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Effects of hydrogen on fatigue life of Ti-4Al-2V titanium alloy①

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Abstract: Four hydrogen contents were employed to investigate the effects of hydrogen on fatigue life of Ti-4Al-2V titanium alloy by means of section-varied samples. Results reveal that the fatigue life of the materials with (116～280) × 10^{-6} hydrogen is higher than that of natural hydrogen material provided that the fatigue load $\Delta\sigma$ is over 550 MPa. At higher $\Delta\sigma$, the content of hydrogen has small effects on fatigue life within (116－280) × 10^{-6} hydrogen. For material containing 280 × 10^{-6} hydrogen, fatigue cracks tend to initiate at sample edges at higher load, in contrast, to initiate at sites of hydrides at lower load. The interstitial hydrogen atoms softening the persistent slip bands(PSB) and hydrides separating from the body become the cause of decrease in fatigue life. Hydrides resolved into the body is observed at lower $\Delta\sigma$ for material with 280 × 10^{-6} hydrogen, which is the result of concentration of hydrogen atoms at crack tips and stress-induced re-precipitation of hydrides.

Key words: titanium alloy; fatigue fracture; hydrogen; hydride
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1 INTRODUCTION

Few researches have placed on the region of hydrogen-influence on fatigue properties of titanium alloys. Because the effects of hydrogen on properties of titanium are different from those on steel, plus that titanium alloys are hydrogen storage materials, the effects of hydrogen are quite complex. Although the functions of fatigue loadings $\Delta\sigma$ amplitude for the initiation and propagation of fatigue cracks of hydrogen charged titanium alloys have not been reported, a lot of studies[1, 2] of those functions on static loadings have been made. Williams[3] proposed that at high stress intensity levels, crack growth is controlled by normal body fracture mechanism; at middle levels, controlled by transferring of hydrogen to some reactive sites through α/β interface, i. e. rate control mechanism; and, at low levels, hydrogen reacts at the sites, leading to strain-induced precipitation of hydrides, and then assists crack propagation by means of restraining plastic flow at crack tips. Yeh[4] pointed out that at high stress levels, hydrogen enhanced localizing plasticity (HELP) occurs, and at low levels, hydrides play an important role. Obviously, the rate control mechanism is relative to microstructure. It is reported[5] that mono α and mono β structure are far less sensitive to hydrogen than $\alpha+\beta$ structure, which is attributed to the diffusion coefficient of hydrogen in α phase smaller several orders than in β phase, and the quite different solubility of hydrogen in both phases.

Recent researches of us showed that similar to delayed loading cracking, effects of hydrogeo on mono α phase Ti-2Al-2.5Zr alloy depend largely on the fatigue cycle numbers and stress amplitudes. At high fatigue stress levels, fatigue life drops drastically for hydrogen charged samples; at middle and low levels, fatigue life decreases with the increase of hydride content.

For any fatigue process, persistent slipping bands(PSBs) will form due to local plastic flow, in which the dislocation density is much higher than other places in the body. When the PSBs move out of the surface, immersion and emersion appear[6]. Essmann[7] and Hunsche[8] et al regarded that interface between PSB and base is discontinuous, and dislocation movement and distribution change suddenly across the interface, so the interface becomes the favorable site for crack initiation. Ma and Laird[9, 10] obtained the direct evidence of crack initiation at such sites in mono-crystalline copper.

As dislocations will trap hydrogen atoms[11], the latter tends to accumulate into PSBs by the action of stress concentration. Studies revealed[12] that hydrogen can assist movement of screw dislocation and thus, soften the material, consequently. It can be predicted that hydrogen will also soften the PSB and promote crack initiation at PSB interface.

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The fatigue life of a material is consisted of the crack initiation life and the propagation life, and the initiation life can cover up to 80% of whole life in a smooth specimen. Therefore, it is important to investigate the effect of hydrogen on fatigue life, especially crack initiation life in titanium alloys.

2 EXPERIMENTAL

2.1 Raw material and sample

Provided by Baoji Nonferrous Metals Mill, Ti-4Al-2V alloy sheet is 2.0 mm in depth and annealed in vacuum furnace. Test sample with dimensions shown in Fig. 1 was machined by string cutting in transverse direction and gage part was polished by 1 000# sanding paper.

