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Dynamic Behavior of a Novel High-Strength and Ductile Near-α Titanium Ti-Al-Mo-Zr-Fe-B Alloy

Chi Yan 1, Chu Wang 1, Miaoxia He 1, Yuecheng Dong 1,* , I. V. Alexandrov 2 and Hui Chang 1,*

1 College of Materials Science and Engineering, Tech Institute for Advanced Materials, Nanjing Tech University, Nanjing 211816, China
2 Department of Materials Science and Physics of Metals, Ufa State Aviation Technical University, 450008 Ufa, Russia
* Correspondence: dongyuecheng@njtech.edu.cn (Y.D.); ch2006@njtech.edu.cn (H.C.); Tel.: +86-25-8358-7270 (Y.D.)

Abstract: In this study, the dynamic compression properties of a new high-strength (>1000 MPa) and ductile (>15%) near-α titanium Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy were investigated at high strain rates of 1620 s⁻¹~2820 s⁻¹ by a split Hopkinson pressure bar (SHPB). The microstructural evolution of the samples before and after the dynamic deformation was analyzed by electron backscatter diffraction (EBSD). The results indicated that the strength of the alloy enhanced significantly under the dynamic loading compared with the quasi-static compression and increased with the increase in the strain rate. An abundance of deformation twins released the dislocation pile-up and coordinated the plastic deformation of alloy during the dynamic loading. The dynamic plasticity constitutive equation of the alloy was obtained by fitting high strain rate experimental data at room temperature by the Johnson–Cook constitutive equation with the modified temperature term.

Keywords: near-α titanium alloy; dynamic deformation; deformation twin; Johnson–Cook equation

1. Introduction

Near-α titanium alloys have high specific strength, excellent corrosion resistance, and good weldability, making them extensively accepted as admirable structural materials in marine engineering [1–5]. In particular, the Ti80 (Ti-6Al-3Nb-2Zr-1Mo) alloy is widely studied and applied as a typical marine titanium alloy because of its corrosion resistance [6,7], impact toughness [8], low fatigue cracking [9,10], and creep behavior [11] as well as its performance after various welding methods [12,13]. However, the presence of the Nb element makes Ti80 expensive; it also gives it a high melting point of up to 2468 °C. The high melting point makes it possible to generate a high density of inclusions in the casting process, which is harmful to the subsequent forging/rolling process. Moreover, the strength of the Ti80 is approximately equal to 800 MPa, which is not high enough as the pressure shell of the 10,000 deep-diving submersible dives into the depth of up to 10,000 m. That is why the new type, the Ti80 alloy, was used in spite of poor weldability.

On the other hand, titanium alloys as structural materials are often subjected to high velocity impact loads, which are required to ensure the structural integrity and continuity with the required dynamic load carrying capacity under the specified impact loads. Since both plasticity and failure behavior of titanium are significantly affected by the strain rate, it is necessary to study the dynamic behavior of titanium at different strain rates [14–16]. Hence, in this work, a novel near-α titanium Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy, based on the Ti80 alloy, was designed, in which Nb was replaced by the microalloying elements Fe and B. The aim of the present investigations was to design a novel, low-cost, high-strength, and high-toughness near-α titanium alloy for marine engineering applications and evaluate the dynamic compression properties of this alloy at different strain rates with the split Hopkinson pressure bar (SHPB).
2. Materials and Methods

A near-α titanium Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy was designed and melted using vacuum arc fusion (VAR) equipment. The chemical composition of the alloy was investigated by inductively coupled plasma mass spectrometry (ICP-MS) analysis; the actual chemical composition is shown in Table 1. The phase transition temperature of the alloy measured by the metallographic method was 995 °C. After forging, heat treatment [17] was used to optimize the microstructure.

