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Subtransus deformation mechanisms of TC11 titanium alloy with lamellar structure

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Abstract: Isothermal compression tests are applied to study the deformation mechanisms of TC11 titanium alloy with lamellar structure under the deformation temperature range of 890−995 °C and strain rate range of 0.01−10 s^{-1}. According to the flow stress data obtained by compression tests, the deformation activations are calculated based on kinetics analysis of high temperature deformation, which are then used for deformation mechanism analysis combined with microstructure investigation. The results show that deformation mechanisms vary with deformation conditions: at low strain rate range, the deformation mechanism is mainly dislocation slip; at low temperature and high strain rate range, twinning is the main mechanism; at high temperature and high strain rate range, the deformation is mainly controlled by diffusion of β phase.

Key words: TC11 titanium alloy; lamellar structure; deformation activation; deformation mechanism

1 Introduction

Titanium alloys, particularly two-phase titanium alloys have been widely used as advanced structural materials in aeronautic applications. Especially, the alloy with a nominal composition of Ti-6.5Al-1.5Zr-3.5Mo-0.3Si (named TC11 alloy), developed on the basis of Russia alloy VT9, is a typical two-phase titanium alloy and has been most widely used in aerospace field in China to produce the compressor discs and blades. Due to the additions of beta isomorphous element Mo, neutral element Zr and beta eutectoid elements Si into Ti–6Al base, TC11 alloy has a higher strength, better creep resistance and more excellent thermal stability, especially at elevated temperatures compared with conventional Ti-6Al-4V alloy[1−2].

Hot forming processing of conventional $\alpha+\beta$ titanium alloy including TC11 alloy often involves a series of steps from cogging of as cast ingot to final part forming with specific microstructure as the goal. Among which, subtransus processing of two-phase titanium alloys after primary cogging in the β phase region plays a key role in the conversion of lamellar to equiaxed alpha structure. Thus, it is of particular interest to study the subtransus deformation mechanisms of such alloys with lamellar structure for microstructure control. SESHACHARYULU et al and SEMIATIN et al[3−5] found dislocation glide and climb to be the main subtransus deformation mechanism of Ti-6Al-4V alloy with lamellar structure through kinetics analysis. KIM et al[6] also studied the deformation mechanisms of the Ti-6Al-4V alloy with lamellar structure using load-relaxation experiments, and the results showed that the deformation mechanisms are inner grain deformation mechanisms including dislocation glide and climb at small strains (about 0.05), and combination of inner grain deformation mechanisms and grain boundary sliding at large strains (about 1.2).

Up to now, a systematically study for the subtransus deformation mechanism of such alloys with lamellar structure is still lacking. And as for the commonly used TC11 alloy in China, early works were mainly focused on constitutive behaviors[7−9] and β processing[10−11], only a few researches were conducted on subtransus processing[12−14] and related to microstructure evolution description using hot processing map method, thus fundamental understanding of the deformation mechanisms in subtransus range is still lacking. The object of the present paper is to primarily investigate the effect of deformation conditions on deformation mechanisms of TC11 alloy during subtransus processing for better understanding of the relationship between the processing condition and microstructure evolution.

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2 Experimental

The chemical compositions of TC11 alloy used in this investigation are given in Table 1. The β transus temperature was measured to be about 1 020 °C. The received wrought bars with a diameter of 10 mm were heated to 1 040 °C, held for 30 min and followed by furnace cooling. The initial microstructure of the heat treated material is given in Fig.1. The microstructure shows a colony α structure consisting of lamellar α colonies in prior coarse β grains (average grain size of about 500 μm), a grain boundary α layer(in thickness of 5 μm) and continuous β layers between the colony α lamellae (in thickness of 3 μm), which has a typical microstructure characteristic before cogging process in α/β phase field.

Table 1 Chemical compositions of used TC11 alloy (mass fraction, %)

Al	Mo	Zr	Si	Fe
6.17	3.41	1.77	0.285	0.054
C	N	H	O	Ti
0.014	0.01	0.000 6	0.100	Bal.

