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Influence of high-energy shot peening on the microstructure and mechanical properties of Ti–6Al–4V welded joints

He Xiaomei 1, Song Guodong and Zhu Xiaoya
School of Metallurgical Engineering, Xi’an University of Architecture & Technology, Xi’an 710055, People’s Republic of China
1 Author to whom any correspondence should be addressed.
E-mail: 85554949@qq.com

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Abstract

In this study, Ti–6Al–4V welded joints were subjected to a high-energy shot peening (HESP) treatment. The structural characteristics of the Ti–6Al–4V welded joints were investigated by optical microscopy (OM) and x-ray diffraction (XRD). The Rockwell hardness and residual stress in different regions of the Ti–6Al–4V welded joints were assessed before and after treatment. Tension tests were then performed to investigate the effect of HESP on the welded samples. Experimental results showed that an Surface nanocrystallization (SNC) layer can be prepared on a Ti–6Al–4V welded joint via HESP. The average grain size of the SNC layer was approximately 30 nm after HESP. HESP also resulted in a higher surface hardness compared with that of the original sample. After treatment, surface compressive residual stresses formed in welded joints, and the residual peak stress of the weld zone was decreased from 650 MPa to −638 MPa. The surface hardness and surface residual stress were distributed more uniformly on the surface of the welded joint. After treatment, the tensile properties were improved.

1. Introduction

Currently, titanium alloys are one of the most important metallic materials used in orthopedics and dental surgery [1]. Ti–6Al–4V is the most commonly used titanium alloy and is widely used in the aerospace field. With the development of aero engines to have high thrust, a high thrust-to-weight ratio, and high reliability, welded structures are being increasingly used in aero-engine structures. However, the inherent heating and cooling during the welding process itself and the process characteristics of adding the welding material determine the inhomogeneous mechanical properties between the base metal and welded areas [2–5]. For welded joints, similar to most materials, material failures such as fretting, fatigue fracture, corrosion and wear occur mainly on component surfaces [6, 7]. Therefore, modifying surface properties of components and parts would be an economical and efficient method for enhancing the overall behavior of materials, thus extending their practical applications [8].

To enhance material surface properties, researchers have examined a wide range of different surface modification techniques. Compared with their coarse-grained counterparts, nanocrystalline materials possess superior properties, including high strength and hardness [9]. Nanostructured materials have received considerable attention among materials scientists due to their unique physical, chemical, and mechanical properties, which can be explored for numerous technological applications [10–17]. Therefore, the nanostructure of the surface layer of a metal material strongly influences its overall performance. Commonly used surface nanocrystallization methods include surface mechanical attrition treatment (SMAT) [9, 18–20], ultrasonic shot peening (USSP) [21, 22], and high-energy shot peening (HESP) [23].

In this study, HESP was used to treat Ti–6Al–4V welded joints. Both of these processes, are cold working surface treatments in which a large number of metallic or ceramic particles impact a treated surface with the aim of
of introducing compressive residual stresses and increasing microhardness in a surface region, thus substantially improving the fatigue life and strength of peened component [24–26]. HESP induces multidirectional plastic deformation and deeper compressive residual stress and creates a superficial nanocrystalline layer. However, HESP is expected to result in thicker nanocrystalline and work-hardened surface layers as well as deeper surface regions with larger residual compressive stresses [27]. The effect of surface nanocrystallization on the microstructure and properties of welded joints was investigated.

2. Experimental

2.1. Material and welding method

2.1.1. Material

The as-received material tested in this study was hot-rolled titanium alloy (Ti–6Al–4V) sheet (1 mm in thickness), which was an equiaxial two-phase alloy. The chemical composition is given in table 1. The microstructure of the hot-rolled sheet is shown in figure 1. It has a fully homogeneous equiaxed microstructure consisting of 70% primary α phase with an average grain size of 10 μm and transformed β with secondary lamellar α thickness of 5 μm.

![Figure 1. Microstructure of the Ti–6Al–4V titanium alloy.](image)

| Element | Al | V | Fe | C | O | N | H | Ti |
|---------|----|---|----|---|---|---|---|----|
| (%)     | 5.5| 3.5| 0.3| 0.1| 0.2| 0.05| 0.015| Balance |

Table 1. Chemical composition of Ti–6Al–4V alloy.

| Welding current | Welding voltage | Shield gas flow (front) | Shield gas flow (back) | Welding wire |
|-----------------|-----------------|-------------------------|------------------------|--------------|
| 60 A            | 12 V            | 5 l min⁻¹               | 10 l min⁻¹             | Ti–6Al–4V    |

Table 2. Welding parameters.