Table 1. Chemical composition of the near-α Ti-6Al-1Mo-2Zr-0.55Fe-0.1B (wt.%) alloy.

| Elements | Al   | Mo   | Zr   | Fe   | B    | Ti   |
|----------|------|------|------|------|------|------|
| wt.%     | 5.8  | 1.07 | 1.85 | 0.57 | 0.044| Bal. |

Quasi-static tensile tests were conducted on the MTS 370.10 hydraulic servo fatigue tester. The tensile specimen length was 100 mm, and the diameter and length of the gauge were ø5 mm and 25 mm, respectively. The strain rate was $7 \times 10^{-3}$ s$^{-1}$. Quasi-static compression experiments were carried out on a universal testing machine with a strain rate of $10^{-3}$ s$^{-1}$ with the dimension of ø3 × 6 mm, which was the sample size. Dynamic mechanical properties were tested by the SHPB equipment at high strain rates of 1620 s$^{-1}$~2820 s$^{-1}$, and the sample size was ø6 × 3 mm. The diameter of the SHPB compression bar was 14.5 mm, and the input pulse maximum value was 1200 MPa. The samples were prepared by wire cutting, and the end faces were polished by sandpaper grinding.

The Japanese Regulus 8100 cold field emission scanning electron microscope was used to characterize the microstructure of the samples as well as the element distribution; the sample size was 4 mm × 4 mm. After dynamic deformation, EBSD samples were prepared by slicing parallel to the compression direction, and the samples were polished using a Shanghai metallographic PFD-2 electrolytic polishing machine with an electrolytic solution of 12 mL perchloric acid, 120 mL methanol, and 68 mL n-butanol. The polishing temperature was stabilized at −20 °C by adding liquid nitrogen; the polishing current was 0.6~0.7 A, and the polishing time was about 50 s. The surface of the specimen was cleaned with alcohol after polishing. The EBSD was carried out on the field emission scanning electron microscope (SEM, JSM-6700F), equipped with an Oxford Instruments EBSD detector and working at an accelerating voltage of 20 KV. The step size of the samples was 0.05 µm with a scanning area of 1600 µm$^2$. The HKL Technology Channel 5 system was used to analyze the EBSD test data.

3. Results and Discussions

3.1. Initial Microstructure

The SEM microstructure and composition distribution map of the Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy is illustrated in Figure 1. Obviously, the bimodal microstructure with the primary equiaxed α phase and β transformation matrix was achieved by the heat treatment, which was accepted as a result of the simultaneous combination of good strength and ductility [18,19]. The size of the primary equiaxed α phase was about 5.25 µm, occupying 21% of the content. The thickness of the secondary lamellar α phase was about 0.52 µm. On the other side, the distribution of each alloying element showed sufficient homogeneity, as indicated in Figure 1b.
Figure 1. SEM results: (a) microstructure and (b) composition distribution maps of the Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy.

3.2. Quasi-Static Tensile and Compression Experiment

Figure 2a shows the true stress—strain curve for the alloy at the strain rate of 0.007 s\(^{-1}\). Result indicated that the yield stress was 927 MPa; the stress increased slightly with the increase in the strain, and the ultimate tensile stress could reach 1005 MPa. In comparison with a slight increase in the work hardening effect during the tensile test, the alloy showed a significantly higher work hardening during the quasi-static compression experiment (Figure 2b). Simultaneously, the ultimate compression stress increased up to 1370 MPa from 1081 MPa of tensile yield stress. Table 2 shows the quasi-static mechanical property data of the new titanium alloy and the Ti80 alloy with the bimodal microstructure. It can be seen that the strength of the novel alloy is superior to that of the Ti80 alloy. Meanwhile, good ductility is kept at the same level.

![Graphs showing tensile and compression stress-strain curves](image)

Figure 2. Quasi-static mechanical properties of the Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy: (a) tension; (b) compression.

Table 2. Quasi-static mechanical property data of the Ti-6Al-1Mo-2Zr-0.55Fe-0.1B and Ti80 alloys.

| Alloy Composition | Tensile Strain (%) | Compr. Strain (%) |
|------------------|--------------------|-------------------|
| Ti-6Al-1Mo-2Zr-0.55Fe-0.1B | 15.1 | 60 |
| Ti-6Al-2Zr-1Mo-3Nb [20,21] | 22 | 50 |
| Ti80 | 1018 | UCS/MPa |

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| Ti80 | 1018 | UCS/MPa |

![](image)
3.3. Dynamic Compression Experiment

Dynamic true stress–strain curves of the Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy at different strain rates are shown in Figure 3a. Obviously, compared to the quasi-static stress–strain curve (Figure 2b), they had no obvious yield stage and strain hardening during dynamic loading. The strength was higher during the high strain rate deformation and increased with the increase in the strain rate. Meanwhile, all the dynamic stress–strain curves fluctuated at the plastic deformation stage. The main data extracted from the dynamic stress–strain curves of the novel alloy and the Ti80 alloy are listed in Table 3. Among them, the absorbed impact energy could be considered as the area under the stress–strain curve of the uniform plastic deformation, which could reflect the dynamic mechanical property of materials more accurately by considering the combination of the strength and plasticity. The formula for calculating the absorbed impact energy is as follows [22]:

\[ E = \int_{\varepsilon_s}^{\varepsilon_f} \sigma \cdot d\varepsilon \]  

