Hot Ductility of Titanium Alloys—A Comparison with Carbon Steels—

H. G. SUZUKI and D. EYLON¹⁾

Formerly R & D, Nippon Steel Corporation. Now at Tsukuba Laboratory, National Research Institute for Metals, Sengen, Tsukuba, Ibaraki-ken, 305 Japan.

1) Graduate Materials Engineering, The University of Dayton, Dayton, OH 45469-0240, U.S.A.

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Characteristics of hot ductility of various kinds of titanium alloys were examined and compared with that of carbon steels. Although carbon steels have three typical embrittling zones above 900 K, titanium alloys do not show any embrittlement in the temperature range between β transus and 1 500 K, which is considered to be best region for hot rolling. The embrittlement occurs just below the β transus in the coarse grain titanium alloys. The ductility is poor in the coarse grain materials, but independent of the strain rate. The mechanism of the embrittlement is considered to be due to the accumulation of strain in the β phase attached to the grain boundary α in α and $\alpha + \beta$ alloys.

KEY WORDS: titanium alloy; titanium; hot ductility; embrittlement; carbon steels; high temperature tensile test.

1. Introduction

In previous papers,^{1–3)} one of the authors reported the embrittlement of carbon steels at elevated temperatures ranging from 900 K to the melting point. They were classified into three temperature ranges. The first zone is from 1 500 K to the melting temperature and the embrittlement is caused by the incipient melting of the grain boundary on heating and the decohesion of dendritic interface on cooling from the melt. The embrittlement is independent of strain rate. The impurity elements such as sulfur and phosphorus which lower the melting point are harmful. The second zone is from 1 200 to 1 500 K, where the austenitic matrix subjected to plastic deformation is dynamically restored. Fracture occurs along austenite grain boundaries where precipitates such as manganese (iron) sulfides exist. Faster strain rates manifest this embrittlement significantly. Cracking during hot rolling is mainly due to this embrittlement. The third zone is from 900 to 1 200 K. The embrittlement is manifested if the strain rate is lowered. The lower the strain rate, the worse the ductility. The major cause of this embrittlement is grain boundary sliding. In the case of carbon steels, the fcc gamma (γ) to bcc ferrite (α) transformation occurs and a proeutectoid ferrite precipitates along the austenite grain boundary. The strength level of this ferrite is almost four times lower than the matrix γ iron. Thus, strain concentrates in this region until terminating fracture. If there are some precipitates along the prior γ grain boundaries, voids nucleate easily at these sites. Therefore, sulfides, nitrides, or oxides on the grain boundary are harmful. In addition to these factors, matrix hardeners such as NbCN (carbonitrides) and solid solution hard-

ening elements make the matrix difficult to deform and result in the accumulation of strain in the grain boundary region.

In the case of titanium alloys, alloying elements lower the melting point but very limited grain boundary precipitation occurs. Then, the embrittlement in the first zone exists somehow intrinsically but the second zone embrittlement is not expected as already reported elsewhere.^{4–6)} It is interesting to study the third zone embrittlement, because titanium has a very ductile bcc phase above beta transus and transforms into an hcp phase having high strength and poor ductility in this temperature range. So, results are expected different from those of carbon steels. In this report, special emphasis was placed on the effect of thermal cycles and strain rate on the hot ductility in the β to α phase transformation region in titanium alloys.

2. Experimental Procedure

Specimens (10 mm ϕ $\times$ 120 mmL) for the hot ductility test were prepared from as cast Ti–6Al–4V(6–4) having β grain size of 2 mm in diameter, β processed Ti–6Al–2Sn–4Zr–2Mo(6–2–4–2, near α) having β grain size of 300 μ m and $\alpha + \beta$ processed 6–4 having α grain size of 10 μ m, and Ti–10V–2Fe–3Al(10–2–3, near β) with 10 μ m, respectively. The details of other alloys such as Ti–5Al–2.5Sn(5–2.5, α alloy), Ti–6Al–6V–2Sn(6–6–2, $\alpha + \beta$ alloy) and Ti–15V–3Cr–3Sn–3Al(15–3, β alloy) were mentioned in the previous paper⁶⁾. Chemical compositions are tabulated in **Table 1**.

A Gleeble tester, model 1500, which is a horizontal tensile testing machine with a resistant heating system, was used to measure hot ductility as well as strength.

Table 1. Chemical compositions of alloys. (wt%)

Heat	Al	V	Cr	Sn	Mo	Fe	O	C	N	H	
5-2.5	5.02	—	—	2.53	—	0.22	0.18	0.006	0.005	0.001	—
6-2-4-2	6.1	1.98	—	—	1.99	0.04	0.06	0.006	0.005	0.001	Zr 4.06
6-4	6.16	4.16	—	—	—	0.152	0.165	0.012	0.012	0.010	—
6-6-2	5.62	5.36	—	2.02	—	0.68	0.16	0.008	0.006	0.001	Cu 0.68
10-2-3	3.0	9.9	—	—	—	2.0	0.080	0.02	0.004	0.012	—
15-3	3.2	15.2	3.17	2.97	—	0.14	0.18	0.010	0.008	0.002	—

Fig. 1.

