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Effect of shot peening on surface mechanical properties of TiB_2/Al composite

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Abstract The influence of shot peening on the surface mechanical properties of the TiB_2/6351Al composite has been investigated. The microstructures were determined by X-ray diffraction line profile analysis. The results showed that the increment of hardness was about 50% in the top surface layer. The matrix proof stress $\sigma_{0.2}$ of the shot peened surface had been increased by 27% and the whole strength increment was about 21% by considering the contribution of the reinforcements. The domain size and the dislocation density in the strengthened surface were 55 nm and 3.67×10^{15} m^{-2}, respectively. The mechanical properties improvement of the modified surface was partially due to the reinforcements but mainly due to the fine domains, high value of dislocation density induced by shot peening.

Introduction

Because of the formation of fine and stable ceramic reinforcements, the in-situ metal matrix composites (MMCs) are found to exhibit excellent physical and mechanical properties, such as specific modulus, strength, as well as thermostability etc., and have been widely concerned [1]. But the large difference in the coefficient of thermal expansion between the matrix and the reinforcements gives rise to residual stresses during manufacturing and subsequent heat treatment. Since these stresses are tensile in the matrix, they may deteriorate the fatigue properties. It is therefore of interest to apply a mechanical surface treatment to enhance the fatigue properties of the MMCs. Shot peening (SP), an effective method widely used in industry, can considerably improve fatigue strength and fatigue life of cyclically loaded metallic components by inducing compressive residual stress and work hardening into the surface region. Many studies dealt with the traditional metallic materials [2, 3], but the studies of the influence of shot peening on metal matrix composites are still insufficient.

The residual stresses of shot peened surfaces are often biaxial and the stress state keeps biaxial under uniaxial load [4]. However, in most of the previous works, only the stress in the loading direction was taken into account. Recently, a new method with a more explicit physical meaning had been put forward to measure the mechanical characteristics of thin films under a biaxial residual stresses state by X-ray stress analysis technique [4, 5]. Using this method, the proof stress of the shot peened surface layer can be determined. For a certain material, the mechanical properties mainly depend on its microstructures. X-ray diffraction (XRD) line profile analysis (LPA) is a powerful method for describing the microstructures of crystalline materials. Several new methods of LPA had been developed through which reliable results of domain size and dislocation density, etc., can be obtained easily [6, 7]. The present study deals with the influence of shot peening on the mechanical properties of the surface layer on TiB_2/6351Al Composite. Besides, the microstructures such as domain size, dislocation density in the peened surface were investigated. The results were also discussed.

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Experimental

The TiB_2/6351Al composite (in-situ, 10 vol.% TiB_2) used in this article was synthesized according to references [8]. The size of reinforcements is between 50 and 500 nm. Before SP, heat treatments were conducted after hot extrusion at a temperature of 450 °C using extrusion ratio of 10:1. The heat treatments condition: solution at 530 °C for 110 min, then quenched into the water, and finally aging at 170 °C for 6 h. The Young's modulus E and proof stress $\sigma_{0.2}$ of the composite are 80 GPa and 300 MPa, respectively, and the same parameters for the monolithic 6351Al alloy, under the same treatments, are 69 GPa and 235 MPa. Dog-bone specimens were cut from the center of the extrusion rods with effective dimensions of 40 × 5 × 2 mm^3. SP treatment was carried out at the both sides of the specimen: 0.3 MPa jet pressure, 1 min time, 0.25 mm average diameter of ceramic beads, 100 mm distance between nozzle and specimen, and 0.24 mm A Almen peening intensity. After SP, the hardness was measured by a HXD-1000B digital microhardness tester, loading weight 15 g, keep time 20 s.

For tensile test, a EMT1503 micro-tensile tester was built on the specimen and the standard distance is 20 mm. Incremental uniaxial tensile loading was applied step by step along the longitudinal direction (extrusion direction). At each step, the applied strain (ε_a) along the longitudinal direction was measured by strain gage technique. The longitudinal stress σ_1 and transverse stress σ_2 were determined by a X350A X-ray stress analyzer under each load level. Since Cr-Kα radiation was used, and the shifts of Al(311) diffraction profiles were detected, the effective penetration depth of the radiation at $\psi = 0^o$ should be 5.6 μm instead of 11 μm (Cu-Kα radiation, both were calculated on the basis of the diffraction intensity ratio of 1−1/e).