(1)

where \( \varepsilon_f \) and \( \varepsilon_s \) are strains at the beginning and the end of the plastic deformation, respectively. It can be seen that the absorbed impact energy of the novel alloy increases with an increase in the strain rate, exhibiting a higher value than the Ti80 alloy at the similar strain rate.

![Figure 3](image-url)  

**Figure 3.** (a) Dynamic true stress–strain curves and (b) strain hardening rate of the Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy.

| Alloy Composition | Ti-6Al-1Mo-2Zr-0.55Fe-0.1B | Ti-6Al-2Zr-1Mo-3Nb [21] |
|-------------------|-----------------------------|--------------------------|
| Strain rate       | 1620 s\(^{-1}\)             | 1500 s\(^{-1}\)          | 2500 s\(^{-1}\)          | 3500 s\(^{-1}\)          |
| UCS /MPa          | 1398 1438 1454 1567         | 1348 1418 1478           |
| Strain            | 0.21 0.21 0.31 0.39         | 0.13 0.13 0.22 0.30      |
| Akv/MJ m\(^{-3}\) | 229 240 335 515             | 158 299 422              |

The strain hardening rate \( \theta \) can be used to express the ability of a material to resist plastic deformation, where \( \theta = d\varepsilon / d\sigma \). Figure 3b shows the strain hardening rate of samples at the high strain rates of 1620 s\(^{-1}\) and 2820 s\(^{-1}\). The strain hardening rate curves developed similarly at the beginning of the deformation, and the alloy underwent continuous plastic deformation, which led to the first peak stress at the strain 0.04. After that, the development of the strain hardening rate was non-monotonous. This implies that the thermal softening effect happened during the dynamic loading. It is well known that adiabatic heating is always generated as a result of accumulated plastic work during the dynamic loading.
when the generated heat cannot dissipate outside in a short time, which increases the local
temperature in materials and leads to the dislocation annihilation. The wave-shape curves
indicate the competition of strain hardening and thermal softening of the alloy during
dynamic loading [23]. When the strain hardening rate $\theta > 0$, the strain hardening effect
dominates in the plastic deformation of the alloy, and when $\theta < 0$, it shifts to be dominated
by the thermal softening effect. Obviously, for the investigated deformation with two strain
rates, the final hardening rate was negative at the strains of 0.19 and 0.3, which shows the
advantage of the thermal softening effect.

The inverse pole figures for the initial and dynamically deformed samples are shown
in Figure 4a–c. The results indicated that the initial microstructure was homogeneous
as a result of heat treatment, and the grain size was about 6.43 $\mu$m. After the dynamic
compression, the grains suffered fragmentation, and the grain size decreased up to 5.51 $\mu$m
and 5.3 $\mu$m after high strain rates of 1620 s$^{-1}$ and 2820 s$^{-1}$, respectively. On the other
hand, deformation twins activated in the microstructure as a result of dynamic loading
and promoted the increase in the strain rate. The morphology of the deformation twins
showed diversity under dynamic loading, including twin boundaries terminated at the
grain boundaries, those terminated within the grains, and even the generated penetrating
twins. Meanwhile, a twin wafer-like microstructure was found in the sample subjected by
dynamic loading at 2820 s$^{-1}$.

![Figure 4](image.png)

**Figure 4.** EBSD results: (a–c) inverse pole figures; (d–f) kernel average misorientation maps;
(g–i) grain boundary orientations for (a,d,g) initial, (b,e,h) deformed with strain rate 1620 s$^{-1}$, and
(c,f,i) deformed with strain rate 2820 s$^{-1}$ samples.

