Dynamic forced shear characteristics of Ti-6Al-4V alloy using flat hat-shaped specimen

Chun Ran\textsuperscript{a,b,⁎}, Qingqing Liu\textsuperscript{b}, Pengwan Chen\textsuperscript{b,⁎}, Qi Chen\textsuperscript{a}

\textsuperscript{a}School of Materials Science and Engineering, Beijing Institute of Technology, Beijing 100081, China
\textsuperscript{b}State Key Laboratory of Explosion Science and Technology, Beijing Institute of Technology, Beijing 100081, China

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\textbf{ABSTRACT}

The relation between stress collapse and nucleation of adiabatic shear band (ASB) remains incompletely understood. To further understand this topic, the evolution of shear deformation of Ti-6Al-4V alloy under forced shear loading is systematically investigated using digital image correlation (DIC) technique. DIC results indicate that ASB nucleates after the maximum stress and drastic stress drop corresponds to the initiation and propagation of crack, and the forced shear process can be divided into homogeneous, inhomogeneous and highly localized deformation stages. Numerical simulation results suggest that thermal softening might not have had a pronounced effect on the onset of adiabatic shear band and dynamic recrystallization formation. "Cohesive fracture" can be identified as the dynamic failure mechanism for Ti-6Al-4V alloy on the basis of the crack propagation features, and the microstructure gets increasingly refined due to the occurrence of dislocations, stacking faults and cell structures.

\section{1. Introduction}

Ti-6Al-4V alloy is of great application value in aerospace, shipbuilding and ordnance industry due to its high strength-to-weight ratio, high-temperature resistance and strong corrosion resistance [1]. However, in practical applications, components and structures made by Ti-6Al-4V alloy are often involved high strain rate deformation during their service. In addition, Ti-6Al-4V alloy is very prone to adiabatic shear localization (ASB) fail under high strain rates loading [2]. Therefore, to identify the dynamic failure behavior of Ti-6Al-4V alloy, a considerable number of investigations on Ti-6Al-4V alloy under high strain rate loading have been conducted [3–9].

The classical explanation of adiabatic shear fail is the competition between hardening effect (strain and strain rate) and thermal softening effect. When softening effects overcome hardening ones, the current state of the material indeed becomes unstable and the overall loading is accommodated by the formation of narrow bands, the adiabatic shear bands in question, in which the deformation progressively concentrates [10]. Hence, thermal softening was believed to be the dominant mechanism that triggers ASB [11]. However, the whole process is very complex, and involves high strain rates, high local temperature, large plastic deformation, etc. [12].

To identify the characteristics of the dynamic deformation process, Marchand and Duffy [13] measured the local strain of HY-100 steel using high-speed cameras. Their results indicated that catastrophic failure corresponds to the formation of the shear band. By contrast, the studies of Duffy and Chi [14] and Cho et al. [15] showed that the sudden drop only corresponds to crack propagation. In contrast to the above results, a recent study revealed that stress collapse is ca several microseconds before ASB formation [16].
To summarize this brief literature survey, it can be observed that the relation between stress collapse and nucleation of ASB is still a debate issue. The problem is that the spatial resolution of the grid patterns (98 lines per cm) deposited on the sample is insufficient to reveal the initiation of ASB. Therefore, one purpose of this work is to clarify the relation between stress collapse and nucleation of ASB by enhancing the spatial resolution of the sample.

A related subject is the evolution of microstructures. Peirs et al. [17] addressed that the microstructural evolution process of Ti-6Al-4V alloy deformed at high strain rates can be divided into three stages: e.g., onset of strain localization, formation of ASBs and initiation and propagation of microcrack. Timothy and Hutchings [2] and Xue et al. [18] found that the main stages of void evolution within shear bands in Ti-6Al-4V alloy are nucleation, growth, and coalescence. Similar coalescence of microvoids is also found in ultrafine-grained pure titanium [19]. Moreover, many elliptical and/or spherical cavities can be observed within the ASBs, and the elongated cavities were induced by void coalescence [9].

To date, the relation of voids, microcrack and ASB has explicitly been addressed. However, the propagation of crack induced by voids and/or microcrack within ASB is rarely reported, and the underlying mechanism remained unclear. Thus, the second purpose of this study is to gain deep insight into the crack propagation and bring experimental evidence on crack propagation path in the microstructures.

