side. It must be noted that the σ phase field of Fig. 6 is quite different from that of Fig. 5, which is very difficult to explain for the 10 °C difference in the two isothermal sections.

6.3 At 1350 °C

None of the alloys that were heat treated at 1350 °C [1995Wea] were single-phase. However, using the tie-line

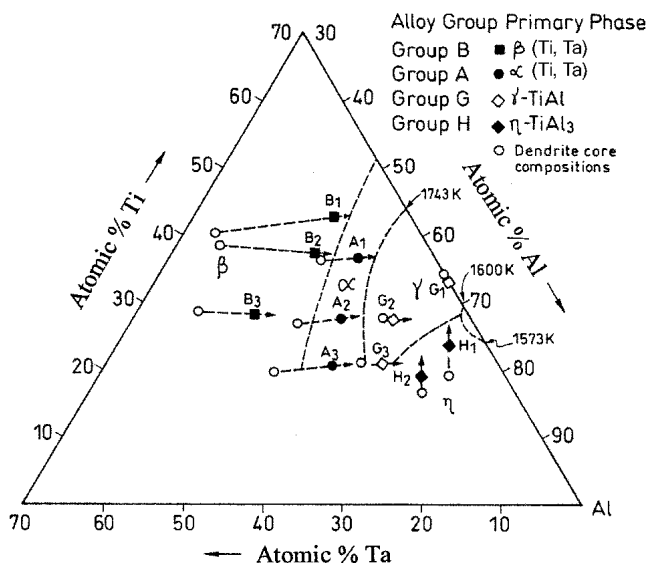


Fig. 4 Partial liquidus surface of the Ti-Al-Ta system. O, the dendrite core compositions for each alloy [1992McC]

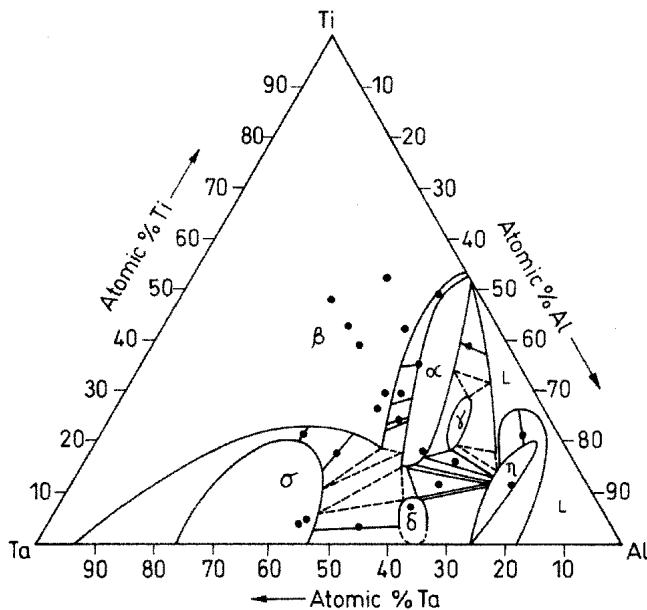


Fig. 5 Isothermal section of the Ti-Al-Ta system at 1450 °C [1995Wea]

and tie-triangle locations of the alloys studied, the single-phase regions at 1350 °C were determined, as shown in the isothermal section in Fig. 7. The single-phase regions thus identified are α , β , γ , σ , and η . More than two phases were not observed in any of the alloys studied at this temperature. Hence, the three-phase regions were estimated on the basis of observations at lower temperature made by [1991Boe], [1991McC], [1992McC], [1991Das1], [1991Das2], [1991Das3], and [1993Das]. The dotted lines in the Fig. 7 indicate the phase boundaries determined from the references stated above. With respect to Fig. 7, one must be aware of the fact that the Al-Ta and Al-Ti side of the ternary diagram are not consistent with the recently published Al-