In order to investigate the microstructures of the shot peened surface, the reflection profiles of the composite and a standard sample were determined by a Dmax/rC diffractometer with Cu-Kα radiation, voltage 40 kV, current 100 mA, step size 0.01°. The standard sample is an Al powder which was annealed at 450 °C for 4 h.

Results and analysis

Microhardness measurements indicated a significant increment of hardness in the peened surface layer. Figure 1 shows the variation of microhardness along the depth from the top surface. In the top surface, hardness reaches 135 Hv, more than 50% higher than the substrate's one. And the depth of hardened layer is about 150 μm. SP is essentially a process of work hardening which will induce domain refinement and high density dislocation. In addition, compressive residual stresses were introduced into the surface layer. Former study showed that nanocrystalline can be formed in shot time, and the nanolayer has much higher hardness [9]. The increment of hardness is mainly due to fine domain, high value of dislocation density and compressive residual stress.

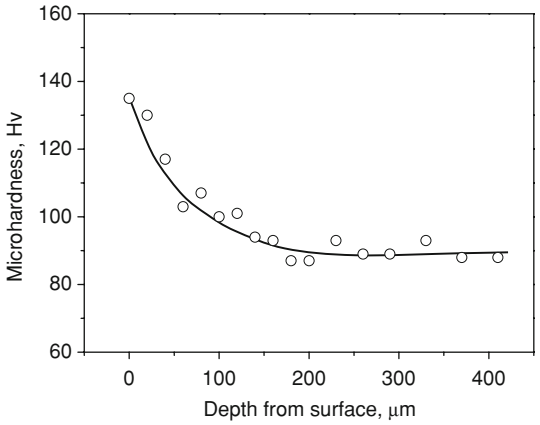

Fig. 1 Measured microhardness as a function of depth from the top surface

The surface yield strength can be determined using X-ray stress analysis techniques [4, 5]. In this method, it is assumed that the longitudinal and transverse directions of a plate specimen are the two principal directions of biaxial residual stress in the surface. And the stress state of the surface stays biaxial when external load is applied incrementally in the longitudinal direction. Then the longitudinal and transverse directions are still the principal directions of the surface stresses. SP residual stress states are approximately equibiaxial in the surface plane of the specimen but truly three-dimensional. Because the effective penetration depth of the X-ray beam is extremely shallow, the diffracting volume can be considered to represent a free surface under plane stress [10]. Thus, the stress in the direction normal to the specimen surface can be neglected. According to the Von Mises yielding criterion, only the equivalent stress versus equivalent strain curve can be properly characterizes the mechanical properties of specimen surface under biaxial stress state. The surface equivalent stress $\bar{\sigma}$ under each load can be expressed as

$$\bar{\sigma} = (\sigma_1^2 - \sigma_1\sigma_2 + \sigma_2^2)^{1/2} \tag{1}$$

For uniaxial equivalent strain$\bar{\varepsilon}$, in the elastic deformation stage

$$\bar{\varepsilon} = \bar{\varepsilon}_e = \bar{\sigma}/E \tag{2}$$

and in the plastic deformation stage

$$\bar{\varepsilon} = \bar{\varepsilon}_e + \bar{\varepsilon}_p \tag{3}$$

where E is the Young's modulus, $\bar{\varepsilon}_e$ is the elastic strain and $\bar{\varepsilon}_p$ is the plastic strain. Using σ_1', σ_1'', σ_2', σ_2'', ε_a', ε_a'', $\bar{\sigma}'$, and $\bar{\sigma}''$ to

indicate the measured values of σ_1, σ_2, applied strain ε_a, and $\bar{\sigma}$ of two adjacent measurement points. The increments of equivalent plastic strain $\Delta\bar{\varepsilon}_p$ for all intervals are