2. Materials and methods

2.1. Materials and specimen geometry

Commercially Ti-6Al-4V alloy was selected in this study for its marked propensity to fail by ASB formation [8]. The β-transus temperature is approximately 990 °C via metallographic observation method. Hat-shaped specimen, originally proposed by Meyer and Hartmann, was proved to be a useful sample to characterize the shear behavior of materials [20]. However, the shear stress measurement is influenced by edge effects [21]. In addition, the real-time local strain evolution cannot be revealed. Therefore, to reduce edge effect of hat shaped specimen in the deforming gage section in real-time during the plastic deformation, flat-hat shaped specimen (FHS), designed by Clos et al. [22], was applied in this study, and they were machined from the as-received bar. The overall configuration of the FHS specimen is illustrated in Fig. 1a. It is well known that axi-symmetric where localization occurs over 360 degrees in hat-shaped specimen [23]. In contrast, planar concentrated deformation will occur in the FHS specimen. As shown in Fig. 1b, FHS specimen can be divided into hat region, edge region and shear region in which forced localization develops. The designed thickness of the shear region is nearly 0.1 mm. It should be noted that the corners of the FHS samples are given a fine polish to remove defects due to machining.

2.2. Experimental setup

The schematic illustration of the experimental setup used in this study is depicted in Fig. 2. The standard SHPB bars and digital image correlation (DIC) measurement system (Kirana-05M, Specialised Imaging Ltd., www.specialised-imaging.com.cn) were utilized for dynamic loading and superficial strain field determination, respectively. An impact pulse was provided with a striker driven by a light gas gun. As shown in Fig. 2, the specimen was sandwiched between incident and transmission bars. It should be noted that the setup used in this work was made of hardened 18% nickel maraging steel bars with diameters of 14 mm and the lengths of the striker, incident and transmission bars were 0.3 m, 1.2 m, and 1.2 m, respectively. To reduce friction and specimen barreling, the bar-specimen interfaces were sufficiently lubricated using grease, and 2–3 FHS specimens were used to test forced shear properties under each loading condition. Note that the shear strain evolution is very complex due to the evolution of the width of the shear region, and this will be addressed next.

In principle, DIC technique is a representative non-contact optical method that acquire images of an object, store images in digital form and perform image analysis to extract full-field shape, deformation and/or motion measurements. It directly provides full-field displacements and strains by comparing the digital images of the specimen surface in the un-deformed (or reference) and deformed states respectively. The reader is referred to Sutton et al. [24] for further details. As a practical and effective tool for quantitative in-plane deformation measurement of a planar object surface, two-dimensional digital image correlation (2D-DIC) is now widely accepted and commonly used as a powerful and flexible tool for the surface deformation measurement in the field of experimental solid mechanics [25]. To date, it has been repeatedly used to monitor the evolution of the shear strain in real-time during the deformation
In addition, it can also be used to obtain the local strain field of materials on grain level [31]. For the sake of brevity, the outline for acquiring full-field deformations will not be described here, and the reader is referred to Yan et al. [29] for further details. However, the following specific points must be noted. To obtain a random black-and-white speckle pattern with a spatial variation in intensity, the specimen was lightly coated, which was appropriate for displacement measurement using computer vision. Then, commercial spray paints (matt white and matt black paints) were adopted for the speckle pattern on all specimens. To make sure the calculation results accurately, suitable speckle pattern applied on the specimens’ surface should better contain 4–6 pixels. The framing rate of the camera was 1,000,000 frames per second and the image resolution was 924 \times 768 pixels. In addition, the camera was placed with its optical axis normal to the specimen surface, and the distance between the camera and the specimen was ca. 250 mm. Furthermore, to acquire high enough spatial resolution to calculate the shear strain field which contains ASB, we just simply re-estimated the magnification factor in 60 pixels/mm for each test. It should be pointed out that the strain evolution was calculated using VIC-2D (Correlated Solutions Inc., correlatedsolutions.com) software.

