Study on Ultrahigh Cycle Fatigue Performance of GH4169 Nickel-Based Alloy at 650 ℃

SONG Zongxian[1,2], *QI Hao*[3], *WU Zhisheng*[1*], *CHEN Xin*[2], *ZHAO Zhenjun*[2], *WU Ruizhe*[2], *LIAN Zheng*[2]

1. School of Materials Science and Engineering, Taiyuan University of Science and Technology, Taiyuan 030024, P.R.China;
2. School of Aeronautics and Astronautics, Tianjin Sino-German University of Applied Sciences, Tianjin 300350, P.R.China;
3. State Key Laboratory of Mechanics and Control of Mechanical Structure, Nanjing University of Aeronautics and Astronautics, Nanjing 210016, P.R.China

(Received 10 August 2020; revised 10 September 2020; accepted 15 October 2020)

Abstract: The fatigue test between 10^5—10^9 cycles of GH4169 nickel-based superalloy commonly used in aircraft engines is carried out by ultrasonic fatigue machine at 650 ℃. The S-N curve is obtained and the fatigue fracture morphology is observed. The fatigue S-N curve presents a "step-like" shape, with the first inflection point near 1×10^7 cycles and the second inflection point near 1×10^8 cycles. There is no engineering fatigue limit, and it still shows a downward trend after 10^7 or even 10^9 cycles. The crack initiation location is related to its life. Cracks are generated on the surface below 10^7 cycles, while it is inside above 10^7 cycles. The crack initiation source in the ultra-high cycle fatigue at 650 ℃ is mainly the local intergranular fracture and casting defect of the matrix. In the phase of crack propagation, the mixed propagation of intergranular and cleavage is the main form.

Key words: nickel-based superalloy; ultrahigh cycle fatigue; crack initiation

CLC number: TG146 **Document code:** A **Article ID:** 1005-1120(2020)06-0970-09

0 Introduction

Superalloy, also known as thermal strength alloy or heat resistant alloy, is a kind of aviation metal material developed in the 1940s. It can withstand complex stress under the conditions of 600—1 100 ℃ oxidation and gas corrosion, and can work reliably in the superalloy for a long time. Nickel-based alloy is the fastest developing and most widely used[1]. Domestic nickel base alloy GH4169 is mainly used in some important parts of aeroengine. When aircraft engine in service process, the change of working condition is extremely intense, vibration is the main reason for the failure of the parts. The pneumatic vibration frequency is very wide, and it may be from a few hertz to several thousand hertz frequency range. High frequency resonance often occurs in blades, which is prone to fatigue fracture in the end. Especially, the fatigue performance in the ultra-high cycle fatigue stage has a great impact on the service life of engines. With the continuous improvement of aircraft performance, the performance requirements of aviation materials are more stringent. "The U. S. Air Force engine Structural Integrity Program" has increased the fatigue life of engines from 10^7 to 10^9 cycles[2].

A large number of achievements have been made in the study of fatigue of GH4169 alloys[3-14], but the fatigue behavior and fracture mechanism of alloys at service temperature of 10^7 cycles or more still need to be further studied. Many studies have shown that the fracture behavior and mechanism of metal materials after their fatigue life enters into the ultrahigh cycle range were different from those of

*Corresponding author, E-mail address: zswu@tyust.edu.cn.

How to cite this article: SONG Zongxian, QI Hao, WU Zhisheng, et al. Study on ultrahigh cycle fatigue performance of GH4169 nickel-based alloy at 650 ℃[J]. Transactions of Nanjing University of Aeronautics and Astronautics, 2020, 37(6): 970-978.
http://dx.doi.org/10.16356/j.1005-1120.2020.06.015

low cycle and high cycle fatigue[15-18]. Therefore, in this paper, 10^5—10^9 cycles of fatigue tests are conducted on nickel-base alloy at 650 ℃ to study its fatigue characteristics, observe the fracture morphology and analyze its fracture mechanism, so as to provide theoretical basis for further application of the alloy in aero-engine high-temperature service field.

1 Materials and Test Methods

The raw material was fabricated by vacuum induction and vacuum consumptive double smelting process. After the forging was formed, standard heat treatment was performed (980 ℃/1 h, air cooling; 720 ℃/8 h, furnace cooling to 620 ℃/8 h, air cooling). The chemical composition of the alloy (in mass): 53.00%Ni, 5.30%Nb, 3.00%Mo, 1.00%Ti, 0.50%Al, 19.00%Cr, 0.05%C, other Fe.

Tensile tests were conducted on Instron 5982 electronic universal material testing machine according to GB/T4338—2006 and the test temperature was 650 ℃. Strain control was used to control the loading speed of the test, which was 0.002/min before yield and 0.02/min after yield. After reaching 650 ℃, heat preservation for 10 min before stretching. The three specimens were tested and the average value was finally taken. Fig.1 shows the size of the specimen of the high temperature tensile equipment and the specimen after fracture, and it can be seen that there are obvious traces of oxidation on the surface of the specimen.