$$\Delta\bar{\varepsilon}_p = \frac{\bar{\sigma}' + \bar{\sigma}''}{\left|\sigma_1' + \sigma_1'' - (\sigma_2' + \sigma_2'')/2\right|}\left|\Delta\varepsilon_a - \frac{\Delta\sigma_1 - \nu\Delta\sigma_2}{E}\right| \quad (4)$$

where $\Delta\sigma_1 = \sigma_1'' - \sigma_1'$, $\Delta\sigma_2 = \sigma_2'' - \sigma_2'$, $\Delta\varepsilon_a = \varepsilon_a'' - \varepsilon_a'$. Then the total equivalent plastic strain $\bar{\varepsilon}_p$for every measurement point can be obtained by successive summation.

The relationships between σ_1, σ_2, and ε_a of the untreated and shot peened surface layers are given in Fig. 2. For each, the point indicated by the arrow is regarded as the boundary between the elastic and plastic stage [11]. E can be obtained by linear regression using the data of the elastic stage in σ_1–ε_a relationship, which is equal to the slope of the regression line. The results show that both the E of untreated and shot peened samples are 70 ± 1 GPa. In the present study, what measured by XRD is the stress in the matrix of the composite. Therefore, both the results are approximated to the E of monolithic 6351Al alloy. In the plastic stages, the relationship curves of σ_1–ε_a deviate from the liner behaviors. The variations of transverse stresses σ_2 of the both specimens are not obvious during the uniaxial tension.

Using the method described above and the data in Fig. 2, the relationships between the equivalent stress $\bar{\sigma}$ and the uniaxial equivalent strain $\bar{\varepsilon}$ are obtained for both samples, as shown in Fig. 3. The dash lines are drawn to determine the proof stresses $\sigma_{0.2}$ corresponding to the permanent plastic strain of 0.2%. The obtained $\sigma_{0.2}$ of the untreated and shot peened surfaces are 221 MPa and 281 MPa, respectively. The $\sigma_{0.2}$ of the untreated surface is approximated to the proof stress of the unreinforced alloy. While for the shot peened surface, the value is increased by about 20%. Comparing $\sigma_{0.2}$ of the untreated surface and the shot peened surface, the increment caused by SP is $(281 - 221)/221 \approx 27\%$. In the present study, the measured $\sigma_{0.2}$ is essentially the proof stress of the matrix in the composite. The actual stress bearing factor of the matrix R along the loading direction were estimated using the following equation,

$$\Delta\sigma_1^r = R\sigma_a \quad (5)$$

where $\Delta\sigma_1^r = \sigma_1^r - \sigma_1^0$, and σ_1^r is the residual stress after unloading at each step, $\sigma_1^0$is the residual stress right after shot peening, σ_a is the applied stress. The relationships between σ_a and $\Delta\sigma_1^r$ were shown in Fig. 4. The R obtained from the slope of the linear regression line was 0.81 for the unpeened sample, which increased slightly to 0.83 after shot peening. Thus, it is reasonable to assume that the contribution ratio of the reinforcements on the whole strength as

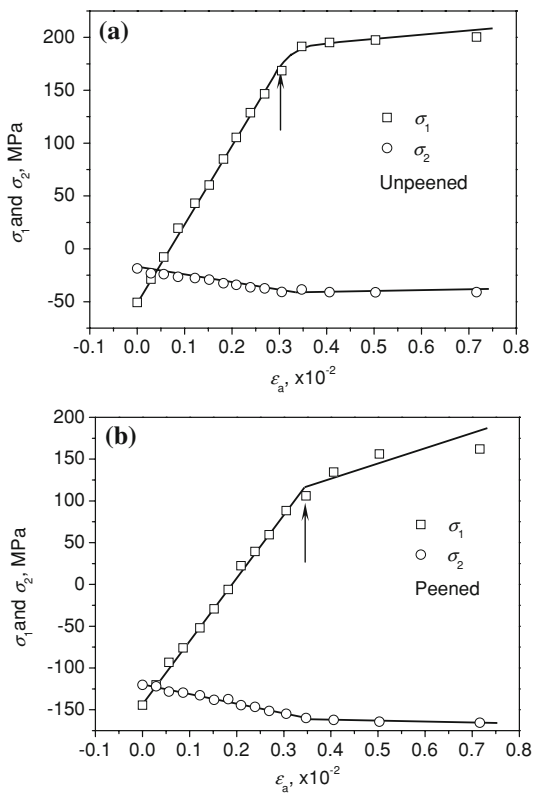

Fig. 2 Variations of the longitudinal stress (σ_1) and transverse stress (σ_2) due to the applied tensile strain (ε_a) **a** unpeened, **b** shot peened