3. Results and discussion

The forced shear tests were carried out at 20 °C by means of SHPB technique. To obtain a prescribed displacement in the plastic deformation region, a series of stopper rings were used to limit the deformation in the punching process. Different nominal shear strains were obtained by varying the thickness of the stopper ring made by high strength steel to make sure only elastic deformation occurred. To gain deep insight into the crack propagation and bring related experimental evidence, this work mainly focuses on the microstructure evolution of Ti-6Al-4V alloy after ASB formation. The punching depths (δ), maximum loadings (F_{\text{max}}), maximum shear stresses (\sigma_{\text{max}}), and velocities of the striker bar (v) are listed in Table 1. Detail information about the shear stress and displacement calculation was addressed in the Appendix A.

The loading and displacement as a function of time are presented in Fig. 3. It can be seen that the duration of shear deformation decreases with increasing punching depth when the velocity of the striker bar maintains constant (9.1 m/s), for example, samples No. 2 and No. 3. Likewise, when the punching depth maintains constant (0.3 mm), for instance, samples No. 2 and No. 4, the loading/shear stress increases with increasing velocity of the striker bar.

3.1. Shear strain measurement and analysis

To identify the process of shear strain’s evolution, the superficial shear strain is calculated by DIC technique. In the present study, a polygon area was selected as the area of interest (AOI), and the subset size of the correlation calculation was 17 \times 17 pixels with a

<table>
<thead>
<tr>
<th>No.</th>
<th>( \delta ) /mm</th>
<th>( F_{\text{max}} ) /kN</th>
<th>( \sigma_{\text{max}} ) /MPa</th>
<th>( v ) /ms(^{-1} )</th>
<th>ASB (Y/N)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.27</td>
<td>17.50</td>
<td>729</td>
<td>8.8</td>
<td>Y</td>
</tr>
<tr>
<td>2</td>
<td>0.30</td>
<td>17.66</td>
<td>736</td>
<td>9.1</td>
<td>Y</td>
</tr>
<tr>
<td>3</td>
<td>0.32</td>
<td>17.68</td>
<td>737</td>
<td>9.1</td>
<td>Y</td>
</tr>
<tr>
<td>4</td>
<td>0.30</td>
<td>17.73</td>
<td>739</td>
<td>9.2</td>
<td>Y</td>
</tr>
<tr>
<td>5</td>
<td>0.31</td>
<td>18.23</td>
<td>760</td>
<td>9.5</td>
<td>Y</td>
</tr>
</tbody>
</table>
step size of 2 pixels. For the sake of brevity, a sample deformed at 9.5 m/s was given as an example. The loading and tensorial shear strain evolution with respect to time are presented in Fig. 4. It should be noticed that the value of engineering shear strain ($\gamma_{\text{eng}}$) is twice the magnitude of $\epsilon_{\text{xy}}$ measured by DIC technique [29].

As illustrated in Fig. 4, the superficial shear strain within the shear region increases with time, while it is nearly no apparent changes in the other region. Apparently, with the development of punching depth, tensorial shear strain within the shear region continues to increase at an accelerated pace, and the severely deforming region continues to narrow down in size. It should be noted that some speckle patterns might peel off due to severe deformation, which may cause failure to calculate strain field using DIC. As might be expected, the severer of the shear deformation, the severer of the peeling of speckle patterns. Thus, shear strain measured by DIC technique is not accurate beyond $t \geq t_f = 61 \mu s$. However, the trend of strain development can be obtained through the variations of the images.

As depicted in Fig. 4, it can be seen that the strain concentration occurs close to the hole when $t = t_1 = 16 \mu s$, and the shear strain distributes homogeneously within the shear region before the maximum loading ($t = t_5 = 52 \mu s$). Hence, it can address that ASB nucleates after the maximum stress, which is consistent with the results reported by Guo et al [16]. It is interesting to note that the
maximum width of the shear concentration region is approximately 1.3 mm. With the development of punching depth, shear strain gradient occurs within the shear region before \( t = t_p = 61 \mu s \). It can be seen that though the deformation is not homogeneous, yet it is not fully localized. A crack nucleates at the corner of the hole when \( t = t_p = 61 \mu s \), and the shear strain is inaccurate on the basis of severe localized deformation. Then, the crack propagates within the shear region with the development of punching depth.

Therefore, according to the DIC results, the process of the deformation can be identified as homogeneous deformation, inhomogeneous deformation and localized deformation. In addition, on the basis of the evolution of shear strain field, the cause of decreasing load-carrying capacity for Ti-6Al-4V alloy is the initiation and propagation of crack.