Samples were divided into 4 groups. Except for the first group (non-hydrogen charged), the rest 3 groups were subject to hydrogen charging to different contents. Gas hydrogen charging was employed in the treatment condition of 650 ℃/2h, cooling down to 100 ℃ in furnace. Hydrogen contents were then measured by LECO RH-2 hydrogen measurement instrument. Results are shown in Table 1.

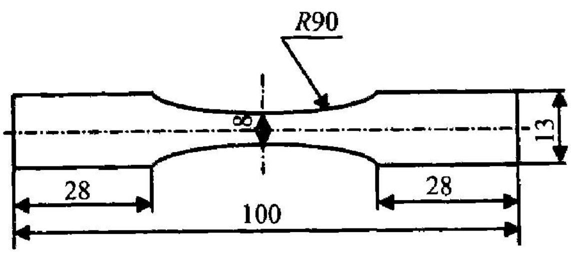

Fig. 1 Test sample dimension

Table 1 Contents of hydrogen and oxygen in each group (mass fraction, 10^{-6})

Group	Hydrogen	Oxygen
1	22	700
2	116	800
3	150	800
4	280	600

2.2 Fatigue and static test

Fatigue test was conducted on the dynamic test machine MTS810. Test was tension stress controlled with parameters of sine waveform, 5 Hz frequency, $\sigma_{min}=0$, and variable σ_{max}. Test process was controlled and automatic recorded by software of TEST-star II, and under the environment of 18 − 22 ℃ and 50% − 60% humidity. Static test was performed on static test machine MTS Sintech65/G.

2.3 Metallograph and fractograph

Microstructures and hydride graphs of samples were observed on optical microscope LEICA/MEF4A and fracture surfaces were observed on scanning electrical microscope HITACH X-650.

3 RESULTS

3.1 Tensile property

From Table 2, it can be seen that after hydrogen charging, both the tensile and yield strengths of Ti-4Al-2V rise by about 16%.

Table 2 Strength of Ti-4Al-2V with various hydrogen contents

Test group	Hydrogen content / 10^{-6}	Tensile Strength σ_b/MPa	Yield Strength $\sigma_{0.2}$/MPa
1	22(un-charged)	651	637
2	116	750	744
3	150	740	732
4	280	772	759

3.2 Stress vs cycle-number curve

Fig. 2 shows the cycle numbers at fracture (N_f) under a series of fatigue stresses $\Delta\sigma$ ($\Delta\sigma=\sigma_{max}-\sigma_{min}$, where $\sigma_{min}=0$) for 4 different hydrogen contents.

From Fig. 2, it can be found that with the drop of $\Delta\sigma$, fatigue lives of all the 4 hydrogen contents decline in index. But the decay rates are distinguishable from each other. As $\Delta\sigma>550$ MPa, un-charged (natural hydrogen content) samples have the lowest life, meanwhile hydrogen-charged samples have close lives as $\Delta\sigma>650$ MPa. When fatigue stress decreasing, lives among hydrogen-charged samples appear increasing distinction, which exhibits the tendency of decay with rise of hydrogen content. After $\Delta\sigma<550$ MPa, life of material with 280×10^{-6} hydrogen content drops even below the natural hydrogen one.

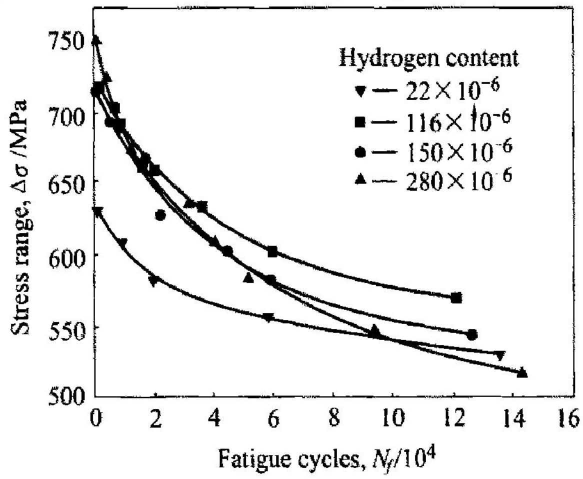

Fig. 2 N_f —$\Delta\sigma$ curves under various hydrogen contents

3.3 Metallographic structure and fractographic feature

Microstructure of Ti-4Al-2V under anneal and un-charged condition is composed of axial α phases with minor β phases. After hydrogen charging, no evidence of structure change is seen. For sample with 280×10^{-6} hydrogen, hydrides between interfaces can be found(Fig. 3), while no hydride appears in other hydrogen content samples.