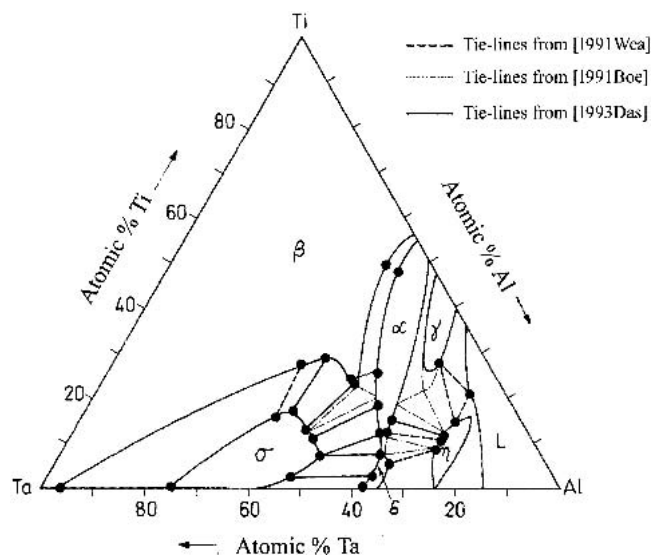


Fig. 6 Isothermal section of the Ti-Al-Ta system at 1440 °C [1993Das, 1993Jew]

Ta [1996Du] and Al-Ti [1992Kat] binary phase diagrams. The $Al_{69}Ta_{39}$ phase from the Al-Ta side and the Ti_2Al_5 and $TiAl_3$ phases from the Al-Ti side are missing.

6.4 At 1330 °C

[1991Boe] prepared arc-melted alloys in the vicinity of the Al_2TiTa composition using 99.99% pure components (i.e., Al, Ti, and Ta). A heat treatment at 1400 °C for 3 h followed by treatment for 1 h at 1525 °C was used to eliminate microsegregation due to casting. Subsequently, the equilibrium was ensured by isothermal holding for times up to 23 h at various temperatures. Samples were subjected to optical metallography, SEM, differential thermal analysis (DTA), and TEM studies. The partial isothermal section of the Ti-Al-Ta system estimated at 1330 °C is shown in Fig. 8. Data obtained at this temperature were used to determine the $\alpha + \gamma + \sigma$ three-phase region. The position of the $\alpha + \beta + \sigma$ three-phase region has been estimated with the help of higher temperature data. The $\sigma + \gamma + \eta$ triangle has been taken from [1991Das1] and is included in Fig. 8 to complete the topology. The binary Al-Ti side is not consistent with the recent assessment of the Al-Ti binary system. The Ti_2Al_5 phase is missing from the binary Al-Ti side.

6.5 At 1100 °C

[1983Sri] prepared the alloys by sintering powder mixtures of the respective elements. Samples were also prepared by arc-melting and subsequent annealing. In the 1100 °C isothermal section of the Ti-Al-Ta system proposed by [1983Sri], as shown in Fig. 9, no ternary phase is present. A complete series of solid solutions between $TiAl_3$ and $TaAl_3$ is indicated by the single-phase field $(Ti,Ta)Al_3$. The formation of these solid solutions takes place through the peritectic reaction $L + TiAl(Ta) + \sigma \rightarrow (Ti,Ta)Al_3$. These solid solutions are nearly ideal because ΔV^{mix} is almost zero. A complete series of solid solutions also occurs, with βTi and