### 3.2. Numerical simulations

To obtain deep insight on the distribution of shear strain and temperature within the shear region, the forced shear process was simulated using the commercial finite element (FE) software package ABAQUS/Explicit [32,33] with 3D full-size model. The physical and mechanical properties of Ti-6Al-4V alloy and SHPB bars used in this study are listed in Table 2. It should be mentioned that this simulation does not intend to exactly reproduce the forced shear behavior or to study the adiabatic shear bands itself. Fig. 5 illustrates the mesh of the FHS specimen. The elements were refined in the shear zones so that the stress/strain conditions can be modeled accurately.

The Johnson-Cook (J-C) constitutive model is widely used to describe the mechanical properties of materials because of its simple structure. The J-C constitutive model for the von Mises flow stress, \( \sigma \), can be expressed as [34]:

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

where \( A, B, \) and \( C \) represent the yield stress, hardening modulus and strain rate sensitivity, respectively. \( \varepsilon \) is the equivalent plastic strain. \( \dot{\varepsilon} \) and \( \dot{\varepsilon}_0 \) represent the current and reference strain rate, respectively. \( T, T_r, \) and \( T_m \) are the current temperature, reference temperature and melting point, respectively. \( n \) and \( m \) are strain hardening coefficient and thermal softening coefficient, respectively. In this work, \( T_r = 20 °C \). The material constants used in this work are given in Table 3.

In this work, a C3D8R (8-node linear brick, reduced integration, hourglass control) element was selected for the SHPB bars, and the SHPB bars were treated as elastic bodies. Likewise, regarding the thermal softening effect, a C3D8RT (8-node thermally coupled brick, trilinear displacement and temperature, reduced integration, hourglass control) element was applied for the specimen. The incident stress wave was applied as pressure on the surface of the incident bar and the fraction of the plastic work converted into heat was set to 0.9. The initial temperature field of the bars and the sample was set to 20 °C. The pressure-overclosure type was “Hard” contact, and the frictional forces on the interfaces were neglected.

Typical distribution of Von Mises stress, equivalent plastic strain and temperature within the shear region of Ti-6Al-4V alloy deformed at 9.5 m/s are illustrated in Fig. 6.

### Table 2

Physical and mechanical properties of Ti-6Al-4V alloy and SHPB bars.

<table>
<thead>
<tr>
<th>Materials</th>
<th>Density/kgm(^{-3})</th>
<th>Young’s modulus/GPa</th>
<th>Poisson ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>SHPB bars</td>
<td>8000</td>
<td>190</td>
<td>0.29</td>
</tr>
<tr>
<td>Ti-6Al-4V</td>
<td>4440</td>
<td>110</td>
<td>0.34</td>
</tr>
</tbody>
</table>
As depicted in Fig. 6a, it can be seen that the stress gradient of the FHS specimen is notable, and the maximum value of Von Mises stress occurs in the centre of the shear region, and the stress is quite uniform within the shear region. By contrast, no apparent plastic deformation forms in the FHS specimen except for the shear region. However, a larger plastic strain is generated at the corners of the shear region due to stress/strain concentration, as shown in Fig. 6b. This is in accordance with the shear strain measurement (see Fig. 4). In addition, this phenomenon is consistent with the FE simulation result of hat-shaped specimen reported by Zhou et al. [6]. Due to larger value of stress and strain, the temperature with the shear region is higher than that of the edge and hat regions. In

### Table 3
Materials constants of Ti-6Al-4V alloy for J-C model [35]

<table>
<thead>
<tr>
<th>A/MPa</th>
<th>B/MPa</th>
<th>C</th>
<th>n</th>
<th>m</th>
<th>c/Jkg⁻¹ °C⁻¹</th>
<th>Tm/°C</th>
<th>Thermal conductivity/Wm⁻¹ °C⁻¹</th>
</tr>
</thead>
<tbody>
<tr>
<td>724.7</td>
<td>683.1</td>
<td>0.035</td>
<td>0.47</td>
<td>1</td>
<td>526.3</td>
<td>1653</td>
<td>6.8</td>
</tr>
</tbody>
</table>

Fig. 6. Typical distribution of the (a) Von Mises stress, (b) Equivalent plastic strain and (c) Temperature of the FHS deformed at 9.5 m/s.
particular, the maximum temperature of the FHS specimen occurs at the corners of the shear region, as illustrated in Fig. 6c.