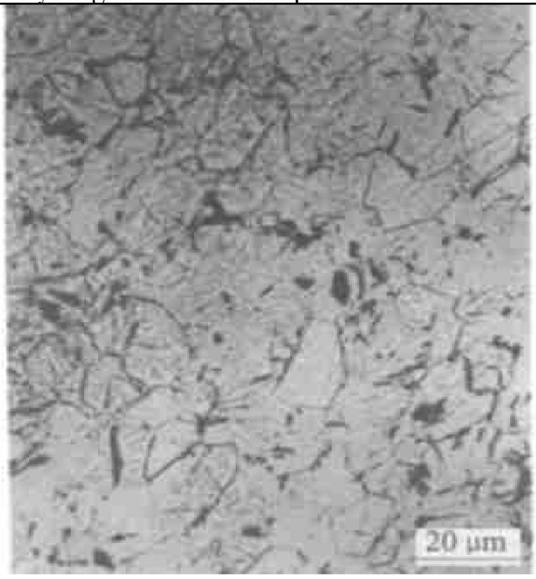

Fig. 3 Microstructure and hydride of 280×10^{-6} hydrogen-charged Ti-4Al-2V

It can be observed that the crack initiation and fracture process depend obviously upon the fatigue loadings and hydrogen contents. For the natural hydrogen samples, the crack initiates at centers in the sample and then grows towards all sides at higher loadings(near the yield strength). In this case, the fracture surface is similar to the static tensile one that dimple pattern prevails besides some visible fatigue strips(Fig. 4). With the decrease of $\Delta\sigma$, the process of crack initiation, propagation and fast fracture is visible clearly, in which the crack initiation happens mainly at the edges of smallest section. For samples containing $(116-150) \times 10^{-6}$ hydrogen, the process of fracture is close to the natural hydrogen one, besides, the edge crack initiation area exhibits quasi-cleavage pattern (Fig. 5).

However, samples containing 280×10^{-6} hydrogen appear some different characteristics. When $\Delta\sigma$ drops to some value, hydrides resolve into the matrix. From Fig. 6 where $\Delta\sigma = 661$ MPa, $N_f = 16\,044$ and $\Delta\sigma = 547$ MPa, $N_f = 93\,157$, it can be found that no hydride is visible at sample of $\Delta\sigma = 547$ MPa. Moreover, from the fracture surfaces in Fig. 7, the second cracks in sample of $\Delta\sigma = 547$ MPa are more in quantity and size than that of $\Delta\sigma = 661$ MPa.

3.4 Locations of crack initiation

Under the condition of lower fatigue loads, the

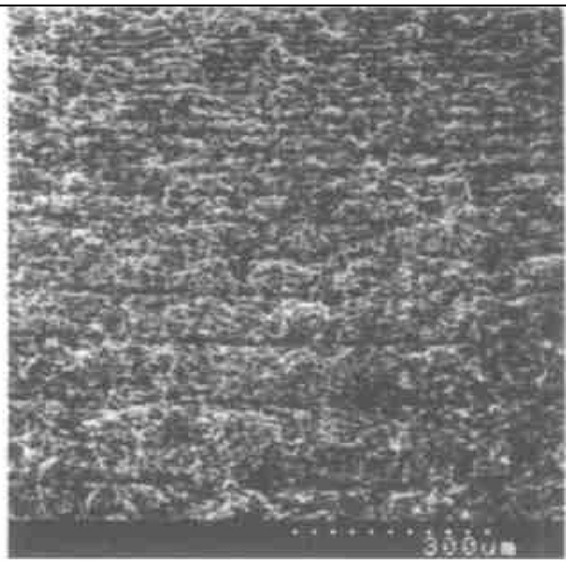

Fig. 4 Fatigue strips and dimples in natural hydrogen Ti-4Al-2V ($\Delta\sigma = 610$ MPa)