Tungsten inert gas (TIG) welding of the Ti–6Al–4V titanium alloy did not open the groove during welding. The welded joint parts were mechanically cleaned. Specifically, they were wiped with a fine abrasive cloth brush to remove the oxide film from the welded part, and then fingerprints on both sides of the head to be welded were removed by scrubbing; organic matter and oil on the surface of the wire were removed with acetone. The welding parameters are given in table 2.
2.2. Sample processing

2.2.1. High-energy shot peening
HESP treatment was carried out on welded joint specimens. The HESP pressure was 0.4 MPa, with a processing duration of 30 min. Shot peening was carried out by stainless steel ball with a diameter of 0.3 mm, the impact distance was 150 mm, and the shot angle was about 85°. The HESP setup is shown in figure 2.

2.2.2. Microstructure
After HESP, the sample was cut by wire electrical discharge machining and then ultrasonically cleaned in acetone. Microstructure characterization was performed using an OLYMPUS GX51 optical microscope. The phase constitution of different zones in the welded joint and sample surface layer at room temperature was characterized via x-ray diffraction (XRD) analysis using a Shimadzu XRD-7000 x-ray diffractometer equipped with a Cu Kα radiation source operated at a voltage and current of 40 kV and 40 mA, respectively. The scan range was 25.00°–80.00°, with a step size of 0.02°.

2.2.3. Hardness
The Rockwell hardness of the weld, heat-affected zone and parent metal of the Ti–6Al–4V-welded joint before and after nanotexturing were measured using a TH300 Rockwell hardness tester. The load was 150 kg, and a diamond indenter was used. Five readings were taken on each sample and the average of all was considered as final reading.

2.2.4. Tensile tests
The tensile specimens were prepared in accordance with ASTM-E8 standards. The specimens were cut in the vertical welding direction, and the welded joints were in the center of the 1 mm-thick tensile specimen (as shown in figure 3). After the tensile test specimen was prepared, HESP treatment was performed. The tensile test was carried out using an Instron 8801 hydraulic servo fatigue machine at a stretching temperature of room temperature (25 °C) and a speed of 1 mm min⁻¹.

After the tensile test, the fracture of the tensile sample was observed with an S-2700 scanning electron microscope.
2.2.5. Residual stress

Residual stress measurements were performed on the sample. The test parameters for this measurement are shown in Table 3.

3. Results and analysis

3.1. Microstructure of the welded joint

Figure 4 shows the microstructure of the welded joint. The lower surface is the microstructure after HESP, and the upper surface is the original welded pattern. The base-metal structure is composed of small equiaxed crystals with a bidirectional structure of $\alpha + \beta$, where white is $\alpha$ crystals and $\beta$ crystal grains are distributed in a sheet shape between the $\alpha$ grains (A). The weld microstructure is the needle-like $\alpha$ phase + the original $\beta$ phase grain boundary; the martensite transformed from the coarse $\beta$ grain, the $\beta$ grain boundary is clearly visible, and the needle-like distribution direction in the same grain is basically the same (C). During the cooling process, the weld first grows toward the center of the weld near the unmelted solid metal close to the fusion line to form martensite. Due to the faster cooling rate of the weld, the degree of supercooling is large and the martensite is relatively small and acicular. The heat-affected zone is substantially changed by the action of the welding thermal cycle. The near-region is organized into needle-shaped martensite at a distance from the center of the weld. This phenomenon is due to the high temperature in this area, which reaches the phase-transition temperature of the $\alpha + \beta \rightarrow \beta$ transition. However, the temperature of this region is lower than that of the weld, and the cooling rate is higher than that of the weld; thus, the grain in this region is finer than that in the weld zone (B). Figure 4 shows that the surface layer with HESP is deformed. As the distance from the blasting surface increases, the plastic deformation gradually diminishes until the surface of the material is strongly plastically deformed, resulting in severe damage to the surface grains.

3.2. X-ray diffraction analysis

Figure 5 shows the XRD patterns of the base metal and weld areas of the original welded joint. The pattern of the base metal and the weld area contrast with the diffraction pattern of Ti–6Al–4V, and they all coincide with the diffraction peak of Ti–6Al–4V, except that the intensity of the peak differs. This difference in intensity is attributed to the titanium alloy sheet exhibiting preferential orientation of grain growth during the rolling process. Compared with that for the base metal, the process of grain growth and orientation differs. In the welding process, the molten metal near the side of the base metal has a temperature gradient, the solidification speed is faster, and the crystal grains are too late to grow. In the center of the weld, the diffraction pattern is different because the solidification speed is slower, the grains are preferentially grown, and the preferred orientation of the grains is different from that of the base metal.