To better understand the distribution of shear strain of the FHS specimen during the forced shear process, a sample deformed at 9.5 m/s, in Fig. 7, was given as an example. Fig. 7a illustrates the evolution of the strain component within the shear region. It can be seen that LE 23 (shear strain) increases nearly linearly from 20 μs to 80 μs, and LE 22 and LE 33 increase apparently from 40 μs to 100 μs. In contrast, the other shear components almost keep around at zero. Moreover, it can be seen that the value of LE 23 is the highest among the strain components. It should be noted that the strain components within the shear zone depicted in Fig. 7a are average values. Fig. 7b shows the distribution of the shear strain (LE 23) located at the lower corner of the shear region (see Fig. 5). It should be addressed here that the centre of the shear region is defined as x-axis, and y-axis is perpendicular to the shear region. As shown in Fig. 7b, the values of shear strain increase with time due to the increasing punching depth (displacement). It is interesting to note that the values of shear strain in the edge region are almost equal to that in the hat region at the same distance. This phenomenon can be interpreted as similar plastic deformation occurs in the edge and hat regions, which is in contradiction with hat-shaped specimen [21], implying that edge effect of hat-shaped specimens can be reduced using FHS specimens.

As aforementioned, the shear strain within the shear region can be calculated accurately when t ≤ 61 μs based on the DIC technique. After that, it is not accurate due to peeling off speckle patterns caused by severe deformation. As shown in Fig. 4, the measured shear strain is about 0.2 when t = 61 μs. By contrast, the FE result (0.3, see Fig. 7a) is around 0.1 higher than that of the DIC result. This discrepancy ascribes to two main aspects. On one hand, for FE simulation, the elements within the shear region were finer than that of DIC measurements. As shown in Fig. 7b, the smaller width of the shear region, the higher of the shear strain. On the other hand, the grains within and close to the ASB are refined during the forced shear process (this will be addressed next), yet is not considered in the FE simulation. This can, in turn, lead to an increase in strength and a decrease in ductility.

As shown in Figs. 6 and 7, it can be seen that plane shear dominates the forced shear process. Therefore, it can be addressed that a shear dominated stress/strain condition can be realized using FHS specimens to investigate the dynamic behavior of materials.

To present a more intuitive comparison, the measured and simulated results of shear stress and temperature within the shear region of Ti-6Al-4V alloy deformed at 9.5 m/s is displayed using a similar scale, as shown in Fig. 8. It can be seen that the maximum simulated shear stress is about 700 MPa, which is ca 70 MPa lower than that of the test result. On the contrary, the duration of the

Fig. 7. Distribution of the (a) Strain component of the shear region and (b) shear strain at the lower corner of the shear region of FHS sample deformed at 9.5 m/s. The centre of the shear region is defined as x-axis, and y-axis is perpendicular to the shear region.

Fig. 8. Comparison between the experimental and simulated results for Ti-6Al-4V alloy deformed at 9.5 m/s.
simulated plastic deformation is approximately 80 μs longer than that of the experimental result. Because of the much longer shear deformation and a little lower shear stress, the numerical simulation will have a larger plastic work. Therefore, the maximum FE simulation temperature (124 °C) is about 60 °C higher than that of the temperature calculated based on the adiabatic assumption \( T = T_0 + \frac{\rho C V}{\rho C V} \).

The difference between the simulated and tested flow stress and temperature curves may result from the inconsistency between the constitutive model and the mechanical property of the present material. In addition, the damage (e.g., Johnson-Cook damage) of the material during the forced shear process is not considered in this FE simulations. Furthermore, the grains within and close to the ASB are refined during the forced shear process, yet is not considered in the FE simulation. Therefore, for a better description of the plastic behavior of the material, the actual stress/strain state must be considered in the tests and FE simulations. This will be further studied in our future work.

### 3.3. Microstructural characterization

The samples for microstructure observation were sectioned along the axial direction by electrical discharge machining, and metallographic specimens were prepared by standard mechanical grinding, polishing, and etched in Kroll’s reagent. Optical microscopy (OM) and transmission electron microscopy (TEM) were performed with LEICA DMI 3000M and FEI Tecnai G2-F30, respectively. It should be pointed out that TEM samples were prepared from the specimens before and after dynamic deformation. For the deformation specimen, the observation area was set to the centre of the shear region including the shear band.