 Hot ductility for typical titanium alloys ($\dot{\epsilon} = 5 \text{ s}^{-1}$, Reheated mode).

Two types of heat patterns were studied. The first was heating up to the test temperature in 60s and tensile testing after holding for 60s (Reheated mode). The second was sensitizing at 1373 or 1473 K for 60s and cooling to the test temperature at the cooling rate of $20 \text{ K} \cdot \text{s}^{-1}$ and holding for 30s followed by tensile test (Sensitized mode). The true stress vs. true strain curves were measured by attaching a strain gauge during the test. The strain rates ranged from $5 \cdot 10^{-3} \text{ s}^{-1}$ to 5 s^{-1} .

Hot ductility was evaluated by reduction of area (RA) and the strength by the flow stress at $\epsilon = 0.1$ ($\sigma_{0.1}$) and maximum strength ($\sigma_{\max}$). Metallographic examinations including SEM were performed.

3. Experimental Results

3.1. Overall View of Hot Ductility and Strength of Titanium Alloys

Figure 1 shows the relationship between ductility and temperature in the reheated mode at a strain rate of 5 s^{-1} . This figure was obtained by the addition of 6-2-4-2 and 10-2-3 data to the previously reported work⁶⁾ (6-4 is as cast material). 15-3 shows very good ductility from 700 to 1600 K, but above this temperature ductility becomes poor. This ductility loss corresponds to the lowness of the melting point by about 100 K compared with 6-4. While 10-2-3 shows good ductility for a wide range of temperature, 6-2-4-2 has poor ductility below the beta transus, 1268 K.

Figure 2 shows the temperature dependence of true

Fig. 2. Temperature dependence of true stress at $\epsilon = 0.1$ for various titanium alloys ($\dot{\epsilon} = 5 \text{ s}^{-1}$, Reheated mode).

Fig. 3. The change of hot ductility and strength against temperature in the as cast Ti-6Al-4V.

stress at $\varepsilon=0.1$ with β sensitization at 1473 K and strain rate of 5 s^{-1} for representative alloys. Low stress level in the β phase region together with good ductility has a good advantage to perform hot rolling in the temperature range above β transus to 1600 K. The sharp increase in stress level is due to the increase of volume of α phase accompanied with the phase transformation from β to α in α and $\alpha+\beta$ alloys. It is also noted that the stress level of 10–2–3 is lower than 15–3 for all temperatures and is even lower than the $\alpha+\beta$ processed fine grain 6–4 below 1100 K.

3.2. Dependence of Ductility on Thermal History

Figure 3 shows the change of hot ductility and strength against temperature in the as cast 6–4. The comparison was made between the results of the reheated and the β sensitized modes and also on the effect of strain rate. It is noted that the degree of improvement in ductility by β sensitizing depends on the microstructural features. Good ductility at 1173 K for β sensitized one is due to the suppression of β to α decomposition and deformation is mainly controlled by good ductile β phase, which is proved by the low level of the strength as shown in the bottom figure of Fig. 3. The dependence of hot ductility on the thermal history has been compared among 5–2.5(α), 6–4($\alpha+\beta$), and 6–6–2($\alpha+\beta$) in the previous paper.⁶⁾ The most hardenable alloy, 6–6–2 showed the remarkable increase in hot ductility by the rapid cooling to suppress β to α transformation. A similar result was shown in the case of 6–4.⁴⁾ Even if β sensitized process is adopted, a slower cooling rate or a hold on the way of cooling to allow the transformation from β to α and the growth of grain boundary α exhibits poor ductility.

Fig. 4. Effect of strain rate on hot ductility in the $\alpha+\beta$ processed fine grain Ti-6Al-4V.

Fig. 5. Effect of strain rate on hot ductility in 6Al-2Sn-4Zr-2Mo (Reheated mode).

3.3. Strain Rate Dependence of Ductility and Strength

One example of large grain 6–4 is already shown in Fig. 3. In this case, the hot ductility improved at lower strain rates, but the data taken by other strain rates scattered to some extent and yielded no systematic trend. The results of $\alpha+\beta$ processed fine grain 6–4 are shown in Fig. 4. The ductility does not depend on the strain rate and shows superior ductility for all the strain rates tested. The strain rate dependence of 6–2–4–2 is shown in Fig. 5. There is a clear ductility trough just beneath the β transus, but there is no indication of strain rate dependence. Thus, the grain size was found being a key factor to control hot ductility. Together with the previously reported result of β alloy,⁷⁾ 15–3, it is concluded that hot ductility is independent of strain rate and poor ductility is manifested in the coarse grain materials.

