Microstructure and Notch Properties of Heat-Treated Ti-4.5Al-3V-2Mo-2Fe Laser Welds

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The microstructure and transformation behavior of Ti-4.5Al-3V-2Mo-2Fe alloy and its laser welds after various heat treatments were investigated. Notch properties, such as impact toughness and notched tensile strength, were also measured on the welds to choose an appropriate post-weld heat treatment (PWHT) of the alloy. The temperature regimes in which β transformed into α or remained as β after quenching were identified and discussed. The results indicated that α could be obtained primarily by rapid quenching from 880–840 °C solution temperatures and was identified as a base-centered orthorhombic (centered on the C face) with the lattice parameters of a = 0.305 nm, b = 0.489 nm, and c = 0.457 nm. The as-welded specimen exhibited fine acicular α in the β matrix with hardness considerably higher than the mill-annealed base metal. For a PWHT temperature lower than 800 °C, the change in microstructure and hardness of the welds depended mainly on the temperature, not on the cooling rate. If the PWHT temperature was higher than 800 °C, both the temperature and the cooling rate were important in altering microstructure and hardness of the welds. The welds after a 704 °C/4 h treatment could prevent notch brittleness, reduce hardness variation in different regions of the weld, and obtain notch properties similar to the mill-annealed base metal. [doi:10.2320/matertrans.MRA2008370]

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1. Introduction

The Ti-4.5Al-3V-2Mo-2Fe alloy is a β-rich α + β titanium alloy, which is also known as SP-700 from its excellent superplasticity at 700 °C. It was designed to improve superplasticity and lower the operating temperature to reduce production cost, as compared with traditional Ti-6Al-4V alloy. Other properties, such as heat-treatability, cold bendability and fatigue properties of Ti-4.5Al-3V-2Mo-2Fe are superior to those of the Ti-6Al-4V alloy. After solution heat treatment, this alloy demonstrates a fast response to age-hardening. For instance, the holding time required to attain the peak hardness at 500 °C is less than two hours, which is considerably shorter than that of most heat-treatable titanium alloys. The Ti-4.5Al-3V-2Mo-2Fe alloy has higher β-stabilizing element content, which modifies the transformation of the β phase and suppresses the diffusional transformation during cooling and quench delay. Among the β-stabilizing elements in the alloy, Fe is an effective strengthening agent and Mo refines the microstructure and retards grain growth during superplastic forming and heat treatment. Other alloying elements, such as Al and V, are used to adjust the β-transus temperature (Tₜ) which can have a significant influence on the superplastic temperature of Ti-4.5Al-3V-2Mo-2Fe alloy. It is well known that thermal history strongly affects the microstructure, as well as the mechanical properties, of α + β titanium alloys. For instance, a coarsening of the acicular structure increases the fracture toughness but decreases the strength of α + β titanium alloys. The Ti-4.5Al-3V-2Mo-2Fe alloy exhibits a wide variety of microstructures depending on heat treatment conditions, however, less attention has been paid to the weld metal microstructure and associated properties of post-weld heat-treated welds. In general, PWHTs in α + β titanium alloys are performed not only for stress relief, but also for improved ductility and toughness of the weld metal. It has been suggested that the stress-relieving temperature range is 538–593 °C for Ti-6Al-4V welds, and 760 °C for Ti-4.5Al-3V-2Mo-2Fe laser welds. The reason for choosing a higher post-weld heat treatment (PWHT) temperature (760 °C/1 h) of the latter was to avoid notch brittleness of the weld metal. However, it should be noted that the PWHT temperature of 760 °C is too high for titanium alloys and may require fixture to avoid causing excessive distortion at high PWHT temperatures. In fact, PWHT involves temperature and time factors, i.e., temperature/time combinations. Accordingly, performing PWHT of Ti-4.5Al-3V-2Mo-2Fe laser welds at a lowered temperature with prolonging time also makes it possible to reduce notch brittleness and requires further study.

Ti-4.5Al-3V-2Mo-2Fe laser welds in the as-welded condition consist of fine acicular α in the β matrix and require a high PWHT temperature to reduce notch sensitivity. As a result, a suitable PWHT of Ti-4.5Al-3V-2Mo-2Fe welds is crucial and of current interest. The present study deals with the microstructural evolution of the heat-treated Ti-4.5Al-3V-2Mo-2Fe alloy and its laser welds. The martensitic transformation of the alloy is also discussed in detail, in particular, the crystal structure of orthorhombic martensite. In addition, the influence of PWHTs on the impact toughness and NTS of the laser-welded specimens is assessed and an appropriate PWHT for Ti-4.5Al-3V-2Mo-2Fe welds is suggested.

