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Micro-analyses of small cracks in 6061-T6 aluminium alloy subjected to high-cycle fatigue

Yoshimasa TAKAHASHI[1,a], Takahiro SHIKAMA[2,b] and Hiroshi NOGUCHI[3,c]

[1]Department of Mechanical Engineering, Kansai University, 3-3-35 Yamate-cho, Suita-shi, Osaka, 564-8680 Japan

[2]Kobe Steel Ltd., Aluminum & Copper Company (Chofu Works), 14-1 Chofu Minato-machi, Shimonoseki, Yamaguchi 752-0953 Japan

[3]Department of Mechanical Engineering, Kyushu University, 744 Motooka, Nishi-ku, Fukuoka-shi, Fukuoka 819-0395 Japan

[a]yoshim-t@kansai-u.ac.jp, [b]shikama.takahiro@kobelco.com, [c]nogu@mech.kyushu-u.ac.jp

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Abstract. The growth of a small crack controlling the high-cycle fatigue life of a precipitation-strengthened 6061-T6 aluminium alloy was critically investigated. As the applied stress lowered, the small crack was arrested for a long period (over 10^6 cycles) at grain boundaries before regrowth, which resulted in a significantly slow growth process. The morphological and crystallographic details of the small crack were then analyzed with focused ion beam and transmission electron microscopy. It was revealed that the small crack was formed along fine persistent slip bands (PSBs) whose structure was fairly different from that reported for other metals. The concept of PSB-limited fatigue strength may be extended to include the present material type.

Introduction

Steels are known to exhibit distinct fatigue limit at high cycle fatigue (HCF) regime; a *knee point*, where fatigue life abruptly increases from 10^5-10^6 orders to over 10^7 orders, exists in the *S*-*N* diagram [1]. Aluminium (Al) alloys, on the other hand, do not exhibit a clear knee point in the HCF regime (see e.g. Fig. 2). This still makes it essentially difficult to conduct reliable fatigue design of machine components made of Al alloys, which needs to be improved through rational approaches.

The above difference in the *S*-*N* diagram is attributed to the different growth properties of initial small cracks (typically less than ca. 1 mm in dimension): the small cracks in steels at low stress level become non-propagating ones soon after their initiation while those in Al alloys continue to grow irrespective of stress level. From a phenomenological viewpoint, such growth properties have been well characterized e.g. by surface replica observations [2]. On the other hand, the underlying micro-mechanism of small crack growth, particularly its metallographic aspects, has mostly been discarded due to the analytical difficulties stemming from the smallness of the target crack.

The aim of this study is to critically investigate the micro-mechanism of small crack growth in a typical precipitation-strengthened Al alloy (JIS 6061-T6) through precise metallographic analyses. The results are then discussed in relation to the HCF life property.

Materials and methods

The chemical composition of the 6061 alloy used in this study is listed in Table 1. The ingot was extruded at 773 K and water-quenched at 293 K. The material was then age-hardened by the T6 heat treatment (450 K×28.8 ks). The 0.2% proof strength ($\sigma_{0.2}$) and absolute strength (σ_B) were 358 MPa and 379 MPa, respectively. Fig. 1 shows the shape and dimension of the smooth round bar specimen machined from the extruded bar. The gauge portion of the specimen was electropolished to remove damage layer. The specimen was then fatigued by using a rotating bending machine (frequency: 55 Hz, stress ratio: -1). The tests were conducted in air at room temperature. Fatigue crack growth was observed by the replica method.

Table 1 Chemical composition of 6061 alloy [wt.%]

Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Al
0. 65	0.08	0.22	0.10	0.92	0. 23	0.01	0.02	Bal.

