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A. Denquin, S. Naka. Various transformation modes observed in two-phase $\gamma+\alpha_2$ TiAl-based alloys. Journal de Physique IV Proceedings, 1993, 03 (C7), pp.C7-383-C7-388. 10.1051/jp4:1993760 . jpa-00252179

HAL Id: jpa-00252179

<https://hal.science/jpa-00252179>

Submitted on 1 Jan 1993

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Various transformation modes observed in two-phase $\gamma + \alpha_2$ TiAl-based alloys

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Abstract

This paper deals with various transformation modes taking place in two-phase $\gamma + \alpha_2$ TiAl-based alloys. These transformation modes are: (1) decomposition of the α phase leading to the precipitation of the γ lamellae ($\alpha \rightarrow \alpha + \gamma$), (2) ordering reaction of the α phase ($\alpha \rightarrow \alpha_2$), (3) massive transformation ($\alpha \rightarrow \gamma$), (4) formation of monolithic γ grains and (5) discontinuous coarsening. By analyzing their kinetics, three types of competitions were identified between these transformation modes: (1) vs. (2), (2) vs. (3) and (4) vs. (5). The occurrence of (2) or (3) and (4) or (5) in binary alloys is strongly dependent on the alloy composition. Therefore, by examining these two competitions in transformation modes, it is possible to evaluate the influence of small additions of ternary and quaternary elements on the response to heat treatments without a precise knowledge of the phase diagrams of complex alloy systems. Another important finding of the present study is related to the formation of the γ phase from the α phase; this can occur through either lamellar precipitation ($\alpha \rightarrow \alpha_2 \rightarrow \alpha_2 + \gamma$, or $\alpha \rightarrow \alpha + \gamma \rightarrow \alpha_2 + \gamma$) or massive reaction ($\alpha \rightarrow \gamma$). The mechanisms involved during these two transformations are investigated and some of the key issues are discussed in the present paper.

Introduction

Titanium aluminide alloys based on TiAl are now considered to be increasingly attractive materials because of their low density and attractive properties at high temperatures, such as high mechanical strength and good oxidation resistance. In particular, the two-phase $\gamma + \alpha_2$ TiAl-based alloys are recognized to exhibit a better ductility and fracture toughness than single-phase γ alloys. However, the mechanical properties of these two-phase alloys appear to be strongly dependent on the microstructure [1]. In order to carefully evaluate the relationship between microstructure and mechanical properties, a precise control of the microstructure is required. The purpose of this study is therefore to provide a better insight into various transformation modes observed during heat treatments in the two-phase $\gamma + \alpha_2$ TiAl-based alloys.

Experimental procedures

The present study was carried out essentially on two binary two-phase TiAl-based alloys with atomic concentrations of $\text{Ti}_{54}\text{Al}_{46}$ and $\text{Ti}_{52}\text{Al}_{48}$. The response to heat treatments was also examined on ternary and quaternary alloys. Their atomic concentrations are as follows: $\text{Ti}_{51}\text{Al}_{47}\text{Cr}_2$, $\text{Ti}_{48}\text{Al}_{48}\text{Cr}_2\text{Nb}_2$ and $\text{Ti}_{50}\text{Al}_{46}\text{Cr}_2\text{Nb}_2$. All the alloys studied were prepared using the vacuum arc melting technique. An argon furnace was used for heat treatments below 1250°C, while a vacuum furnace which allowed the oil quenching option was employed for heat treatments at higher temperatures. Microstructural observations were performed both by optical microscopy and transmission electron microscopy (TEM) on a JEOL 200CX. Since the details of the ternary and quaternary phase diagrams are not available, and because of the small addition of ternary and quaternary elements, the

binary phase diagramme proposed by McCullough et al. [2] was used as reference for understanding phase transformations in all the alloys investigated. In the present work, heat treatments were performed in three phase fields of this phase diagramme: solution heat treatments in the α single-phase field and heat treatments in the $\alpha+\gamma$ and $\alpha_2+\gamma$ phase fields, either after a solution heat treatment or directly from the as-cast state.