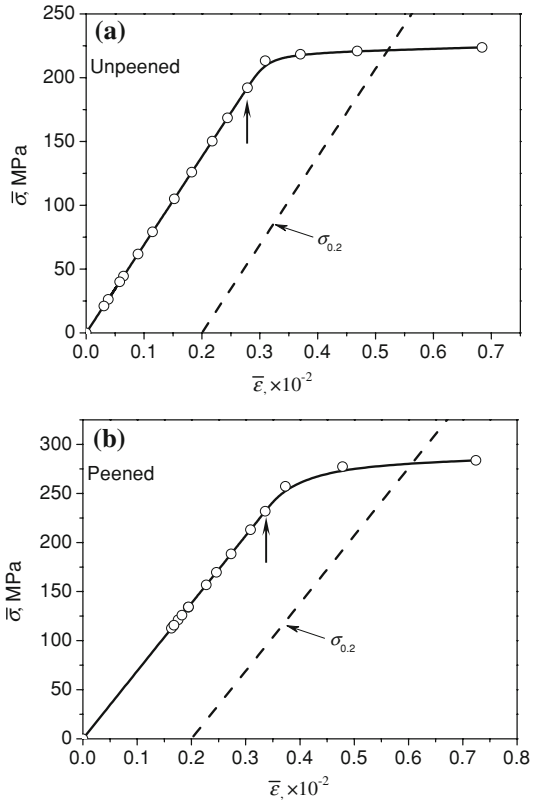

Fig. 3 Relationships between the equivalent stress ($\bar{\sigma}$) and equivalent uniaxial strain ($\bar{\varepsilon}$) under the tensile load **a** unpeened, **b** shot peened

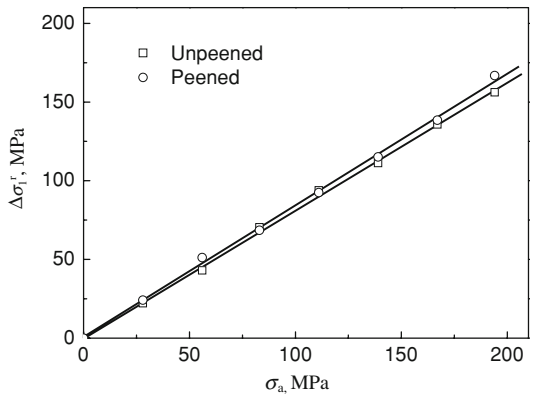

Fig. 4 Relationships between $\Delta\sigma_1^r$ and the applied stress σ_a

constant. As the strength of the matrix had been improved by shot peening, the contribution of the reinforcements to the whole strength increased also. Then the real proof stress of the shot peened surface is $281 + (300 - 235)(1 + 27\%) \approx 364$ MPa. And the real increment of the strength is $(364 - 300)/300 \approx 21\%$. These results indicate that SP is an effective method to improve the surface strength of the TiB_2/6351Al composite.

The mechanical properties of metallic materials are directly depended on their microstructures. The domain size and the dislocation density of the shot peened surface matrix were determined by modified Warren-Averbach (MWA) method [6, 7, 12]. In X-ray diffraction, domain means a region over which diffraction is coherent. It can be dislocation cell, subgrain, or the region between fault planes [13]. The results of XRD line profile analysis is an average ones. The XRD patterns of the shot peened surface and the standard sample are shown in Fig. 5. The basic equation of MWA method is

$$\ln A(L) \cong \ln A^s(L) - \rho B L^2 \ln(R_e/L)(K^2\overline{C}) + o(K^4\overline{C}^2) \quad (6)$$

where $A(L)$is the real part of the Fourier coefficients and it can be obtained through Stokes deconvolution [14], $A^s(L)$