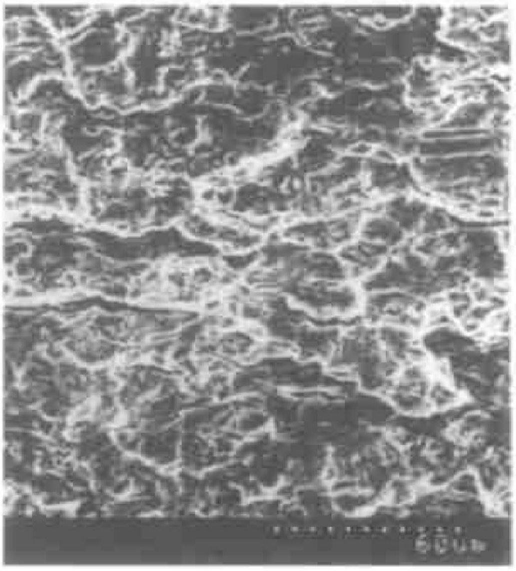

Fig. 5 Quasi-cleavage feature in 150×10^{-6} hydrogen content sample ($\Delta\sigma = 610$ MPa)

cracks initiate at three locations, i.e. at the edge, on the side and on the face(Fig. 8), which are related to the contents of hydrogen. For natural hydrogen specimen, cracks initiate at an edge principally, and additional side cracks initiation present for 116×10^{-6} and 150×10^{-6} hydrogen contents, and for 280×10^{-6} hydrogen, face cracks can even be found. Fig. 9 shows a face crack near the fracture surface exhibiting mix of larger transgranular and minor intergranular configuration.

4 DISCUSSION

Higher fatigue loadings (near yield strength) bring about macro deformation in entirety, and in this case, the yield strength determines the fatigue lives of materials with various hydrogen contents. Because of the 16% rise of strength for hydrogen-charged materials, the fatigue lives are higher than

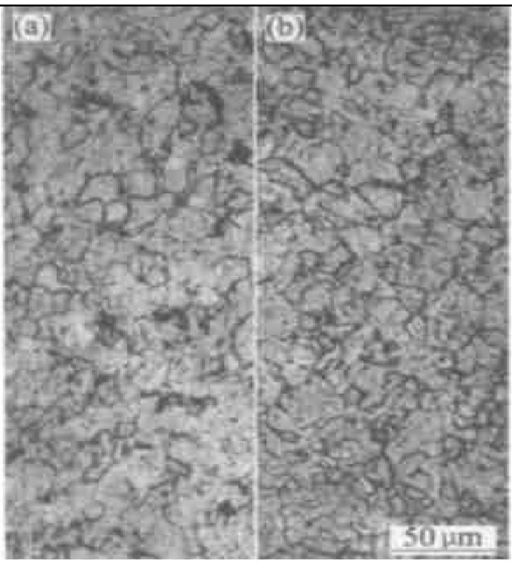

Fig. 6 Hydride resolved in lower fatigue load near fracture area with 280×10^{-6} hydrogen content
(a) $-\Delta\sigma$= 661 MPa; (b) $-\Delta\sigma$= 547 MPa

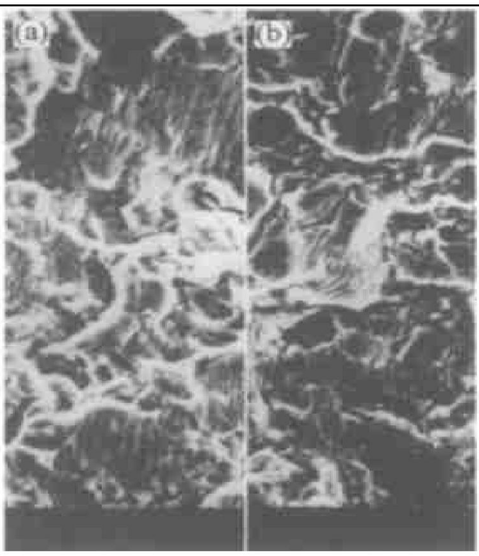

Fig. 7 Second-cracks comparison under higher and lower fatigue load
(a) $-\Delta\sigma$= 661 MPa; (b) $-\Delta\sigma$= 547 MPa

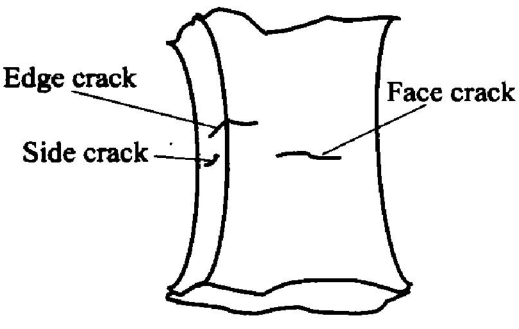

Fig. 8 Three locations of crack initiation at smallest section

natural hydrogen material.