Results

1. As-cast conditions

In the as-cast conditions, the alloys showed the lamellar grained structure (Fig. 1a), which is characteristic of two-phase TiAl-based alloys. Some monolithic γ regions, resulting from the interdendritic segregation of aluminium during solidification, can also be observed (for example on the 48 at.% Al alloy in the case of binary alloys). The lamellar structure is generally characterized by a set of several γ lamellae, having the $L1_0$ structure, alternating with a single α_2 lamella, having the DO_{19} structure. The two phases α_2 and γ possess the orientation relationship : $\langle 11\bar{2}0 \rangle_{\alpha_2} // \langle 1\bar{1}0 \rangle_{\gamma}$ and $(0001)_{\alpha_2} // \{111\}_{\gamma}$. The habit plane is the (0001) plane of the α_2 phase [3]. Crystallographically, all the α_2 lamellae in a given grain are oriented in the same manner, due to the fact that α_2 corresponds to a simply ordered form of the initial α phase before decomposition. By examining in more detail the γ lamellae, it is noticed that each γ lamella is divided into numerous order related domains (Fig. 1b and c). These domains are generated due to the fact that the positions of Ti and Al atoms on the $\langle 1\bar{1}0 \rangle$ and $\langle 01\bar{1} \rangle$ axis of the $L1_0$ structure are not equivalent. It is noteworthy that all the domains of a given lamella have the same stacking sequence of $\{111\}$ planes, i.e. ABCABC.. or ACBACB..., while two adjacent lamellae generally have opposite stacking sequences of the $\{111\}$ planes, i.e. ABCABC.. vs ACBACB... The domains of the two adjacent lamellae are then twin or pseudo-twin related [4].

Figure 1: (a) Optical and (b) TEM observation of the lamellar structure, (c) Schematic representation of (b)

2. Formation sequence of the lamellar microstructure

According to the Ti-Al phase diagramme, the lamellar microstructure is the result of the decomposition of the disordered hexagonal α phase into the two phases α_2 and γ . Heat treatments including solution heat treatments in the α phase field have been performed in order to examine the formation sequence of this lamellar structure.

Optical and TEM observations of the binary 46 at.% Al alloy directly quenched from the α phase field (Fig. 2a and b) show a microstructure constituted by large single-phased α_2 grains, with some lamellae formed preferentially around grain boundaries. As shown on figure 2b, the γ lamellae always cut through the antiphase boundaries (APB's) of the α_2 matrix. These results lead to the conclusion that the nucleation of the γ lamellae starts subsequently to the $\alpha \rightarrow \alpha_2$ ordering reaction.

However, if a sample of the same alloy is cooled from the α phase field down to 1230°C (which corresponds to the middle of the $\alpha+\gamma$ phase field) and held for 72 hours before quenching, the microstructure is entirely constituted by large lamellar grains (Fig. 2c and d). This lamellar microstructure is composed by well developed γ and α_2 lamellae. Moreover, the α_2 lamellae contain small APB's. These observations indicate that the formation of γ lamellae can also precede the ordering of the α phase.

The decomposition of the α phase leading to the lamellar structure can occur through two routes: (1) $\alpha \rightarrow \alpha_2 \rightarrow \alpha_2+\gamma$, and (2) $\alpha \rightarrow \alpha+\gamma \rightarrow \alpha_2+\gamma$. This result indicates that during a continuous cooling of the alloy from the single-phase α field, the precipitation of γ lamellae probably starts in the $\alpha+\gamma$ phase field and finishes in the $\alpha_2+\gamma$ phase field.

Figure 2: Optical and TEM observations of the Ti₅₄Al₄₆ alloy, (a) and (b) quenched from the α single-phase field and (c) and (d) cooled down from the α phase field and held for 72 hours at 1230°C before quenching

3. Massive formation of the γ phase ($\alpha \rightarrow \gamma$)

The binary 48 at.% Al alloy quenched from the α phase field shows an optical microstructure quite different from the previous 46 at.% Al alloy (Fig. 3a), although the heat treatment is the same. All the grains are extensively occupied by "massively transformed" regions. A TEM study reveals that these

regions are of the γ phase. Numerous order, twin and pseudo-twin related domains, like those observed in the lamellar γ phase are present in the massive γ phase. In addition, thermal γ APB's are frequently observed (Fig. 3b). When examining the above domains, it is found that they constitute colonies, which are not related to each other by any simple orientation relationship. Some lamellar regions, resulting from the decomposition of the retained α phase, are also observed. A crystallographic analysis of the two regions (massively transformed and lamellar) indicates that the parent α phase and the various colonies of the γ phase do not have any simple orientation relationship. It is also noteworthy that the boundaries between two colonies as well as between the lamellar regions and the massively transformed regions do not seem to follow any particular orientation.