Figure 6 shows that, after HESP, the peak intensity in the XRD pattern decreases due to the deformation layer formed on the surface layer; in addition, the base metal and the weld bead tend to be uniform. The diffraction patterns before and after the high-energy treatment show that the characteristic peaks of crystal faces (1010), (0002) and (1011) are substantially broadened. The severe plastic deformation induced by HESP distorts the

| Target | Diffraction crystal plane | $2\theta$ | Poisson’s ratio | Elastic modulus | Voltage | Current | Scanning range | Field of view |
|--------|---------------------------|----------|----------------|----------------|---------|---------|----------------|--------------|
| Ti     | (110)                     | 137°     | 0.36           | 120–200 MPa    | 30 kV   | 8 mA    | 130–144°       | φ3 mm        |

Table 3. Residual stress measurement parameters.

Figure 4. Microstructure of Ti–6Al–4V titanium alloy TIG welded joint.
lattice in the strengthening layer and refines the grain, thereby causing a change in the XRD intensity.

Broadening of the Bragg diffraction peaks is a consequence of three effects: grain refinement, an increase in microscopic stress and instrumental broadening. Because the instrumental broadening effect of different samples is constant, the grain size and microscopic stress of the sample after HESP could be judged, revealing a considerable change. At the same time, the change in the full-width at half-maximum (FWHM) reflects the degree of work hardening, the magnitude of microscopic residual stress and the level of dislocation density inside the crystal. The degrees of work hardening, grain refinement, microscopic residual stress and dislocation density increase, which can lead to broadening of the diffraction peak. These effects can also indicate that, after the HESP treatment, the hardness of the surface layer of the material as well as the dislocation density is increased.

The Scherrer formula [28] is used to calculate the grain size and micro strain of the base metal surface and welding area of Ti–6Al–4V welded joint.

The Scherrer formula is expressed as follows:

$$D = \frac{K\lambda}{\beta \cos \theta}$$

where $K$ is the Scherrer constant, which has a value of 0.89; $D$ is the grain size (nm); $\beta$ is the diffraction half-width height (rad); $\theta$ is the diffraction angle; $\lambda$ and is the XRD wavelength, with a value of 0.154 056 nm.

Microscopic strain refers to the elastic deformation caused by compressive stress or tensile stress between grains and grains, between mosaic blocks, or between certain regions. The size and direction are statistically distributed. Within the macro size range, the average is zero.

The calculation results are shown in table 4. The grain size of the base metal and the weld zone were reduced to the nanometer scale after HESP.
3.3. Hardness
The test results of the hardness of the welded joint are shown in Table 5. The hardness of the Ti–6Al–4V welded joint near the center is the highest, and the hardness decreases toward both sides. The hardness of the material after HESP increased from 33.8 to 35.6, indicating that HESP can improve the surface hardness of the material. After HESP of the material, the surface hardness of the welded joint tends to be homogeneous.

Table 4. Average grain size and average microscopic strain of HESP Ti–6Al–4V.

| Sample          | Grain size (nm) | Microscopic strain (%) |
|-----------------|-----------------|------------------------|
| HESP (base metal) | 25.7 ± 3.0      | 0.417 ± 0.044          |
| HESP (weld)     | 35.8 ± 4.8      | 0.475 ± 0.062          |

Table 5. Hardness value test results (HRC).

| Sample          | Weld     | Heat-affected zone | Base metal |
|-----------------|----------|--------------------|------------|
| Original sample | 35.4     | 38.0               | 33.8       |
| HESP            | 36.0     | 39.0               | 35.6       |

Table 6. Ti–6Al–4V tensile strength and yield strength.

| Sample         | Tensile strength (MPa) | Yield strength (MPa) | Elongation (%) |
|----------------|------------------------|----------------------|----------------|
| Original sample| 860                    | 802                  | 4              |
| HESP           | 1023                   | 836                  | 8              |

3.4. Tensile test
Table 6 shows the Ti–6Al–4V tensile strength and yield strength. When the titanium alloy was subjected to HESP, the tensile strength was increased from 860 MPa to 1031 MPa and the yield strength increased from 802 MPa to 846 MPa. With surface nanocrystallization, both the tensile strength and the yield strength increased. The elongation rate also increased from 4% to 8%.

The tensile strength of the welded joints of the HESP specimens was improved. The HESP treatment also improved the fracture strength and toughness of the Ti–6Al–4V welded joints, indicating that it is an ideal method for improving the mechanical properties of the welded joints.