Typical microstructure of the well-developed shear band of Ti-6Al-4V alloy deformed at 9.2 m/s are displayed in Fig. 9. It can be seen that the width of the shear band is about 22 μm. As shown in Fig. 9b, one of the prominent features is that elliptical voids with a crack occur at the left side of the ASB. Furthermore, the crack propagates in the centre of the ASB or nearly close to the ASB/matrix interface. With further deformation, the ASB/matrix interface will be debonded, and the surfaces of both matrixes will be covered by fractured ASB. Similar microstructural features have been found in the work of Bai et al. [9] and Lee and Lin [35]. It is interesting to note that the crack propagation is independent on the loading mode (e.g., dynamic compression [35] and dynamic torsional [9]) and geometrical feature (e.g., two-flange thin wall tube specimen [9], thick-walled cylinder specimen [18], cylinder specimen [35] and hat-shaped specimen [17]). Moreover, the phenomena (crack propagation within ASB) have also been observed in other materials deformed at high strain rates, e.g., commercially pure titanium (grade two) [23], copper [36], and ultrafine-grained titanium [19].

Combined with the previous work [9,35], the sequence of the microstructural evolution within an ASB for Ti-6Al-4V alloy can be summarized, and Fig. 10 is the schematic representation. Firstly, ASBs form due to the occurrence of severe strain concentration (Fig. 10a). Then, elliptical voids initiate at a higher strain localization point within the ASB (Fig. 10b). With further deformation, adjoining voids coalesce to bigger voids (Fig. 10c). Ultimately, a crack forms and propagates within the ASB up to failure or fracture.

![Fig. 9. Typical OM micrographs of the microstructure of Ti-6Al-4V alloy deformed at 9.2 m/s: (a) Lower magnification and (b) Higher magnification of A.](image)

![Fig. 10. Schematic diagram showing the mechanisms of dynamic failure for Ti-6Al-4V alloy: (a) ASB formation; (b) voids occurred; (c) adjoining voids coalesce; (d) crack formed and propagated within ASB.](image)
Assume here that the ASB and the matrix are “adhesive” and “adherend”, respectively. Then, the evolution of the ASB in Ti-6Al-4V alloy can be identified as “cohesive fracture” relying on the crack propagation features, a distinction that was not emphasized in previous studies. Hence, we tentatively identified “cohesive fracture” as the dynamic failure mechanism for Ti-6Al-4V alloy, which is absolutely different from the results we have found in Ti-55511 alloy [37]. It should be noted that this “failure mechanism” cannot be generalized to other metallic materials until further tests are carried out.

Typical TEM micrographs of Ti-6Al-4V alloy are shown in Fig. 11. The initial microstructure of Ti-6Al-4V alloy (undeformed) shows low dislocation density in the crystal (see Fig. 11a). The microstructures of the dynamically deformed specimen within the shear region are shown in Fig. 11b–e. Planar stacking fault and its bordering partial dislocations are displayed in Fig. 11b, which is in accord with the results reported by Guder et al. [38]. Fig. 11c is the high-resolution TEM images showing the dislocations and stacking faults corresponding to the dotted rectangle area of Fig. 11b. The initiation of cells as the dislocations from various slip systems begin to tangle and pile up are depicted in Fig. 11d. These small areas, or cells, exhibit very few or no individual dislocations and therefore are outlined by broad boundaries [39], and intense dislocations are distinguished in the junction of the grain boundaries. The feature of dislocation pile-up group is illustrated in Fig. 11e, and the dark field is inserted. It can be seen that some refine grains with low dislocation density form, and the grain boundaries are free of defects. The size of these grains is ca 150 nm, which is in accord with the results reported by Meyers et al. [23]. Hence, as deformation proceeds, the microstructure gets increasingly refined due to the occurrence of dislocations, stacking faults and cell structures.

The recrystallized grain size can be estimated as [40]:

$$\lambda = \frac{KbG}{\sigma} \tag{2}$$

where $K = 15$, $G$, $b$ and $\sigma$ are the shear modulus, burger’s vector and applied stress, respectively. For Ti-6Al-4V alloy, $G$ and $b$ are 45GPa and $2.7 \times 10^{-10} \text{ m}$, respectively. $\sigma$ is ca 1.5 GPa. Thus, the estimated recrystallized size is ca 140 nm, which agrees well with the experimental observations (see Fig. 11e).