Figure 3: (a) Optical and (b) TEM observations of the $\text{Ti}_{52}\text{Al}_{48}$ alloy quenched from the α phase field.

4. Coarsening of the lamellar structure

It is important to note that the mechanical characterization of the two-phase TiAl-based alloys is generally performed on samples heat-treated in the $\alpha+\gamma$ and/or $\alpha_2+\gamma$ phase fields without any solution heat treatment in the single-phase α field. In order to get a better understanding of the microstructures mechanically tested, various microstructural modifications of the lamellar structure were examined essentially starting from the as-cast state.

The binary 48 at.% Al alloy held at 1150°C for one hour decomposes into numerous monolithic γ grains (Fig. 4a). These monolithic γ grains are located preferentially near a grain boundary or at interdendritic regions enriched in aluminium. On the contrary, the binary 46 at.% Al alloy held at the same temperature is essentially composed of coarser and less regular secondary lamellae (Fig. 4b), whose formation originated from grain boundaries without any significant formation of monolithic γ grains. A part of the microstructure consists of primary lamellae, practically unchanged morphologically. This second coarsening process is called discontinuous coarsening and accompanied by the migration of the grain boundaries. This migration involves a dissolution of the precipitates (γ) of the neighbouring grain [5]. It should be mentioned here that the discontinuous coarsening process can be also observed in the binary 46 at.% Al sample quenched from the α single-phase field and held in the $\alpha_2+\gamma$ phase field (e.g. at 1000°C). After an immediate formation of the primary lamellar structure, new lamellar regions appear around grain boundaries and spread into the primary lamellar grains [6, 7].

Figure 4 : (a) $\text{Ti}_{52}\text{Al}_{48}$ held 1 hour at 1150°C, (b) $\text{Ti}_{54}\text{Al}_{46}$ held 72 hours at 1150°C.

5. Classification of various two-phase TiAl-based alloys

An examination of the kinetics of all the transformation modes observed in the two-phase binary TiAl-based alloys leads to the identification of three competition mechanisms [7]:

- competition between the ordering reaction of the α phase ($\alpha \rightarrow \alpha_2$) and the formation of γ lamellae ($\alpha \rightarrow \alpha + \gamma$),
- competition between the $\alpha \rightarrow \alpha_2$ ordering reaction and the massive formation of the γ phase ($\alpha \rightarrow \gamma$),
- competition between the formation of monolithic γ grains and the discontinuous coarsening.

The first competition indicates that the formation of the γ lamellae can occur both through the disordered α phase and the ordered α_2 phase.

The last two competitions are strongly dependent on the alloy composition. Similar heat treatments performed on the ternary and quaternary alloys indicate that these alloys can be classified into various categories by performing only two heat treatments: a quench from the α phase-field and a heat treatment at 1150°C. Both the nature of the microstructural modifications and the response to heat treatments can then be evaluated without knowing the exact phase diagram corresponding to these alloys. The figure 5 summarizes the classification.

Figure 5 : Classification of various two-phase $\gamma + \alpha_2$ TiAl based alloys

Discussion

Among the various microstructures observed, the lamellar microstructure is the most important ingredient of two-phase $\gamma + \alpha_2$ microstructures, playing a key role in the mechanical behaviour of this class of alloys. The formation mode of the γ phase from the supersaturated α matrix can be either lamellar, or massive, depending on the rate of cooling from the single-phase α field. In the following, we briefly discuss the mechanisms involved in these two transformation modes by using the results of TEM investigations on the nature of numerous γ order-domains introduced during both these transformations. More detailed analyses of the mechanisms are presently at hand.