2. Material and Experimental Procedures

The materials used in the investigation were 3 mm thick sheets of mill-annealed Ti-4.5Al-3V-2Mo-2Fe, with the chemical composition in weight percent of 4.74% Al, 2.26% V, 1.89% Mo and 1.84% Fe. Prior to laser welding,
the mill-annealed sheets (200 mm × 30 mm × 3 mm) were mechanically ground with 240 grit SiC papers to remove surface contamination and cleaned with acetone. A Rofin-Sinar 5 kW CO2 laser was utilized to manufacture bead-on-plate welds with the welding direction perpendicular to the rolling direction of the sheets. Laser welding parameters including a laser power of 3 kW and a travel speed of 1000 mm/min were used to achieve full-penetration welds. After laser welding, small specimens (10 mm × 30 mm × 3 mm) were cut from the welded sheets and then heat-treated in an argon furnace. Such specimens were heat-treated either at temperatures below Tp (~900°C) for 1 h or at 920°C for 0.5 h, followed by water quenching as well as air cooling. These specimens were used to study the phase transformation and microstructure of Ti-4.5Al-3V-2Mo-2Fe welds under various heat treatment conditions. Since microstructural examinations were made at the middle-thickness region of all specimens, a thin oxide layer (~100 μm) formed after heat treating did not affect either the interior microstructure or the transformation behavior.

Charpy impact and notched tensile tests were performed to assess notch properties of the welds after PWHTs. Impact specimens (sub-size, 2.5 mm thick) according to ASTM E23-07a standard were employed to conduct impact tests of laser welds at room temperature. Double-edge notched specimens (2.8 mm thick), which have been discussed elsewhere, were used to determine the NTSS of the welds after distinct PWHTs. Notched tensile tests were carried out in air at a displacement rate of 1 mm/min. Impact and notched tensile specimens were evaluated to find out a suitable PWHT for Ti-4.5Al-3V-2Mo-2Fe laser welds. The weld criterion was essentially based on the mechanical properties, i.e., the weld after a suitable PWHT should have impact toughness and NTSS comparable to its base metal. In order to simply the description of a heat-treated specimen, the identification numbers were assigned according to the region of welded specimens (BM: the base metal; WM: the weld metal), heat treatment temperature (temperature in °C), and cooling medium (Q: water quenched; A: air cooled) combinations. For instance, the BM-760A specimen represents the base metal which was heat-treated at 760°C/1 h and then air-cooled; the WM-920Q specimen stands for the weld metal which was heat-treated at 920°C/0.5 h followed by water-quenching.

The specimens for optical metallography and scanning electron microscope (SEM) were polished and etched in a solution composed of 10% nitric acid, 5% hydrofluoric and 85% water. Microhardness measurements of various specimens were taken by a Vickers microhardness tester. For detailed microstructural observations, transmission electron microscopy (TEM) specimens were sectioned in slices within the WM and BM regions of a given specimen and examined with a 200 kV microscope. Thin foils were prepared by a standard twin-jet polisher using an electrolyte consisting of 60% methanol, 34% butanol and 6% perchloric acid at −40°C.

3. Results and Discussion

3.1 Laser-welded Ti-4.5Al-3V-2Mo-2Fe alloy

Figure 1(a) is the cross-sectional view of a typical laser weld in the as-welded condition, in which three distinct regions, i.e., the WM, heat-affected zone (HAZ), and the BM, are obvious. The WM consists of large columnar grains with fine acicular α in the β matrix as shown in Fig. 1(b). Detailed TEM examinations of such a specimen have been reported in a previous study. The HAZ adjacent to both sides of the WM is about 1 mm and contains different grain sizes that can be roughly divided into coarse-grained and fine-grained HAZs as marked in Fig. 1(a). Presumably, the coarse-grained and fine-grained HAZs can be distinguished from their peak temperatures, i.e., higher and lower than Tp in the welding process. In the coarse-grained HAZ, the grain size (10~100 μm) is larger than that of the BM of (3~5 μm) and the microstructure consists of fine acicular α in the β matrix, as displayed in Fig. 1(c). Further away from the fusion boundary, the fine-grained HAZ has similar structures to the mill-annealed BM consisting of elongated α + β as illustrated in Fig. 1(d).