Relying on the FE simulation evidence, the maximum temperature within the shear band is ca 120 °C. Such a temperature is much lower than those of $\alpha \rightarrow \beta$ phase transformation (approximately 1000 °C) and dynamic recrystallization (DRX) (0.4T_m, nearly 477 °C). In addition, for Ti-6Al-4V alloy, the dynamic failure energy (the area of the dynamic stress–strain curve up to the maximum stress), also called as critical strain energy, should be practically constant [41,42]. In other words, there should not be apparent difference for plastic work when deformed at various impact loading. Therefore, this observation strongly suggested that the formation of ASB does not mean phase transformation occurs. Moreover, thermal softening might not have had a pronounced effect on the onset of ASB and DRX formation. This is consistent with our previous finding in Ti-55511 alloy [21,37]. Similar findings have also been observed in commercially pure titanium (grade two) [16], AM 50 [42], Ti-6Al-4V alloy [41,42] and some steels [22].

4. Conclusions

A series of forced shear tests of Ti-6Al-4V alloy was performed by SHPB setup using FHS specimens. The sequence of the forced shear process was monitored using an ultra-high-speed camera, and the evolution of superficial shear strain was calculated by DIC technique. Moreover, the forced shear process was numerical simulated. According to the experimental and simulation findings, the following conclusions can be drawn:

Based on the superficial shear strain evolution characters, the forced shear process can be identified as homogeneous deformation,
inhomogeneous deformation and highly localized deformation stages. In addition, ASB nucleates after the maximum stress and the drastic drop in load carrying capacity was caused by the initiation and propagation of crack.

Shear dominated stress/strain state can be realized using FHS specimens to investigate the dynamic behavior of materials. Relying on the FE simulation evidence, thermal softening might not have had a pronounced effect on the onset of ASB and DRX formation.

The features of crack propagation in Ti-6Al-4V alloy under high strain rate loading is inherent, which is independent on the loading mode and geometrical feature. Therefore, “cohesive fracture” can be identified as the dynamic failure mechanism of Ti-6Al-4V alloy.

Relying on the microstructural analysis, as deformation proceeds, the microstructure gets increasingly refined due to the occurrence of dislocations, stacking faults and cell structures.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Appendix A

When one-dimensional stress-waves in the bars are achieved, and the specimen is in a state of uniform stress, the total axial loading history in the specimen can be determined by [43,44].

\[ F = E_0 A_0 \epsilon_r (t) \] (A1)

where \( E_0 \) is the Young’s Modulus of the Hopkinson bars, \( A_0 \) is the cross-section area of the bars, and \( \epsilon_r(t) \) is the strain produced by transmitted wave.

The stress in the shear region of the FHS specimen can be approximated as follows. Assuming that \( F \) is uniformly distributed, then, \( F \) can be considered as shear part \( (F_s) \) and compression part \( (F_c) \). Since the overlapping distance is about 0.1 mm, which is very small compared to the thickness of the deformed section, so the angle of the sloped shear section \( (\theta) \) is very small. Then, the components of \( F \) and the area of the shear section \( (A) \) can be expressed as:

\[ F_s = F \sin \theta \approx F \bar{y} = F \frac{(a - b)}{2h} \] (A2)

\[ F_c = F \cos \theta \approx F \] (A3)

\[ A = \sqrt{(a - b)^2 + h^2} \times b \approx h \times b \] (A4)

where \( b \) represents the thickness of the FHS specimen.

Then, the shear stress, \( \tau \), of the specimen can be approximated by:

\[ \tau \approx \frac{\pi E_0 r_0^2 \epsilon_r (t)}{2ht} \] (A5)

where \( r_0 \) is the radius of the Hopkinson bars.

The punching depth (displacement), \( \delta \), can be estimated approximately as

\[ \delta = -2C_0 \int_0^t \epsilon_r (t) dt \] (A6)

where \( C_0 \) is the elastic bar wave speed in the bar material, and \( \epsilon_r(t) \) is the reflected strain histories in the bars at the specimen ends. Here, \( C_0 = 4900 \text{ m/s} \).

References


[5] Chen G, Ren CZ, Lu LP, Ke ZH, Qin XD, Ge X. Determination of ductile damage behaviors of high strain rate compression deformation for Ti-6Al-4V alloy using