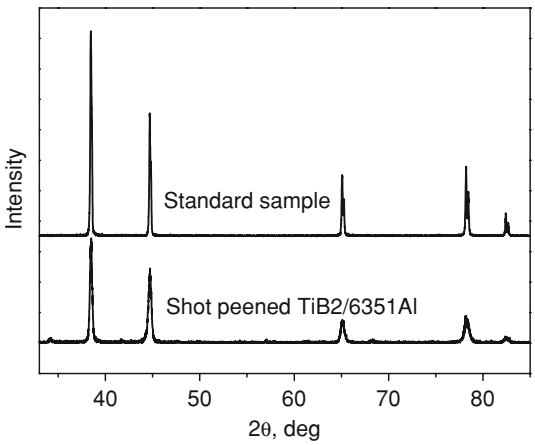

Fig. 5 XRD patterns of shot peened and standard samples

is the size Fourier coefficient, $B = \pi b^2/2$, R_e is the effective outer cut-off radius of dislocations and o stands for higher order terms in $K^4\overline{C}^2$. L is the Fourier length defined as $L = na_3$, where $a_3 = \lambda/2(\sin\theta_2 - \sin\theta_1)$, n are integers starting from zero and$(\theta_2 - \theta_1)$is the angular range of the measured diffraction profile. $K = 2\sin\theta/\lambda$, the average contrast factors $\overline{C}$ can be calculated from reference [12, 15]. The size parameter corresponding to the Fourier coefficients is denoted by L_0. It is deduced from the size Fourier coefficients $A^s(L)$ versus L by least-squares method according to the formula provided by Wang [16]

$$A^s(L) = a - L/L_0 \quad (7)$$

where a is the quantity expressing the "hook" effect. Denoting the coefficients of the second term on the right hand side of Eq. 6 by $X(L)$, the average dislocation density ρ can be determined from the following relation

$$X(L)/L^2 = B\rho \ln R_e - B\rho \ln(L) \quad (8)$$

Using Eqs. 6–8 and the data in Fig. 5, the values of domain size L_0 and dislocation density ρ are obtained to be: $L_0 = 55$ nm, $\rho = 3.67 \times 10^{15}$ m^{-2}. This value for the dislocation density is consistent with the value reported by Ungár [7] and considered a high value in ordinary material. Fine domains and a large amount of dislocations are generated during the process of SP. In deformed metals, microstrain mainly comes from dislocations because not only they are always the main defects, but also they are the major components or can play an important role in other defects [17, 18]. For metal matrix composites, the reinforcement particles in matrix always act as sink sources of dislocations during repeated deformation [19].

In the present study, the Young's modulus of the reinforcements TiB_2 (500GPa) is quite higher than the matrix's one. Therefore, the microstructures of the reinforcements were hard to change during the process of SP. And due to the compressive residual stresses in the surface layer, the microcracks were not easy to nucleate and propagate in the reinforcement/matrix interface. The contribution of the reinforcements to the whole strength increased with the improvement of the matrix strength. Therefore, the increment of the mechanical properties of the strengthened surface was partial due to the reinforcements but mainly due to the fine domain, high value of dislocation density induced by shot peening.

Conclusions

The mechanical properties of the shot peened surface on the TiB_2/6351Al composite were investigated and the microstructures were determined using LPA. Results showed that the proof stresses $\sigma_{0.2}$ of the matrix in the shot

peened and untreated surfaces are 281 MPa and 221 MPa, respectively. And the whole strength increment was about 21% by taking account of the contribution of the reinforcements. The hardness of the shot peened surface had been improved by more than 50%. Microstructures investigation revealed that fine domain and high density dislocations had been introduced to the strengthened surface, which is 55 nm and 3.67×10^{15} m^{-2}, respectively. The contribution of the reinforcements to the whole strength increased with the matrix strength improving. Nevertheless, the improvement of the surface mechanical properties was mainly due to the fine domain, high value of dislocation density. SP is an effective method to improve the surface strength of the TiB_2/6351Al composite.

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