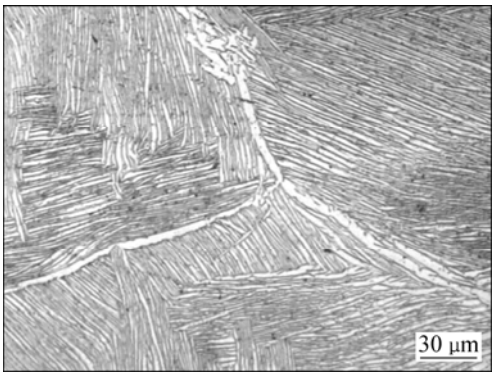

Fig.1 Initial microstructure of tested TC11 alloy

Compression specimens machined from the heat treated bars were subjected to 70% isothermal hot compression tests on a Gleeble 3500 thermal simulator in the temperature range of 890−995 °C and strain rate range of 0.01−10 s^{-1}. The specimens were heated to test temperatures at a heating rate of 5 °C /s, held for 3 min, compressed to given reduction and then water quenched to room temperature. The true stress—true strain curves at different temperatures and strain rates were obtained.

The deformed specimens were sectioned parallel to the compression axis for microstructure analysis. The samples for optical metallographic examination were made by mechanically polishing and etching with a solution consisting of 10% HF, 20% HNO_3 and 60% H_2O. Disc-shaped samples with a diameter of 3 mm were used for transmission electron microscopy (TEM) studies. These discs were mechanically thinned down to about 100 μm and then electropolished using a double jet with a 30% n-butanol and 6% perchloric acid solution in methanol at −30 °C and 30 V. TEM examinations were performed using a Tecnai 20 operating at 200 kV.

3 Results and discussion

3.1 Stress-strain behavior

The stress—strain curves of TC11 alloy with lamellar starting structure achieved by isothermal compression test under different deformation conditions are shown in Fig.2. The stress—strain curves indicate that the flow stress increases with the increase of strain rate and decrease of temperature. Similar features can be found for stress—strain relations under different deformation conditions: with increasing strain, the flow stress rapidly increases to a peak value at very low strains (around 0.05), then starts to decrease continuously, and finally reaches a steady value when the strain is over 0.7 at low strain rate (0.01 s^{-1}) or tends to reach steady flow status at higher strain rates (0.1−10 s^{-1}).

The rapid strain hardening plus continuous flow softening behavior indicated by TC11 alloy, which is a typical phenomenon observed for two-phase titanium alloys with lamellar structure[1−4], should be related to the deformation mechanisms.

3.2 Kinetics analysis

As the high temperature deformation of alloys is controlled by thermal activation, the temperature and strain rate dependence of flow stress is generally expressed in terms of Arrhenius kinetic rate equation controlled by thermal activation given by[15]:

$$\dot{\varepsilon} = A\sigma^n \exp(-\frac{Q}{RT}) \tag{1}$$

where $\dot{\varepsilon}$ is the strain rate (s^{-1}), A is a material constant, σ is the flow stress (MPa), n is the stress exponent, Q is the activation energy (kJ/mol), R is the gas constant, and T is the temperature in Kelvin (K).

In order to identify the mechanisms of hot deformation, the kinetic parameters, n and Q, are evaluated. Eq.(1) can also be written as:

$$\ln\dot{\varepsilon} = \ln A + n\ln\sigma - \frac{Q}{RT} \tag{2}$$

Thus, the following expressions can be derived:

$$\frac{1}{n} = \frac{\partial\ln\sigma}{\partial\ln\dot{\varepsilon}}\Big|_T, \quad \frac{Q}{nR} = \frac{\partial\ln\sigma}{\partial(1/T)}\Big|_{\dot{\varepsilon}} \tag{3}$$

And the stress exponent n and the activation energy Q can be defined as:

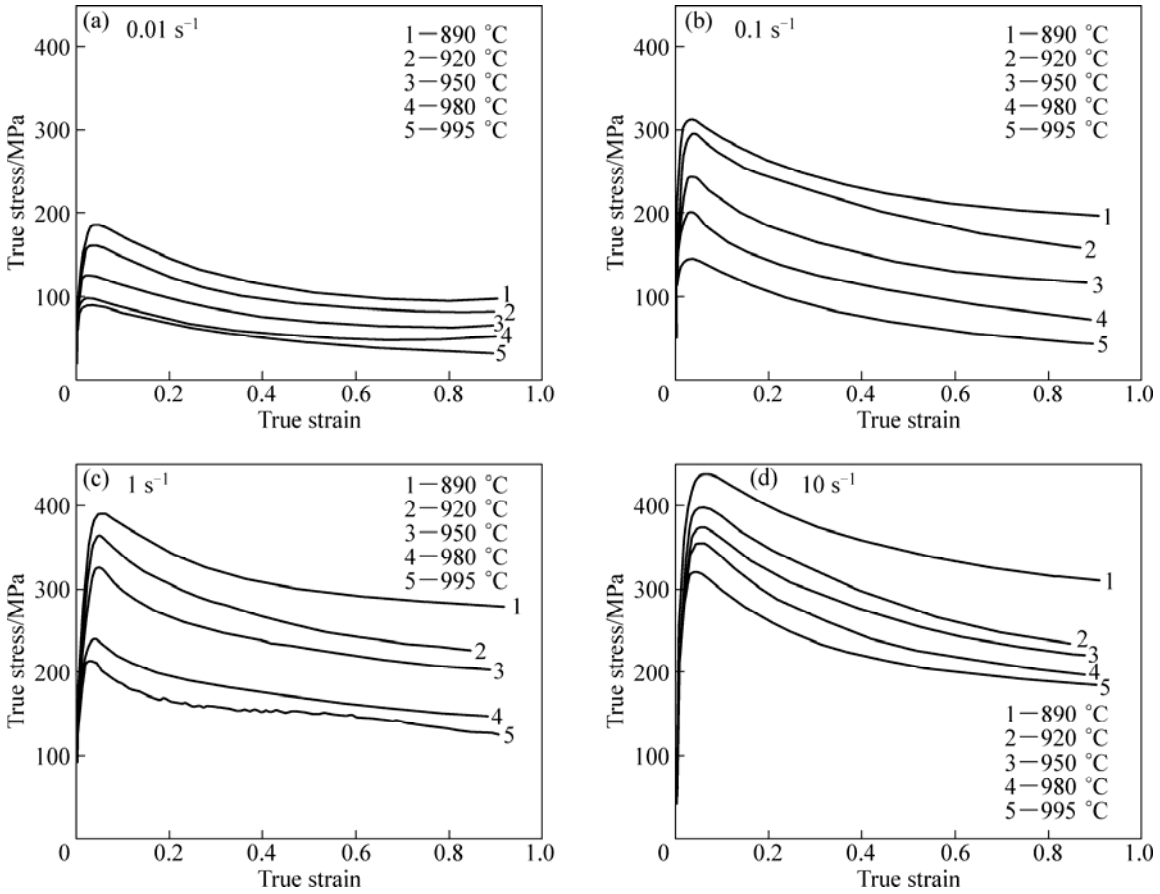

Fig.2 True stress—true strain curves at various deformation conditions during hot compression

$$n=\frac{\partial \ln \dot{\varepsilon}}{\partial \ln \sigma}\Big|_T,\ Q=R\cdot\frac{\partial \ln \sigma}{\partial(1/T)}\Big|_{\dot{\varepsilon}}\cdot\left(1/\frac{\partial \ln \sigma}{\partial \ln \dot{\varepsilon}}\Big|_T\right) \tag{4}$$

For kinetics analysis, feature stress data (peak stress or steady state stress) in the stress—strain curves are often used for flow stress σ in Eq.(1), and the peak flow stress is applied in the present work. It can be seen from Eq.(4) that if there is linear relation between $\ln\sigma$ and $1/T$ at given strain rate and between $\ln\sigma$ and $\ln\dot{\varepsilon}$ at given temperature, then the activation energy Q should be a constant value. However, from the relation of $\ln\sigma$ vs $1/T$ and $\ln\sigma$ vs $\ln\dot{\varepsilon}$ given in Fig.3 for TC11 alloy with lamellar structure during subtransus temperature, non-linear relations can be found for $\ln\sigma$ vs $1/T$ and $\ln\sigma$ vs $\ln\dot{\varepsilon}$ at different deformation conditions, which indicates different activation energy values corresponding to different deformation mechanisms at different deformation conditions.