1. Formation of the lamellar structure

As suggested by Blackburn [3], the nucleation of the γ phase probably takes place on the stacking faults of the supersaturated α or α_2 phase, which are produced by the movement of Shockley partial dislocations on the basal planes of the hexagonal phase. In this case, there are two possible stacking sequences of the $\{111\}$ planes of an FCC-type structure, i.e. ABCABC... or ACBACB... . Those two stacking sequences provide the origin of the formation of twin or pseudo-twin boundaries observed between two adjacent lamellae. The formation of the L1_0 structure during the ordering reaction takes place at a number of sites in a given stacking fault and gives rise to the numerous ordered γ domains. The ordering can occur in a different manner in these domains. The encounter of two growing domains can result in the formation of order-domain boundaries or APB's. However, only few γ APB's have ever been observed in the lamellar structure [8]. A possibly high growth rate of the antiphase domains may explain the lack of APB's. The other boundaries present in the lamellar structure may act as "sink" sites for these

APB'S. Indeed, order-domain, twin and pseudo-twin boundaries can also be coupled with a further translation corresponding to an antiphase vector ($a/2\langle 01\bar{1} \rangle$) [8]. All these boundaries may be called "mixed" boundaries. However, these "mixed" boundaries cannot be differentiated from the others by usual TEM contrast analyses. Work is now underway in order to verify their existence by means of the high resolution transmission electron microscopy.

The lateral growth of the nascent γ phase is probably achieved simultaneously both by shearing (ledge movement [9]) and ordering, since no disordered stage has ever been observed, even during the early stages of the transformation. This growth mechanism leads to a very good atomic matching between the two phases, the habit plane of the two phases lying on the (0001) basal plane of the hexagonal α or α_2 phase.

2. Massive formation of the γ phase

Although the exact mechanisms of both nucleation and growth of the massive formation of the γ phase is still under debate, the fact that the γ phase and the parent α phase do not show any simple orientation relationship, and in particular not the one observed in the lamellar structure, indicates that the nucleation process does not involve the propagation of Shockley partial dislocations or even a shearing of particular atomic planes. Therefore, the transformation seems not to be a martensitic-type but rather a massive-type.

The growth process probably occurs via the movement of the phase interface by thermally activated jumps of atoms because the boundaries between domains or colonies do not follow any specific plane, as already mentioned. The fact that thermal APB's together with order, twin and pseudo-twin related domains can be found in each colony suggests the occurrence of a metastable disordered stage of the γ phase during the transformation [8].

Conclusion

Various transformation modes have been observed in the two-phase TiAl-based alloys. Apart from the lamellar structure, which is observed for all the compositions after a continuous cooling from the α phase-field, the other transformations, i.e. ordering reaction ($\alpha \rightarrow \alpha_2$), massive transformation ($\alpha \rightarrow \gamma$), discontinuous coarsening and formation of monolithic γ grains, appear to be strongly dependant on the alloy composition. The formation mode of the γ phase from the high temperature α phase is either lamellar or massive, depending on the cooling rate from the α single-phase field. These two transformations induce the formation of numerous domains (order, twin and pseudo-twin related). However, a detailed study of the microstructures indicates that the mechanisms involved during the two transformation modes are quite different.

Acknowledgements - This work was supported by a financial support of SNECMA and TURBOMECA. The authors would to thank A. Bachelier-Locq for her technical assistance and P. Thevenin for the alloy preparations. The authors are grateful to Drs P. Veyssiere and G. Vandershaeve for their help in understanding phase-transformations.

References

1. Y.W. KIM, Acta Met., **40**, 1121 (1992).
2. C Mc CULLOUGH, J.J. VALENCIA and R. MEHRABIAN, Acta Met., **37**, 1321 (1989).
3. M.J. BLACKBURN, The Science, Technology and Applications of Titanium, Ed. R.I. Jaffee and N.E. Promisel, Pergamon Press, Oxford, 633 (1970).
4. Y.S. YANG and S.K. WU, Scripta Met., **24**, 1801 (1990).
5. R.D. DOHERTY, Physical Metallurgy, Third Edition, Elsevier Science Publisher, BV, (1983).
6. Y. YAMABE, N. HONJO and M. KIKUCHI, Proc. of Int. Symp. on Intermetallic Compounds, Structure and Mechanical Properties, (JIMIS 6), Ed. O. Izumi, The Japan Inst. of Metals, 821 (1991).
7. A. DENQUIN, S. NAKA and T. KAHN, Proc. of the Seventh World Conf. on Titanium, San Diego (CA) (1992), to be published.
8. A. DENQUIN and S. NAKA, Phil. Mag. Letters, **67**, (1993), to be published.
9. S.R. SINGH and J.M. HOWE, Phil. Mag. A, **66**, 739 (1992).