Using fine grain 6–4, deformation behavior was analyzed based on the constitutional equation by Sellars and Tegart.⁸⁾ Figure 6 is the relation between true stress

Fig. 6. Temperature dependence of true stress at $\epsilon=0.1$ for various strain rates in the $\alpha+\beta$ processed fine grain Ti-6Al-4V.

Fig. 7. Relation between strain rate and temperature in the $\alpha+\beta$ processed fine grain Ti-6Al-4V.

at $\epsilon=0.1$ and temperature for various strain rates. It is noted that stress depends sharply on the temperature in the $\alpha+\beta$ region while little in the β region. Figure 7 is the relation between strain rate and temperature at

indicated stress level, which was derived from Fig. 6. From the slope of these lines, the activation energies for deformation were obtained as $268 \text{ kJ}\cdot\text{mol}^{-1}$ in the β region and $596 \text{ kJ}\cdot\text{mol}^{-1}$ in the $\alpha+\beta$ region. The activation energy in the β region is similar to the activation energy of β -Ti, and the value reported earlier,⁷⁾ but it is more than two times larger in the $\alpha+\beta$ region. Several factors are considered to be involved. One is the strain induced phase transformation which requires energy and another factor may be the energy related to the superplasticity. Fine grain 6-4 is well known for its superplasticity in this region.⁹⁾

4. Fractographic Study

The fractographic analysis of 6-4 in the embrittled region was already described.⁴⁾ Fracture occurred in the vanadium enriched β region adjacent to the grain boundary α . So, macroscopically fracture surface looks like intergranular fracture, but it consists of fine dimples due to microvoid coalescence. Good ductile, fine grain materials showed transgranular dimples. It is also true for the case of 6-6-2 and even 5-2.5. Intergranular dimple fracture was also observed in the coarse grain poor ductile β alloy, 15-3.⁷⁾

5. Discussion

There are some similarities and dissimilarities of hot ductility between carbon steels and titanium alloys. The embrittlement in the first zone is related to the melting phenomena, so to some extent it is intrinsic in nature. The alloying elements such as Fe, Co, Ni and Cu which have a partition coefficient less than 0.5 and make low melting intermetallic compounds in titanium, are embrittling agents in this zone.

The second zone does not show any embrittlement in the titanium alloys because there are no grain boundary precipitates weakening the grain boundary strength and also any alloying elements in bcc β titanium have high diffusivity¹⁰⁾ and thus dynamic recovery occurs quite easily, resulting in very little stress concentration along the grain boundaries.

The embrittlement in the third zone differs between carbon steels and titanium alloys. The most significant difference is that there is no strain rate dependence of hot ductility in the titanium alloys. Carbon steels, including austenitic stainless steels, show clear embrittlement with slower rates of straining and good ductility with rapid pulling rate. Figure 8 is a schematic drawing of the grain configuration and temperature dependence of strength around the phase transformation temperatures for carbon steels and titanium alloys. A grain boundary film is first formed when cooled from higher temperatures. Softer bcc α iron is formed in the case of carbon steels while harder hcp α titanium is formed and the remaining matrix consists of softer bcc β matrix in the case of titanium alloys. Differences of the fracture mode are also apparent. The grain boundary sliding and the intergranular ductile dimple type fracture within the film-like proeutectoid are the causes of fracture in carbon steels. Slower pulling is more favorable for

Fig. 8. Schematic drawing of the intergranular fracture in the temperature range between $0.5 T_m$ and $T\beta-50$ K.

grain boundary sliding and causes most of the strain to accumulate in this film. On the other hand, intergranular ductile dimple fracture occurs along the β region sticking to grain boundary α in the titanium alloys. Therefore, the main cause of embrittlement in titanium alloys is the difference of the strength between the matrix β and grain boundary α . Thus, plastic deformation is always restricted to this narrow β region regardless of strain rate. So, materials having larger grain size such as the as cast 6-4 shows the remarkable embrittlement.

6. Conclusion

Deformation characteristics of titanium alloys at elevated temperature were studied and compared with those of carbon steels.

(1) β titanium alloys covering near β alloy, Ti-

10V-2Fe-3Al, showed excellent ductility for wide range of temperature from 700 to 1700 K. However, melting point is lowered by about 100 K compared with Ti-6Al-4V.

(2) In the β phase region, the temperature being above β transus to 1600 K, titanium alloys have an excellent ductility. This temperature range should be utilized more for forging or rolling processes because of good ductility with low flow stress.

(3) The embrittlement occurs just below β transus in coarse grain α , near α and $\alpha + \beta$ titanium alloys due to the strain accumulation to the narrow region of β phase adjacent to the grain boundary α . The embrittlement gives rise to the cracking during hot rolling as well as hot forging in the $\alpha + \beta$ processing.

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