The fractures are all inclined approximately 45°, and the fractures in the heat-affected zone or the junction between the heat-affected zone and the parent metal exhibit a necking phenomenon. The fracture surface and the principal stress (tensile stress) are at an angle of approximately 45°, and the macroscopic fracture is cup-shaped.

Figure 7 shows that the fracture mouth has a dimple that is dense and small. Furthermore, the dimples after HESP are finer than the original dimples because the second phase in the matrix is separated from the matrix to form micropores, which grow and polymerize until the fracture is formed. The dimple depth and size differ. In some places, the fracture surface is relatively flat and exhibits the characteristics of brittle fracture, which is caused by the uneven plastic deformation of the material. In general, samples in the heat-affected zone or the heat-affected zone at the junction of the parent metal undergo ductile fracture. The fracture surface of the specimen fractured at the weld is perpendicular to the principal stress (tensile stress), and the fracture is grain-like, with obvious steps, which is caused by the expansion of the crack on the cleavage plane with different heights. However, a few dimples are also present because the material also undergoes slight plastic deformation before the brittle fracture occurs. In general, a test piece that breaks at the weld exhibits brittle fracture. This behavior is mainly due to the presence of incomplete penetration defects, which destroy the continuity of the grain boundaries. The fracture position of the welded joint is close to the weld in the heat-affected zone, indicating that the strength of the heat-affected zone adjacent to the weld is reduced and that the plasticity and toughness are greatly reduced. Therefore, the welding heat is minimized to improve the toughness index of the welded joint.

Through the observation and analysis of the fracture morphology of all tensile specimens, it is found that the surface of the fracture surface is bright and grainy, almost all of which belong to the form of cleavage fracture. It is also proved that the main reason for the poor joint performance is the grain size.
3.5. Residual stress test

The major problem in welded structures is the tensile residual stresses that are inevitably produced during the welding process. Surface tensile stresses can threaten the performance of the weldments because they act as an accelerant in fatigue crack initiation and failure [29]. High residual tensile stress is present in the weld zone and its surroundings, and the longitudinal stress is much larger than the transverse stress. The axial residual stress measurement results are shown in table 7.

According to table 7, the axial residual stress of the untreated original welded joint is tensile stress in the weld and its vicinity, and the compressive stress is at the base material away from the weld; in addition, the tensile stress at the weld is 650 MPa. The heat-affected zone is also under tensile stress, and its value is 331.7 MPa. However, the base-metal zone is under compressive stress due to the unaffected welding process, and the stress difference between the weld bead and the base metal reaches approximately 1000 MPa. This result is consistent with the distribution law of welding residual stress of welded joints. Compared with the weld, heat-affected zone and base metal, the HESP specimens exhibit the compressive stress; the difference in stress values between the latter three regions is less than 100 MPa.

The tensile residual stresses in welded structures reduce the tensile strength and plasticity in the original welded joint. After HESP, the residual stress of the welded joint changed from tensile stress to compressive stress. The compressive residual stress can be used as static stress and offset part of the tensile stress values during stretching [30], so the tensile strength in tensile test increased by 163 MPa. At the same time, the compressive residual stress that have smaller difference makes the deformation more uniform, so the plasticity is also improved. As shown in table 6, the elongation has doubled.

The compressive residual stress can also improve the fatigue life [31–33]. The compressive residual stress can increase fatigue crack threshold, so the cracks are difficult to initiate; Once the cracks have initiated, the compressive residual stress can block Fatigue crack propagation, and reduce the fatigue crack growth rates; Finally, the high compressive residual stress on the surface pushes the crack origin to subsurface layers.

4. Conclusions

In this paper, the influence of HESP on the microstructure and properties of Ti–6Al–4V welded joints was investigated. The HESP treatment was used to successfully prepare a uniform layer of nanolayers on the surface layer of the base material, heat-affected zone and weld zone of the welded joint. Homogenization of the surface nanocrystallization layer and surface hardness of the welded joint was realized. The results proved that the hardness of the surface layer of the three contiguous zones tended to be consistent. The HESP treatment led to higher hardness compared with that of the original sample. The treatment also formed a residual compressive stress in the three areas of the welded joint. The residual stress values of the three regions of the welded joint
tended to be homogeneous. After treatment, the tensile properties were improved, leading to increases in the residual compressive stress, tensile strength, yield strength and elongation. Due to the formation of the surface nanolayer, the tensile properties after HESP were improved.

ORCID iDs

He Xiaomei @ https://orcid.org/0000-0001-5780-1958

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