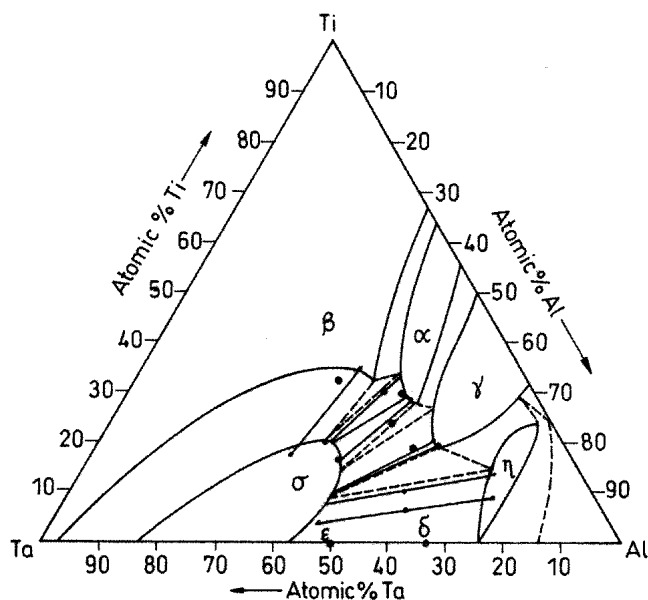


Fig. 7 Isothermal section of the Ti-Al-Ta system at 1350 °C [1995Wea]

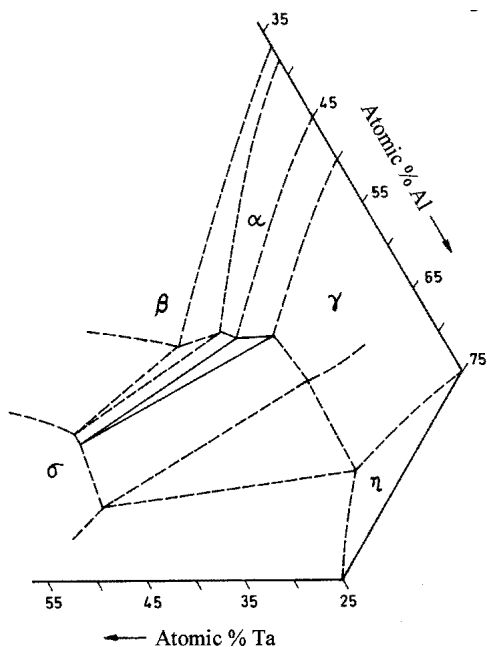


Fig. 8 Partial isothermal section of the Ti-Al-Ta system at 1330 °C [1991Boe]

Ta having considerable solubility of Al. The σ phase, besides having a large homogeneous region in the binary Ta-Al system, also dissolves Ti (Ti-aluminide) by as much as 60 at.%. The Ti-aluminides TiAl and Ti_3Al dissolve Ta (Ta-aluminide) up to 10 and 15 at.%, respectively. The Fig. 9 shows the presence of the three-phase equilibria, namely, $\sigma + (Ti,Ta)Al + (Ti,Ta)Al_3$, $\sigma + (Ti,Ta)_3Al + (Ti,Ta)Al$, $\sigma + \beta(Ti,Ta) + Ti_3Al$, and $\alpha(Ti) + \beta(Ti,Ta) + Ti_3Al$. Ta being a β stabilizer suppresses the $\alpha(Ti)$ field. As a result, the

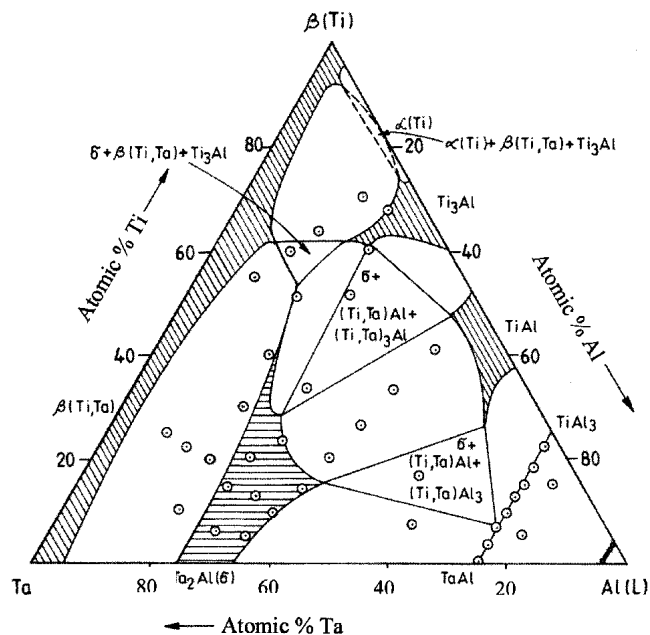


Fig. 9 Isothermal section of the Ti-Al-Ta system at 1100 °C [1983Sri]

three-phase field, $Ti_3Al(Ta) + \alpha - Ti_{ss} + (\beta Ti, Ta)$, shown by the dotted line on Fig. 9, is also very narrow. At this temperature, Al is in the liquid state and in two-phase equilibrium with the $(Ti, Ta)Al_3$ solid solution.