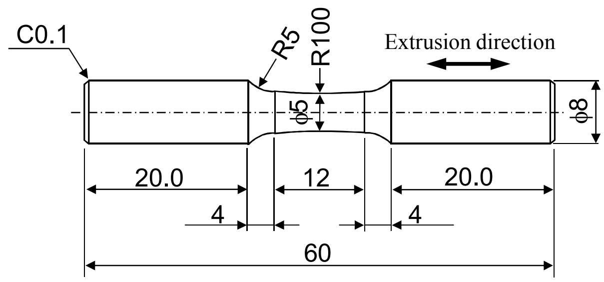

Fig. 1 Shape and dimension of specimen

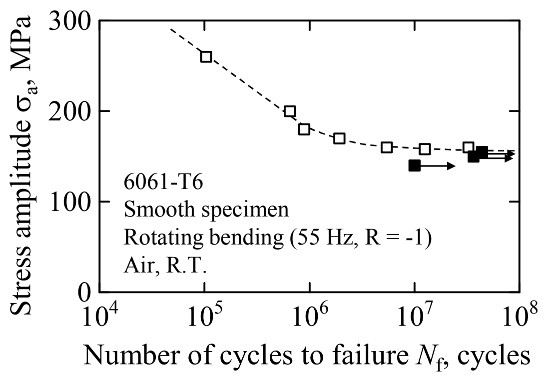

Fig. 2 *S-N* curve of 6061-T6 alloy

Results and discussion

Fig. 2 shows the obtained *S-N* curve of 6061-T6. The curve has no clear knee point around 10^5-10^6 cycles, which is in contrast to those for steels that exhibit clear knee points [1]. Fig. 3 shows the initial small crack observed in a specimen subjected to relatively low stress amplitude. The crack grew oblique to the principal stress plane. The crack tip was frequently arrested for more than 10^6 cycles and then began to grow at a relatively large rate. An electron backscatter diffraction (EBSD) analysis revealed that the crack tip was arrested at grain boundaries. Note that the regrowth of crack did not necessarily occur from the previous crack tip; a microcrack was sometimes formed near the crack tip and it coalesced with the main crack. The intermittent growth of the initial oblique crack occupied over 90% of the total life in this specimen. This explains why the slope of the *S-N* curve shown in Fig. 2 is reduced after 10^6 cycles: as the stress level lowers, the number of fatigue cycles consumed during the crack arrest increases.

Fig. 4 shows the secondary electron images of a small crack (left; plan view, right; cross section) observed by a focused ion beam (FIB). The crack extends from specimen surface and reaches only a few tens of micrometers inside; the aspect ratio of the crack (depth from surface/surface length) is less than 0.1, which is smaller than a normal semi-circular surface crack (aspect ratio: ≈ 0.5). The sub-surface crack plane is significantly tilted away from the principal stress plane, indicating that

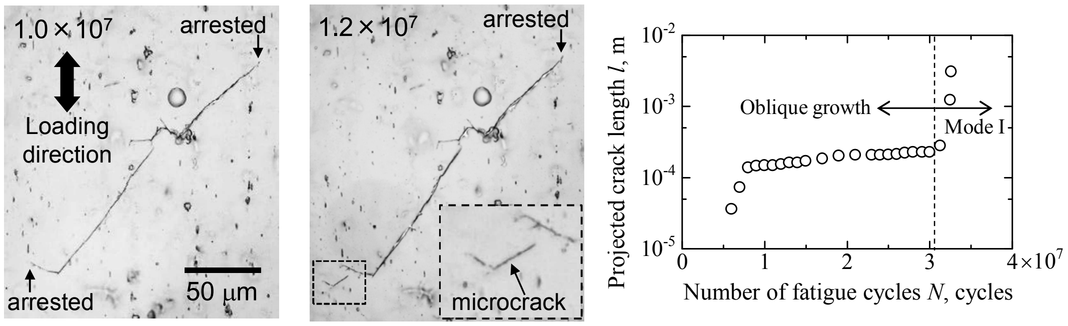

Fig. 3 Example of crack growth observed by replica and crack growth curve (σ_a = 160 MPa)

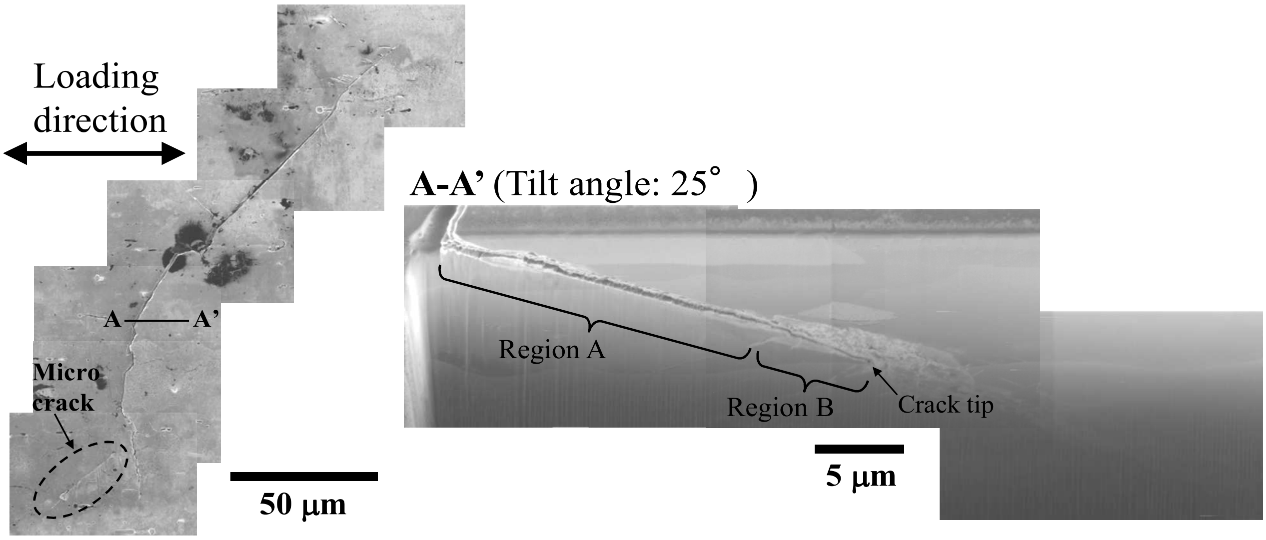

Fig. 4 A small crack observed by FIB (σ_a = 155 MPa, $N = 4.3 \times 10^7$ cycles)

the crack tip was subjected to strong shear-mode stress intensity. The crack path of Region A is rather straight and accompanies no eminent plastic zone, while that of Region B is undulated and accompanies a severely deformed plastic zone. The three-dimensionally *anisotropic* crack growth is probably caused by the different growth resistance between Regions A and B.