The calculated n and Q values at different deformation conditions are given in Table 2, and some values for titanium alloys with lamellar structure reported in literatures are also given in Table 3. It can be seen that the calculated n and Q values which consider the effect of deformation conditions lie in the same range with that in references.

Table 2 Average n and Q values under different deformation conditions

Deformation condition		n	Q/(kJ·mol^{-1})
Strain rate/ s^{-1}	Temperature/°C		
0.01	890−980	4.05	351.57
	995	3.93	288.88
0.1−10	890 −980	12.19	601.12
	995	5.53	875.54

Table 3 n and Q values reported in literatures for different titanium alloys

Alloy	Deformation condition	n	Q/(kJ·mol^{-1})
Ti-6Al-4V[5]	10^{-4}–10^{-2} s^{-1}, 850−950 °C	–	455
Ti-6Al-4V[4]	10^{-3}–10^{1} s^{-1}, 815−955 °C	3.86	700
TC11[16]	10^{-3}–10^{0} s^{-1}, 800−900 °C	–	191.38
	10^{-3}–10^{0} s^{-1}, 950−1 050 °C	–	336.72
TC11[17]	10^{-3}–10^{-1} s^{-1}, 800−980 °C	4.42	490.8

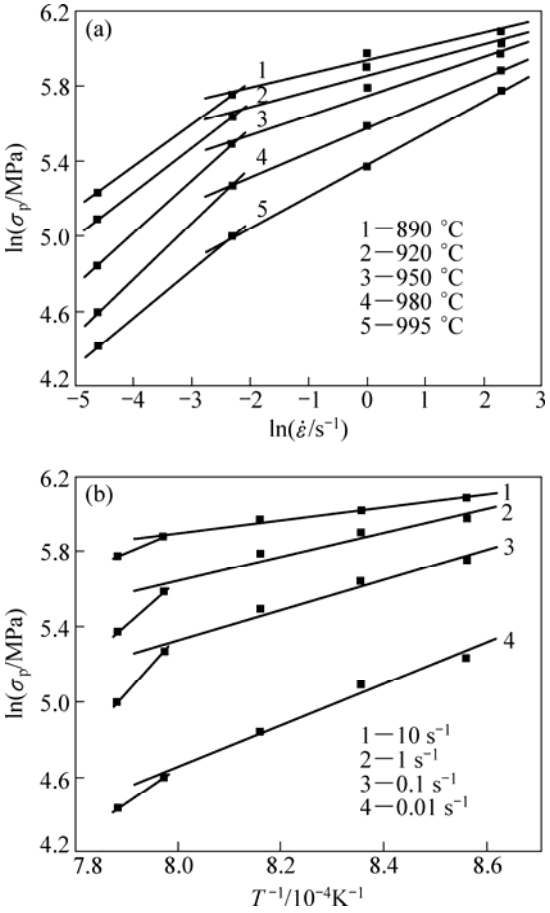

Fig.3 Curves of $\ln\sigma$ vs $\ln\dot{\varepsilon}$ and $\ln\sigma$ vs $1/T$ relations of TC11 alloy during hot deformation at different conditions

3.3 Deformation mechanisms analysis

According to the calculated activation energy values combined with microstructure observation of the deformed specimens at different deformation conditions, the deformation mechanisms of TC11 alloy with lamellar structure can be analyzed as follows.