[1991Das1], [1991Das2], [1991Das3], and [1993Das] used diffusion couples as well as bulk, homogenized alloys for the phase equilibria study. Diffusion couples were prepared by hot isostatic pressing of Ta powder on $\gamma TiAl$ alloy at 207 MPa and 1200 °C. The authors also prepared diffusion couples by sputter-deposition of different compositions of the Ti-Ta system (30-150 μm thick) on $\gamma TiAl$ substrate. The diffusion couples were coated with yttria, wrapped in Ta foil, and subsequently annealed at 1100 °C for various lengths of time up to a maximum of 100 h in an argon-filled quartz tube. The annealing process was followed by quenching in water. For a detailed study of the ternary phase obtained from the diffusion couple experiments, an alloy with a similar composition was prepared by an electric arc-melting technique. This alloy was initially homogenized at 1300 °C for 36 h and then at 1100 °C for 120 h followed by furnace cooling. Samples for TEM were obtained from this homogenized alloy.

The 1100 °C isothermal section of the Ti-Al-Ta system based on the experimental study of the bulk alloy samples and diffusion couples is shown in Fig. 10 [1993Das]. Figure 10 shows extensive solubilities of the β , B2, α_2 , γ , σ , and η phases in the ternary phase fields. The other phases, namely, α , δ , and ϕ (TaAl), that are present at this temperature exhibit less extensive solubilities than the aforementioned phases. The tie-triangles in Fig. 10 indicate the following three-phase equilibrium: $\alpha_2 + B2 + \gamma$; $\eta + \gamma + \phi$; $\gamma + \sigma + \phi$; $\sigma + \gamma + \phi$; and $\sigma + B2 + \gamma$. The $\phi + \gamma$ two-phase field lies between the three-phase fields of $\eta + \gamma + \phi$ and $\gamma + \sigma + \phi$. Similarly, the $\sigma + \phi$ two-phase field was deter-

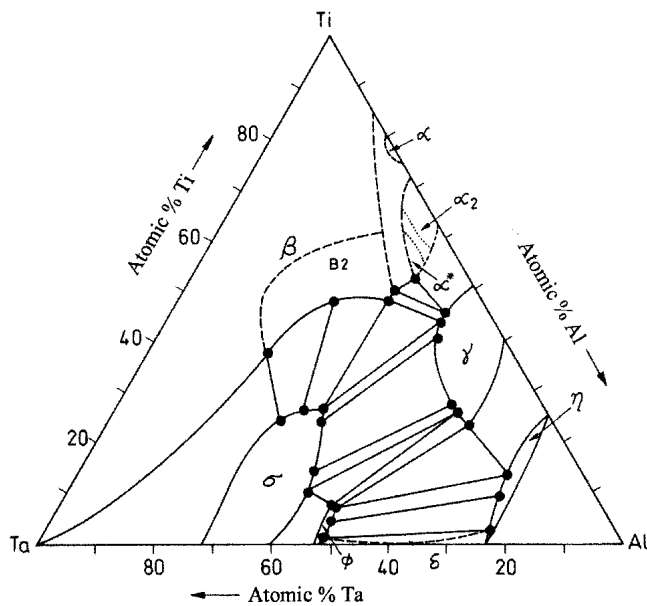


Fig. 10 Isothermal section of the Ti-Al-Ta system at 1100 °C [1993Das]

mined by the boundaries of other phase fields rather than the tie lines established for the samples within that region. This $\sigma + \phi$ phase field is bounded by the Al-Ta binary system and the $\phi + \gamma + \sigma$ three-phase field. The B2 + γ two-phase field lies between the $\sigma + \gamma + B2$ and the $\gamma + B2 + \alpha_2$ three-phase fields. Most of the information available for the $\gamma + \alpha_2$, $\beta(B2) + \alpha_2$, $\beta + \alpha_2 + \alpha$, $\alpha_2 + \alpha$, or $\beta + \alpha$ phase fields were obtained from the bulk diffusion couples. The results show that the α_2 phase extends significantly into the ternary system, whereas the α phase has limited solubility into the ternary phase.