When loadings are much lower, plastic deformation becomes localized , appearing distinct

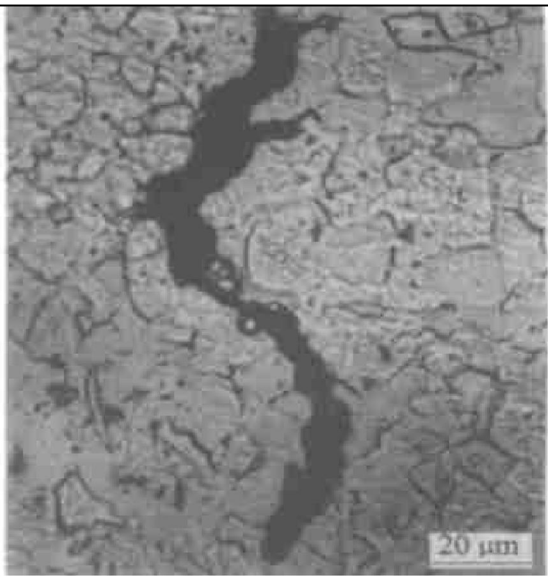

Fig. 9 Face cracks showing mixed fracture modes

processes of crack initiation, propagation and rapid cracking. At lower loadings, the fatigue crack initiation location is apparently related to hydrogen content. For natural hydrogen sample, cracks initiate at the location of high stress concentration, i. e. edges of the smallest section. With the increase of hydrogen, other locations of less stress concentration like sides and faces also become the initiation sites, which confirms that hydrogen promotes initiation of fatigue cracks.

Previous researches revealed that if no defect exists, the fatigue cracks initiate along PSB interface in a smooth sample. This is suitable for our case of edge crack initiation due to stress concentration at higher loadings. Hydrogen gathering in PSB can soften PSB, subsequently shorten the initiation life in comparison with no hydrogen charged samples. As to 280×10^{-6} hydrogen samples, besides at PSB, crack initiation has something to do with hydrides. A crack initiating at the front face of the sample in Fig. 8 shows the partial intergranular fracture feature, which represents separation of interface hydrides from body due to dislocations(within PSB) moving encountering hydrides and local stress accumulating and promoting separation, thus resulting in samples containing 280×10^{-6} hydrogen having the lowest fatigue life, and the lower the hydrogen contents, the higher the fatigue lives.

Our investigation on mono α phase alloy Ti-2Al-2. 5Zr discovered that hydrides exhibited larger effects on the fatigue properties but no hydrides resolving into body happened. For Ti-4Al-2V alloy, as a few number of β phase exist as well as far higher diffusion coefficient and saturated solubility of hydrogen in β than in α phase, hydrides at α/β interfaces may resolve into body under the action of local concentrated stress and after enough time. After that, the resolved hydrogen diffuses to the main crack tip by stress induction and hydrides re-precipitate. Phenomenon in Fig. 7 gives the fact that as hydrides in front of crack tip increase in number and size, more secondary cracks arise. The change of hydrides in the material makes the crack propagation life drop.

5 CONCLUSIONS

1) The yield strengths of Ti-4Al-2V elevate by about 16% for materials containing $(116-280)\times10^{-6}$ hydrogen in comparison with that of natural hydrogen content materials, which becomes the cause that fatigue lives of samples containing $(116-280)\times10^{-6}$ hydrogen are higher than the natural hydrogen ones under the fatigue loadings over 550 MPa.

2) At high fatigue loadings (about 650 MPa higher), the fatigue lives of Ti-4Al-2V appear little difference among hydrogen contents of $(116-280)\times10^{-6}$.

3) At lower fatigue loadings, the crack initiation location is affected by the contents of hydrogen. With the increase of hydrogen content, the locations of less stress concentration like sides and faces also become crack initiation sites, which confirms hydrogen promoting initiation of fatigue cracks.

4) Interstitial hydrogen atoms softening the persistent slip bands lead to crack initiating in advance, subsequently the decay rates of hydrogen-charged materials under high $\Delta\sigma$ are faster than that of natural hydrogen material, and meanwhile hydrides blocking PSB and separating from the matrix become the cause of decrease in fatigue life at lower loadings for higher hydrogen content.

5) For 280×10^{-6} hydrogen material, hydrides resolves into matrix at lower fatigue loadings, and then the resolved hydrogen diffuses to the main crack tip by stress induction and hydrides re-precipitate, which makes the crack propagation life drop.

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