Microhardness measurements indicated that the hardness values of the WM and the BM were 420 and 330 Hv in the as-welded specimen, respectively. The hardness of the HAZ was in between the WM and the BM, with the coarse-grained HAZ (~400 Hv) being harder than the fine-grained HAZ (~350 Hv). The same tendency has been observed for the laser welds subjected to various heat treatments as discussed below. The fine α + β structures in the WM (Fig. 1(b)) and the coarse-grained HAZ (Fig. 1(c)) indicated that the cooling rates in these regions were not fast enough to form martensite in the laser welding process. The HAZ microstructure in this alloy was quite complicated because thermal cycles and peak temperatures varied at different locations in the HAZ zone. In the following sections, the attention is focused on microstructural analyses of the WM and BM specimens subjected to various heat treatment procedures. This is particularly important for the welds to perform PWHT or stress-relieving after welding.

3.2 Microstructure of the water-quenched welds

In the Ti-4.5Al-3V-2Mo-2Fe alloy, the β-phase can undergo transformation to martensite (α' and α") or be retained as β-phase upon water quenching. The phase transformation behavior of the BM and WM specimens is the same. Metallographic observations and phase identifications of distinct specimens were further confirmed by TEM examinations. Optical micrographs of the laser-welded specimens that were solution-heated at the temperature range of 800 to 920°C and then water-quenched are displayed in Fig. 2. The microstructures of the BM- and WM-920Q specimens are entirely α" (hcp martensite) which has a needle-like structure, as shown in Fig. 2(a). The major difference between these two specimens is the grain size, with the latter being significantly larger than the former. Decreasing the solution temperature, for example, to 845°C, the microstructure of the BM-845Q specimen consists of α' (orthorhombic martensite) and αp (primary α, small particles at grain boundaries), as shown in the left half of Fig. 2(b). It is clear that the β and α phases equilibrated at 845°C were transformed to α" and remained as αp upon quenching, respectively. For the solution temperature less than Tp, the amount of αp in the BM specimens depended on
temperature, i.e., the lower the temperature the greater the amount of $\alpha_p$. By lowering the solution temperature in the two-phase region, the $\beta$-phase could be retained upon quenching due to the enriched $\beta$-stabilizing elements in that phase. In the case of the WM-845Q specimen, the microstructure consists of small elongated $\alpha$ in the $\alpha''$ matrix along with grain boundary $\alpha$ particles as shown in the right half of Fig. 2(b). As the solution temperature is decreased, the BM- and WM-800Q specimens contain smaller but higher volume fraction of $\alpha_p$ in the $\alpha''$ matrix relative to the BM- and WM-845Q specimens, as shown in Fig. 2(c).

The crystal structure of the above-mentioned phases such as $\alpha$, $\beta$, and $\alpha'$ are well known in titanium alloys. However, the crystal structure of $\alpha''$ and its lattice parameters in the Ti-4.5Al-3V-2Mo-2Fe alloy have not been fully investigated. As a result, the structure of $\alpha'$ martensite was reported to have either a face-centered or a base-centered structure in titanium alloys. In this study, the crystal structure of $\alpha''$ has been identified as a base-centered (centered on the C face) orthorhombic lattice. Figures 3(a) and 3(b) reveal the microstructure of $\alpha''$ and the associated electron diffraction pattern, respectively. It is important to note that a base-centered orthorhombic can only be distinguished from a face-centered orthorhombic in some selected zone axes, e.g., [110] and [311], of the diffraction patterns according to structure factor calculations. The lattice parameters, which are composition dependent, were calibrated with the aid of coating a thin Pt layer on TEM foils. Figure 3(c) shows the superimposed patterns of $\alpha''$ and polycrystalline Pt, in which the d-spacing of (111)$_{\text{Pt}}$ and (200)$_{\text{Pt}}$ rings can be used for calibration. The lattice parameters of $\alpha''$ were determined to be $a = 0.305\,\text{nm}$, $b = 0.489\,\text{nm}$, and $c = 0.457\,\text{nm}$ were determined.