A phase equilibria study [1993Das] has indicated the presence of a ternary phase (B2) near the composition Ti-25Al-25Ta, as shown in Fig. 10. It is seen from Fig. 10 that the B2 phase extends at least to the composition Ti-30Al-20Ta. The occurrence of an order/disorder transformation near 1200 °C has been reported [2003Das]. From the TEM analysis, it is clear that the ternary phase is ordered bcc, B2.

The investigation of the phase boundaries for the $\alpha + \gamma$ and $\alpha_2 + \gamma$ two-phase fields has led to the identification of an island-like region, which was tentatively identified as α^* in Fig. 10 of [1993Das], and has the approximate composition of Ti_4Al_3Ta [1993Jew]. When individual experiments are concerned, there is no conclusive evidence for the existence of the α^* phase. However, when taken as a whole, the information is in agreement with the formation of a new single-phase region. The α^* phase appeared not only in the quenched bulk samples but also in some of the diffusion couples.

[1992McC] used Al (99.99%), Ti (200 ppm O), and Ta (99.99%) as starting materials for preparing the alloys. The required alloys were made in two stages to minimize aluminum evaporation. In the first stage, two master alloys, a Ti-Ta alloy and a Ti-Al alloy, were prepared. In the second stage, the master alloys were crushed, mixed, and melted together to produce the final Ti-Al-Ta alloy of the required

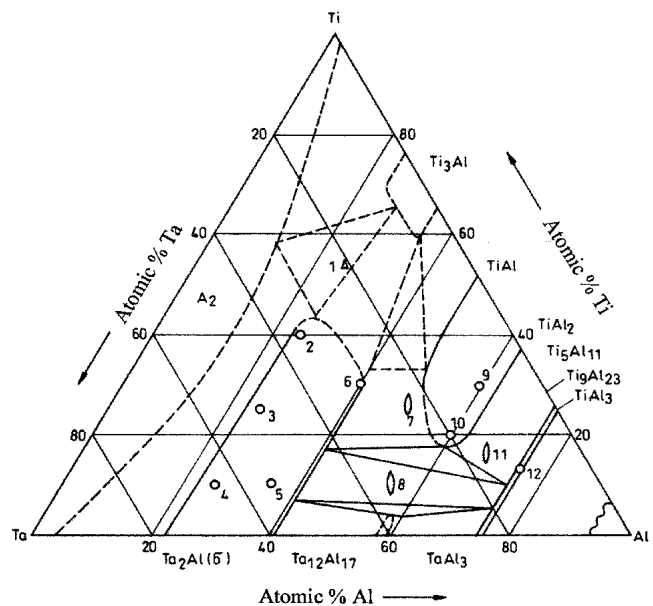


Fig. 11 Isothermal section of the Ti-Al-Ta system at 1000 °C [1966Ram]

composition. The samples were subjected to two types of heat treatments. One consisted of heat treatment for 4 h at 1200 °C for the purpose of relieving any supersaturation in the cast sample, and the other one, which was for equilibrating the samples, consisted of a three-step ramp heat treatment of 1400 °C for 24 h, 1300 °C for 24 h, and 1100 °C for 48 h. For alloys with >60 at.% Al, the 1400 °C step has not been carried out. The techniques used for the characterization of the samples were SEM, energy dispersive x-ray spectroscopy (EDS), XRD, and TEM. The three-step heat treatment was performed to promote interdiffusion between the phases, which would subsequently enable equilibration of the samples at 1100 °C. Most of the microstructures obtained after prolonged heat treatments are in general agreement with the published isothermal section at 1100 °C, even when full homogenization is not achieved.