Figure 4d–f shows the kernel average misorientation (KAM) plot of the new alloy,
which can represent the density distribution of the geometrically necessary dislocation
(GND) boundaries macroscopically. Those boundaries with KAM $> 0.5^\circ$ are considered
as high-density dislocation boundaries (shown in green in the KAM plot), and those with
KAM $\leq 0.5^\circ$ are considered as low-density dislocation boundaries (shown in blue in the KAM plot). In the initial state, high-density dislocations were rare and only existed as a result of previous heat treatment. However, the dislocation density of the sample increased significantly after dynamic loading, and dislocations were distributed not only at grain boundaries but also in grain interiors, which showed strong strain rate dependence. On the other hand, the fraction of low-angle grain boundaries (LAGBs) increased with the increase in the strain rate, and the fraction of LAGBs increased to 68.84% and 82.37% after the high strain rates of 1620 s$^{-1}$ and 2820 s$^{-1}$, respectively. Some other relevant parameters of different state samples obtained through EBSD data statistics are listed in Table 4.

| State       | Avg GS (μm) | HAGB (%) | LAGB (%) | The Fraction of Grain in Different States (%) |
|-------------|-------------|----------|----------|---------------------------------------------|
| Initial     | 6.43        | 45.58    | 54.42    | Recrystallized 44.4, Substructured 51.9, Deformed 3.7 |
| 1620 s$^{-1}$ | 5.51        | 68.84    | 31.16    | 6.6, 67.6, 25.8 |
| 2820 s$^{-1}$ | 5.3         | 82.37    | 17.63    | 5.2, 21.7, 73.1 |

Figure 5 shows the pole figures of the initial sample and the samples after the dynamic impact testing at 1620 s$^{-1}$ and 2820 s$^{-1}$, respectively. The maximum texture intensity of the initial sample was 18.12. The sample after the impact testing with 1620 s$^{-1}$ and 2820 s$^{-1}$ showed a maximum texture intensity of 24.84 and 23.66 along the $<0001>$ pole. Thus, more grains in the deformed samples must be aligned along the $<0001>$ pole direction, demonstrating the large influence of the strain rate on the crystallographic texture evolution.

![Figure 5](image_url)

**Figure 5.** Pole figures of (a) initial, (b) deformed with strain rates 1620 s$^{-1}$, and (c) 2820 s$^{-1}$ samples.

It is worth mentioning that the high-strength titanium alloy kept integrity and did not show any instability after dynamic loading, showing excellent comprehensive mechanical properties. The new Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy is classified as the near-$\alpha$ titanium alloy according to the $K_\beta$ stability coefficient calculation [24]. As is commonly known, the $\alpha$ phase of a titanium alloy has a hexagonal close-packed (HCP) structure, with only three slip systems, but the dislocation slip on prismatic surfaces is facilitated to launch due to the $c/a$ value (1.587) below the ideal HCP value (1.633). Furthermore, the addition of Al elements is considered to reduce the $c/a$ value and benefit the dislocation movement [25–27]. Additionally, an abundance of deformation twins released the dislocation pile-up and
coordinated the plastic deformation of the alloy during the dynamic loading [28,29]. On the other hand, the damage of materials under the dynamic loading was always related to the occurrence of adiabatic shear bands, which were considered to accompany the dynamic recrystallization process [30–32]. In our investigation, the adiabatic shear bands have not been observed by optical microscopy and EBSD, and the content of recrystallized grains decreased seriously; even the imposed strain rate was equal to 2820 s\(^{-1}\), as indicated in Table 4. Obviously, the novel titanium alloy has a good potential for the applications due to the simultaneous high strength and ductility, although the deformation mechanism still needs to be revealed further.

4. The Johnson–Cook Constitutive Equation

As mentioned above, when materials suffered high strain rate deformation, the plastic work generated from the deformation could be converted into heat energy, which increased the local temperature of materials. Theoretically, the temperature rise calculation was carried out by the following equation:

\[
\Delta T = \beta \frac{W_p}{\rho C_p}
\]

(2)

where \(\beta\) is the ratio of the mechanical energy transferred into the thermal energy, generally taken as 0.9; \(W_p\) is the plastic work; \(W_p = \int_0^{\varepsilon_f} \sigma \, d\varepsilon\); \(\rho\) is the material density; \(\rho = 4.45 \text{ g/cm}^3\); and \(C_p\) is the constant pressure specific heat capacity of the material 0.526 J/g K. The temperature rise of the new alloy at different strain rates can be obtained from the Equation (2); the result is presented in Figure 6. The fitting can be obtained as follows:

\[
\Delta T = 26.478e^{0.0007\dot{\varepsilon}}
\]

(3)

![Figure 6. Temperature rises of the Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy at different strain rates.](image)

Obviously, the temperature rises of the new alloy increase exponentially under the dynamic loading, especially when the strain rate exceeds 2000 s\(^{-1}\).