1) At low temperature (890−980 °C) and low strain rate (0.01 s^{-1}) range, the calculated activation energy value (351.57 kJ/mol) , which is far larger than the self diffusion activation energy values of β-Ti (161 kJ/mol[16]) and α-Ti (204 kJ/mol[16]), lies between the dislocation creep activation energy values of α-Ti (200−360 kJ/mol[17]) and has a similar value with the apparent deformation activation energy of α-Ti (346 kJ/mol[16]). And the stress exponent value (4.05) is also consistent with the stress component for deformation of α-Ti. Therefore, it indicates that the deformation is mainly focused on α phase under the above conditions with the dislocation movement as the controlled mechanism[18].

Further study on the TEM observation of deformed specimen at this deformation condition range (Fig.4) indicates that the main mechanism controlling α phase deformation is dislocation glide/climb accompanied by cross slip mechanism indicated by a few bended slip lines.

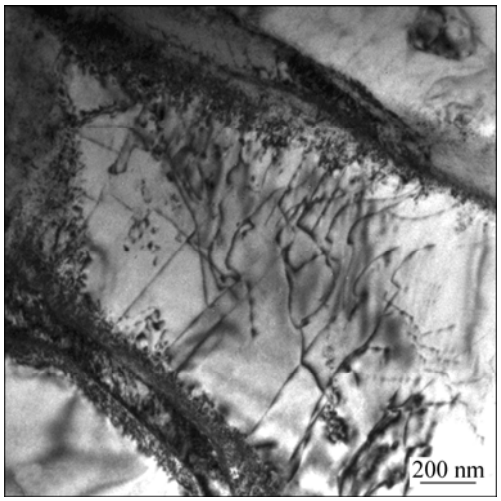

Fig.4 Dislocation morphology in α lamellar after deformation at 980 °C with strain rate of 0.01 s^{-1}

2) Under high temperature (995 °C) and low strain rate (0.01 s^{-1}) condition, the calculated activation energy value (288.88 kJ/mol) lies between the self diffusion activation energy values of α-Ti (204 kJ/mol[16]) and the apparent deformation activation energy of α-Ti (346 kJ/mol[16]), which indicates that with temperature increasing, the diffusion ability is enhanced and the dislocation can glide, climb and cross slip on both the base slip plane and non-base slip plane through self diffusion of α phase. Therefore, the main deformation mechanism changes from dislocation glide/climb at low temperature to dislocation cross slip controlled by self diffusion of α phase.

3) At low temperature (890−980 °C) and high strain rate (0.1−10 s^{-1}) range, the calculated activation energy value (601.12 kJ/mol) is not only far larger than the self diffusion activation energy values of α-Ti and β-Ti, but also far larger than the apparent deformation activation energy values of α-Ti and β-Ti (178 kJ/mol[16]), and the stress exponent value is also far larger than those during deformation controlled by diffusion and dislocation[18]. Different researchers related such high deformation activation energy to different mechanisms including phase transformation[4], the effect of stress on activation energy[5] and lamellar globularization[16−17]. TEM study of the deformed specimen (Fig.5) under such conditions shows that the high activation energy is caused by the mechanism of deformation twining.

Seen from Fig.5, integral slip across the whole thickness of α lamellar occurs (Fig.5(a)), and further observation shows lots of crossing twins inside α lamellar (Fig.5(b)) with $\{10\bar{1}0\}$ as the twining plane determined by the SAD result (Fig.5(c)). Under low temperature and high strain rate deformation condition,