In both Fig. 9 and 10, Al-Ti and Al-Ta sides are inconsistent with the newest Ta-Al [1996Du] and Ti-Al sides [1992Kat, 2000Ohn, 2002Pal]. For example, the Al_3Ta , Al_3Ta_2 , and Al_7Ta_5 phases are not shown in Fig. 9. Similarly, the Al_3Ta_2 and Al_7Ta_5 phases are missing from Fig. 10. The Al_2Ti and Al_5Ti_2 phases are missing from both Fig. 9 and 10.

6.6 At 1000 °C

[1966Ram] used Al (99.999%), Ti (99.999%), and Ta as the starting materials. The purity of Ta was not stated. The alloys were prepared in an arc furnace under a commercial-purity argon atmosphere. The cast alloys were homogenized by annealing at 1000 °C. The isothermal section at 1000 °C, which was evaluated by [1966Ram], is shown in Fig. 11. No ternary phase was observed in this system. The σ phase was found to form in the homogenized alloys. The Ti and Ta atoms are of almost equal size and can replace each other in a phase. This results in the σ phase taking up about 40 at.%

Ti. The TiAl (CuAu-type) phase also has a large volume occupied by Ta atoms, which replace the Ti atoms. The mutual replacement of Ti and Ta atoms is also evident in the alloy $Ti_{12.5}Ta_{12.5}Al_{75}$ in both the cast and annealed forms. After homogenizing, this alloy consists of a $TiAl_3$ -type phase. The region of the $TiAl_3$ phase is shown in Fig. 11 as a continuous mixed-crystal series. The lattice constants of these phases remain constant. Other important phases in this system are $Ta(Ti)_{12}Al_{17}$, $(Ti,Ta)Al_3$, $(Ti,Ta)Al$, $(Ti,Ta)_3Al$, and $Ta(Ti,Al)$. Figure 11 is not consistent with the recent binary Ti-Al and Al-Ta systems.

[2000Kai] discussed the effect of alloying elements on the shapes of the $\alpha_2/\alpha_2+\gamma$ and $\alpha_2+\gamma/\gamma$ boundaries in the Ti-Al-X (X = V, Cr, Mn, Fe, Co, Ni, Cu, Zr, Nb, Mo, Ta, and W) system at 1000 °C. It was found that an addition of Ta shifts the $\alpha_2+\gamma/\gamma$ boundary toward the Ta-Al side. The authors showed the $\alpha/\alpha+\gamma$ and $\alpha+\gamma/\gamma$ phase boundaries as well as the $\beta/\beta+\alpha$ and $\beta+\alpha/\alpha$ phase boundaries at 1200 and 1300 °C.

7. Comparison between 1440 and 1100 °C Isothermal Sections

The most striking differences between the 1440 and 1100 °C isothermal sections are the changes in the solubilities of the α , β , and γ phases. The solubility of the α phase decreases drastically from about 30 at.% Ta at 1440 °C to about 10 at.% Ta at 1100 °C. The region of stability for the β phase undergoes a considerable shift away from the Al corner of the system when the temperature decreases from 1440 to 1100 °C. The general shape of the β phase field remains the same at both temperatures. The γ phase field undergoes an expansion in width with the decrease in temperature. The ϕ phase is absent in the 1440 °C isothermal section. Quenching and DTA studies performed on the alloys in this region reveal the decrease in homogeneity range of the ϕ phase as the temperature increases from 1100 °C, and finally it disappears via a peritectoid reaction. The δ phase in the Ta-Al system has almost no solubility of Ti in the ternary system at 1100 °C, but steadily increases in Ti content to extend into the ternary system with the rise in temperature.

8. Conclusions

The solidification behavior of the Ti-Al-Ta system has been studied by several groups. The results have been used by them to develop the partial liquidus surface of the system and the isothermal sections of the Ti-Al-Ta system at several temperatures in the range of 1450 to 1000 °C. The most detailed study has been carried out at 1100 °C.

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