Figure 4 contains TEM micrographs of the BM and WM specimens which were solution-treated at various temperatures and then water-quenched. These micrographs also confirmed the metallographic observations and the description of the phases in the various specimens as illustrated in Fig. 2. A typical micrograph of both the BM- and WM-920Q specimens is shown in Fig. 4(a), in which the microstructure contains entirely $\alpha'$ with internal twins and stacking faults. In fact, the co-existence of $\alpha'$ and $\alpha''$ could be observed in the specimen which was solution-treated at temperatures from slightly below $T_\beta$ to about $880\,^\circ\text{C}$, and then water-quenched. Quenching from a solution temperature range of $880\sim840\,^\circ\text{C}$, e.g. $845\,^\circ\text{C}$, the transformation of $\alpha''$ to $\alpha'$ in the WM-845Q specimen is similar to that in the BM-845Q specimen, as shown in Fig. 3(a). For the solution temperature lower than $840\,^\circ\text{C}$ but higher than $800\,^\circ\text{C}$, the transformation of $\beta$ to $\alpha''$ in the WM-845Q specimen was not complete and some $\alpha''$ retained upon quenching. The volume fraction of $\alpha''$ decreased and the amount of retained $\beta$ increased as the temperature lowered accordingly. For the weld solution-treated at $800\,^\circ\text{C}$, fine $\alpha' + \beta$ structures in the WM grew substantially at that temperature. Upon water quenching, the resultant microstructures of the WM-800Q specimen consist of only coarsened $\alpha' + \beta$ structures without the presence of $\alpha''$, as shown in Fig. 4(b). A typical micrograph of the BM-800Q specimen is displayed in Fig. 4(c), in which the size and distribution of $\alpha$ in the $\beta$...
matrix are slightly different from those of the WM-800Q specimen. For the welded specimens equilibrated at a temperature less than 800°C, the structures formed at that temperature were not affected by the subsequent cooling rates (discussed in the following section). In summary, five heat-treated temperature regimes can be roughly identified with regard to the quenching products that are observed in the β-phase regions: (1) \( T > 900°C, \beta \rightarrow \alpha' \); (2) \( 900°C > T > 880°C, \beta \rightarrow \alpha' + \alpha'' \); (3) \( 880°C > T > 840°C, \beta \rightarrow \alpha'' \); (4) \( 840°C > T > 800°C, \beta \rightarrow \alpha' + \beta \); (5) \( T < 800°C, \beta \) retained.

### 3.3 Effect of cooling rates

The effects of heat treatment temperatures and cooling rates (air cooling and water quenching) on the hardness in the WM and BM regions of laser welds are displayed in Fig. 5. The results clearly indicate that the cooling rate did not affect the hardness of the WM (Fig. 5(a)) and BM (Fig. 5(b)) regions for heat treatment temperatures less than approximately 800°C. Nevertheless, low temperature heat treatments increased the hardness of the WM considerably, e.g., from 350 Hv of the WM-760Q specimen to 420 Hv of the WM-593Q specimen and 480 Hv of the WM-482Q specimen. The coarsening of \( \alpha' + \beta \) structures as well as the grain boundary \( \alpha \) (Fig. 6) was responsible for the decreased hardness of the WM with increasing PWHT temperature. In contrast, the temperature-hardness dependence of the BM (mill-annealed) specimens was minor and exhibited only a small increase in hardness as the temperature was decreased.

For the heat treatment temperature higher than 800°C, the difference in cooling rates affected the hardness of the WM and BM regions significantly and the air-cooled specimens were apparently harder than the water-quenched specimens. Figure 7(a) shows a typical micrograph of both the WM-920A and BM-920A specimens, consisting of acicular \( \alpha \) in the \( \beta \) matrix and having the same hardness of about 420 Hv.
It is very different from the entirely $\alpha'$ structure in the WM-920Q specimen, as shown in Fig. 4(a). When the solution temperature is lowered, e.g., to 870°C, the resulting microstructures of the WM-870A and BM-870A specimens are shown in Figs. 7(b) and 7(c), respectively. The microstructures of these two specimens consisted of secondary $\alpha$ ($\alpha_{sec}$) in the $\beta$ matrix along with $\alpha_p$ of different morphologies. Unlike martensite in steels, titanium martensite has no strengthening effect in the titanium alloys. As a result, the hardness of the WM-920A specimen (419 Hv) was significantly higher than that of the WM-920Q specimen (322 Hv). The same trend was found for the WM and BM specimens solution-treated at temperatures between $T_\beta$ and 800°C, however, the difference in hardness was reduced as the temperature lowered (Fig. 5). For instance, the WM-870A and WM-870Q specimens had hardness values of 388 and 315 Hv, respectively.