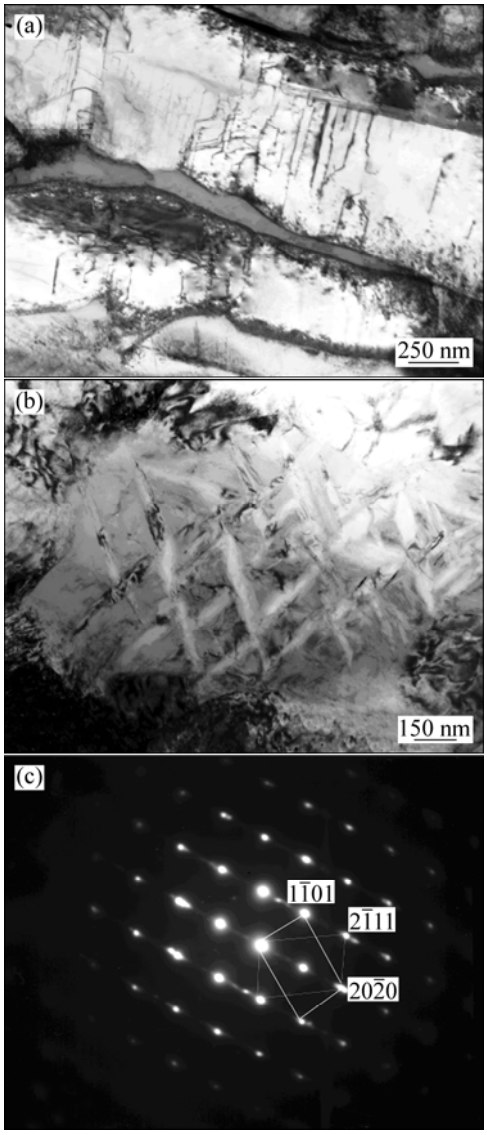

Fig.5 TEM morphologies of α lamellar after deformation at 980 °C and strain rate of 0.1 s^{-1}: (a) Slip of α lamellar; (b) Twins in lamellar; (c) Selected area diffraction (SAD) pattern of twins

the dislocation glide/climb is restricted and then serious pile-up of dislocation can cause the twining to change grain orientation for further deformation. Therefore, the main deformation mechanism is twining of α phase combined by few lamellar slip[19].

4) Under high temperature (995 °C) and high strain rate (0.1−10 s^{-1}) condition, the calculated activation energy (875−875.54 kJ/mol) is overlarge to be related to any mechanism. The microstructure observation (Fig.6) shows that the deformed microstructure changes from α+β phase lamellar to β transformed structure (which indicates single β phase at deformed temperature). Therefore, phase transformation occurs during the compression deformation due to high deformation temperature and deformation heating effect by high strain rate. According to the results by BRIOTTET et al[20], the phase transformation during deformation causes an over high activation energy, so the calculated overlarge activation energy has no physical meaning. According to the deformed microstructure, the deformation is mainly focused on transformed β phase and the mechanism should be controlled by diffusion, which is similar to the high temperature deformation of single β phase.

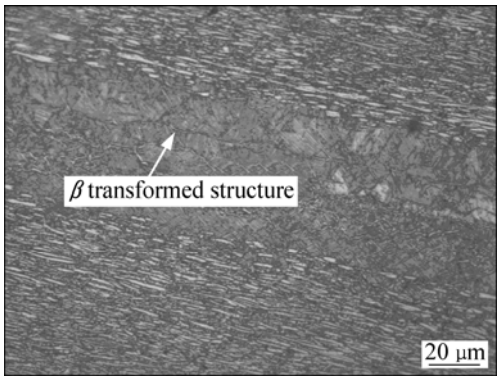

Fig.6 Optical microstructure of specimen deformed at 995 °C and strain rate of 1 s^{-1}

4 Conclusions

Under the deformation temperature range of 890−995 °C and strain rate range of 0.01−10 s^{-1}, the deformation mechanisms of TC11 titanium alloy with lamellar structure varies with the deformation conditions:

1) At low temperature (890−980 °C) and low strain rate (0.01 s^{-1}) range, the main mechanism controlling α phase deformation is dislocation glide and (or) climb accompanied by cross slip mechanism.

2) Under high temperature (995 °C) and low strain rate (0.01 s^{-1}) condition, the principal deformation mechanism changes to be dislocation cross slip controlled by self diffusion of α phase.

3) At low temperature (890~980 °C) and high strain rate (0.1−10 s^{-1}) range, twinning becomes the main deformation mechanism.

4) Under high temperature (995 °C) and high strain rate (0.1−10 s^{-1}) condition, the deformation is mainly controlled by diffusion of β phase.

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