Fig. 5 shows the TEM image of crack tip region located near the specimen surface. The beam direction was set parallel to the <110> direction of G1 so that the {111} planes in G1 are edge-on to the printed surface. The crack grows within a significantly localized slip band structure (i.e. persistent slip band; PSB) whose width is not more than 200-300 nm. The fine PSB and the associated crack are very different from the reported ones in pure FCC metals that accompany the formation of dislocation walls (e.g. vein, labyrinth or ladder structure [3]). The characteristic appearance of the fine PSB/cracking probably stems from the *inhomogeneity* (i.e. precipitation-reinforced matrix structure) of the present material type; once the precipitates on a certain slip plane are sheared, further slip along the same plane is facilitated [4].

The results suggest that the observed small crack growth in HCF regime can essentially be regarded as the initiation and extension of PSB cracking that exclusively occur on specimen surface. This process, the unit of which is grain or a group of sub-grains, is entirely crystallographic; it is easily impeded by grain boundaries, and the subsequent process depends primarily on the position of active dislocation sources in the adjacent grain. The micro-mechanism shown here is believed to be applicable to other precipitation-strengthened alloys such as SUH660 stainless steel [5] in which similar small crack growth behavior is reported.

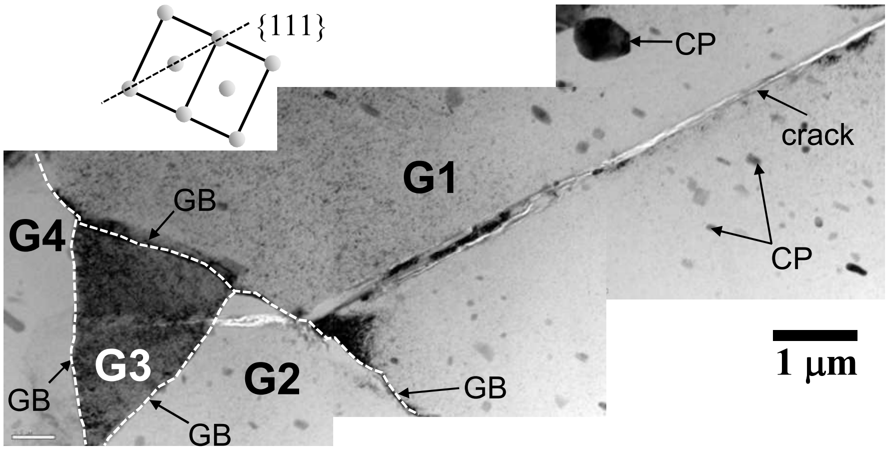

Fig. 5 TEM image of crack wake corresponding to Region A in Fig. 4 (GB; grain boundary, CP; coarse precipitate). Note that the fine precipitates reinforcing the matrix are not resolved in this image.

Finally, it should be added that the PSB-limited fatigue strength concept [6] still appears to be hold for the present material at very high-cycle fatigue (VHCF) regime of over 10^9 cycles according to our recent investigation using ultrasonic fatigue tester (see Fig. 6). The results will be further compared with the present data and discussed in detail elsewhere.

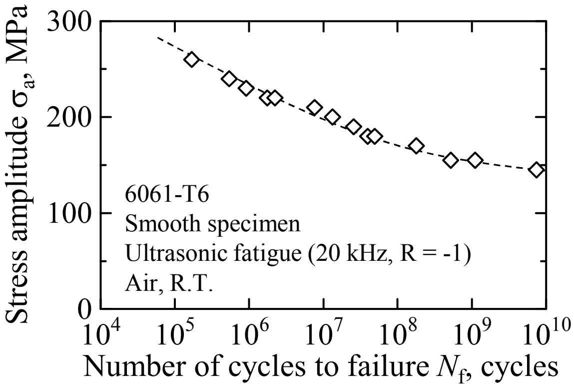

Fig. 6 S-N curve of 6061-T6 alloy including VHCF regime

Summary

The initial small fatigue crack in 6061-T6 alloy was critically analyzed by FIB and TEM. It was revealed that the small crack growth was attributed to the initiation and extension of PSBs whose structure was very different from those in pure FCC metals. The concept of PSB-limited fatigue strength may generally apply to the same material type (i.e. precipitation-strengthened alloys).

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