### 3.4 Notch properties of laser welds

In practical applications, a solution treatment (temperature higher than 800°C) of the titanium welds after welding is usually not performed. As a result, the focus of this section is on the PWHT temperature of Ti-4.5Al-3V-2Mo-2Fe laser welds less than 800°C. Notch properties, such as impact toughness and notched tensile strength, of the welds are
important WM properties to be considered in the selection of an appropriate PWHT procedure. To some extent, they indicate the ability of the welds to resist fracture and can be determined easily. The corresponding specimens, such as single V-notched impact and double V-notched tensile specimens, are used to assess notch sensitivity of the welds. It is generally agreed that low notch sensitivity is associated with ductile materials, while high notch sensitivity is associated brittle materials.\(^\text{18)}\) In most cases, the PWHT of a given weld should be considered to avoid large differences in mechanical properties among the WM, HAZ, and BM regions. The impact energy and notch tensile strength of the WM after PWHTs would best be equivalent to those of the BM in a weld. Ti-4.5Al-3V-2Mo-2Fe laser welds in the as-welded condition were very brittle and had an impact energy of approximately 1.2 J. Figure 8 shows the impact energy of the welds heat-treated at various PWHT temperatures for 1 h, in which the impact energy of the mill-annealed BM is also included for comparison. It also indicated that the weld aged at 482°C/1 h, i.e., the WM-482A specimen, exhibited the lowest impact toughness (0.8 J) among the welded specimens. Apparently, PWHTs at higher temperatures led to an obvious increase in impact toughness of the welds. For example, the impact values were 5.9 and 8.8 J for the welds after 704°C/1 h and 760°C/1 h PWHTs, respectively.

The NTS of Ti-4.5Al-3V-2Mo-2Fe laser welds that were heat-treated at various PWHT temperatures for 1 h exhibited the same trend as the impact energy-temperature relationship and the lowest NTS was also associated with the peak-aged (482°C/1 h) weld.\(^\text{12)}\) To minimize the difference in notch properties of different weld regions, the NTS (1100 MPa) and impact toughness (7.2 J) of the mill-annealed BM can be used as criteria for choosing suitable PWHTs of the welds. In other words, the notch properties of the WM should be comparable to those of the BM. Table 1 gives the NTS and impact values of laser welds after some PWHT procedures. Clearly, the welds after 704°C/1 h and 760°C/1 h treatments could not meet the criteria; however, the weld subjected to a 704°C/4 h treatment could improve the notch properties to meet the requirements. The NTS (1148 MPa) and impact energy (9.0 J) of the WM after a 704°C/4 h treatment were slightly higher than those of the mill-annealed BM. Obviously, notch properties of the welds could be improved considerably by a further coarsening of \(\alpha + \beta\) structures with prolonging time to 4 h at 704°C (Fig. 9). Additionally, the HAZ had hardness values between the WM (~352 Hv) and the BM (~340 Hv) regions for the weld after a 704°C/4 h treatment. Accordingly, the strength among various regions of the welds after such a PWHT was expected to be similar. It should also be noted that a PWHT temperature of 760°C is too high for titanium alloys and should not be considered in practice. Based on the results of impact and notch tensile tests as well as hardness measurements, a 704°C/4 h PWHT seemed to be appropriate for stress-relieving Ti-4.5Al-3V-2Mo-2Fe welds.
4. Conclusions

(1) Five heat treatment temperature regimes could be roughly identified with regard to the quenching products that were observed in the β-phase regions of the BM and WM specimens: (1) T > 900°C, β → α’; (2) 900°C > T > 880°C, β → α’ + α”; (3) 880°C > T > 840°C, β → α”; (4) 840°C > T > 800°C, β → α” + β; (5) T < 800°C, β retained.

(2) The transformation to martensite (α’ or α’’) of Ti-4.5Al-3V-2Mo-2Fe could be obtained only by rapid quenching from high solution temperatures. The crystal structure of α” could be identified as a base-centered orthorhombic (centered on the C face) with the lattice parameters of a = 0.305 nm, b = 0.489 nm, and c = 0.457 nm.

(3) The cooling rate did not affect the hardness of the weld and base metals for PWHT temperatures less than 800°C. Nevertheless, the hardness of the WM specimens decreased considerably as the PWHT temperature increased due to the coarsening of the acicular α structure. In contrast, the BM specimens exhibited only a small increase in hardness as the temperature decreased.

(4) For solution temperatures higher than 800°C, the cooling rate affected the hardness of both weld and base metals significantly and the air-cooled specimens were apparently harder than the water-quenched specimens. Apparently, the cooling rate altered transformation products, i.e., microstructures, to cause changes in mechanical properties.

(5) The welds post-weld heat-treated at 704°C/4 h led to a coarsening of α + β structures that could avoid notch brittleness, reduce hardness differences among various regions of the weld, and obtain notch properties similar to the mill-annealed BM. As a result, a PWHT at 704°C/4 h was recommended for stress relief of Ti-4.5Al-3V-2Mo-2Fe welds.

REFERENCES