The thermal-viscoplastic constitutive model is commonly used to describe the dynamic mechanical response of metallic materials, such as the Johnson–Cook model, the Zerrilli–Armstrong model, and the Steinberg–Guinan model. Among them, the Johnson–Cook model is simple and can better describe the strain hardening, strain rate effects, and thermal softening effect of the material. In this paper, the Johnson–Cook model is used to describe the dynamic intrinsic structure of the new titanium alloy, and the model expression is as follows:

\[
\sigma = (A + B\varepsilon^n) \left[1 + C \ln\left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right)\right] \left(1 - T^m\right)
\]

(4)
where $A$, $B$, $C$, $n$, and $m$ are coefficients to be determined; $\sigma$ is the flow stress; $\varepsilon$ is the equivalent plastic strain; and $\dot{\varepsilon}_0$ and $\dot{\varepsilon}$ are the reference strain rate and the equivalent plastic strain rate. $T^*$ is the dimensionless temperature ($T^* = (T - T_0)/(T_m - T_0)$), where $T_0$, $T$, and $T_m$ are the reference temperature, the deformation temperature, and the melting point temperature of the material, taking $T_0 = 293$ K and $T_m = 1800$ K.

When $\dot{\varepsilon}_0 = 10^{-3}$ s$^{-1}$, $T = T_0$, for the compression experiment performed under quasi-static loading at room temperature, the equation can then be decoupled as follows:

$$\sigma = A + Be^n$$  \hspace{1cm} (5)

It is rewritten as $\ln(\sigma - A) = \ln B + n \ln \varepsilon$, substituted into the stress–strain data under the quasi-static compressive loading at room temperature for linear fitting, and the final strain-hardening parameters are obtained as $A = 1018$ MPa, $B = 774$ MPa, $n = 0.97$.

The strain rate strengthening parameter $C$ corresponds to the strain rate effect of the material and can be fitted using stress–strain data for different strain rates at room temperature, where the equation is as follows:

$$\sigma = \left(1018 + 774 \cdot \epsilon^{0.97}\right) \left[1 + C \cdot \ln \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right)\right]$$  \hspace{1cm} (6)

Obviously, the value of $C$ is the slope of $\sigma/(1018 + 777 \cdot \epsilon^{0.97}) - 1 = C \cdot \ln \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right)$, and after substituting the stress–strain data for different strain rates at room temperature, the value of the strain rate strengthening parameter $C$ is 0.013.

Using the adiabatic temperature rise in Equation (5) plus the reference temperature as the deformation temperature, brought into Equation (6), the value of $m$ is obtained by fitting it to the experimental data and taking the average value of 1.72. Then, the complete J–C constitutive equation is as follows:

$$\sigma = \left(1018 + 774 \cdot \epsilon^{0.97}\right) \left[1 + 0.013 \cdot \ln \left(\frac{\dot{\varepsilon}}{0.001}\right)\right] \left(1 - T^*^{1.72}\right)$$  \hspace{1cm} (7)

Figure 7 shows the results of the fitting of the experimental data of the alloy at high strain rates. The J–C constitutive equation constructed in this paper can effectively predict the mechanical response of the alloy under high strain rate deformation, especially in the plastic deformation stage.
5. Conclusions

(1) A novel near-α titanium Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy was developed, which possesses simultaneous high strength (>1000 MPa) and ductility (>15%).

(2) The strength of the alloy under dynamic loading enhanced significantly compared with the quasi-static compression and increased with the increase in the strain rate. An abundance of deformation twins released the dislocation pile-up and coordinated the plastic deformation of the alloy during the dynamic loading.

(3) The Johnson–Cook constitutive equation was modified by the adiabatic temperature rise term, and a nonlinear fit was performed to construct a dynamic constitutive relationship at room temperature, which could describe the rheological behavior of the alloy under high strain rate conditions at room temperature.

The novel near-α titanium Ti-6Al-1Mo-2Zr-0.55Fe-0.1B alloy possesses excellent mechanical properties, a low-cost design, and good weldability, making it a good potential application in marine engineering.

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