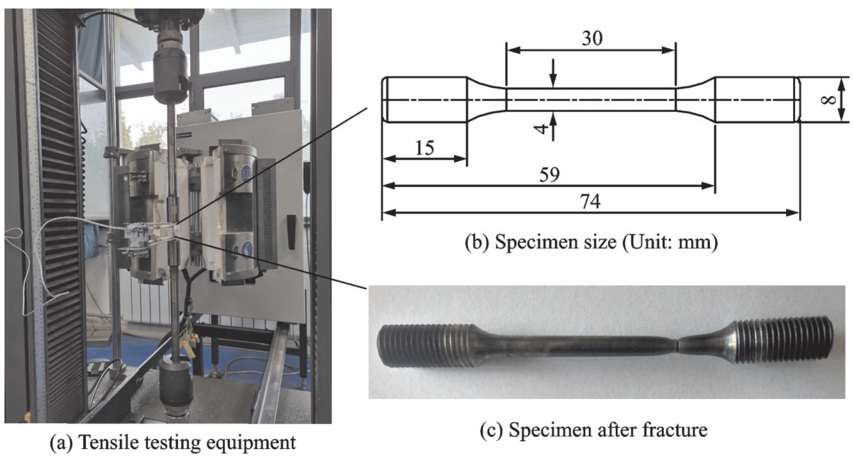


Fig.1　Tensile testing equipment and specimens at high temperature

The fatigue test adopted USF-300 ultrasonic fatigue testing machine with fatigue loading frequency of 20 kHz and symmetrical tension and compression cyclic load. The equipment was composed of ultrasonic fatigue power supply and vibration terminal. The terminal of the testing machine included transducer and lug rod. The specimen was connected with the lug rod by bolts (Fig.2). The specimen shape was designed as a dog bone shape with variable cross section in the middle(Fig.3). The middle part of the specimen was polished. The specimen was heated by induction coil and the temperature in the middle of the specimen was monitored by a thermometer. The test terminated automatically when the fatigue crack or other fatigue damage had grown enough to cause a 5% change in resonance frequency. The equipment would stop automatically when the cycle set by the system reached 1×10^9. After fatigue test, the fracture was cleaned by ultrasonic cleaning equipment. The morphology and energy spectrum of the fracture were analyzed by SEM.

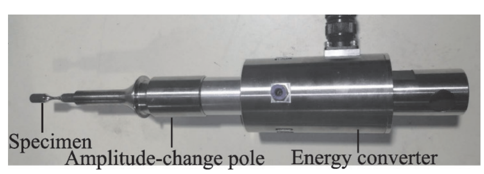


Fig.2　Terminal components of ultrasonic fatigue testing machine

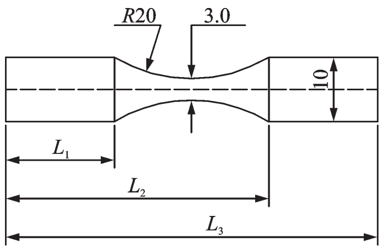


Fig.3 Dimensions of ultrasonic fatigue specimen at high temperature (L_1=15 mm, L_2=39 mm, L_3=54 mm)

2 Experimental Results and Discussion

2.1 Tensile property

The tensile properties of the three GH4169 alloy specimens at 650 ℃ are shown in Table 1. Fig.4 shows the stress-strain curve of 1# specimen and Young's modulus is obtained by fitting the elastic state. As can be seen from Table 1, GH4169 alloy has excellent high temperature tensile properties of yield strength of 937 MPa, ultimate tensile strength of 1 166 MPa, elongation of 22.75%, and elastic modulus of 167 GPa, which are far beyond the requirements of high temperature mechanical properties stipulated in the aviation materials manual[1].

Table 1 Tensile properties of GH4169 alloy at 650 ℃

Specimen	Yield strength/MPa	Ultimate tensile strength/MPa	Elongation/%	Young's modulus/GPa
1#	938.8	1 157.7	24.56	160.1
2#	941.8	1 170.5	23.60	169.3
3#	930.5	1 170.4	20.08	171.0
Mean value	937.03	1 166.2	22.75	166.8

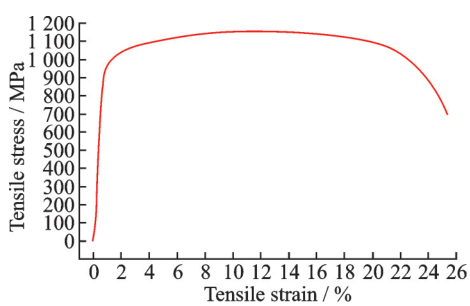


Fig.4 Stress-strain tensile curve of 1# specimen at 650 ℃

2.2 Fatigue S-N curve

The S-N curve of GH4169 alloy is shown in Fig.5. The arrow in the figure indicates that the test was stopped without failure after reaching 1×10^9 cycles. It can be seen from the figure that the S-N curve presents a "step-like" shape, appearing at the first inflection point near 1×10^7 cycles and the second inflection point near 1×10^8 cycles. The S-N curve of ultra-high cycle fatigue of alloy materials usually has a "stepped" or "double linear" shape with two inflection points[19], which is a typical feature different from low cycle fatigue and high cycle fatigue.

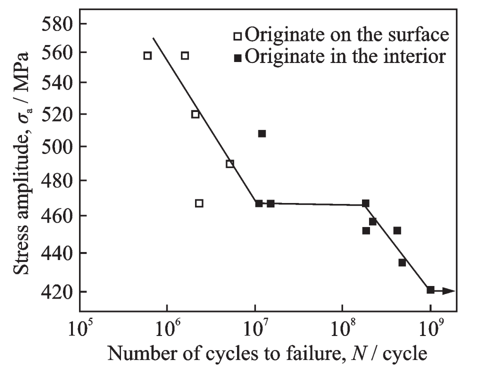


Fig.5 Fatigue S-N curve of GH4169 alloy at 650 ℃

Through the curve, it can also be found that the crack initiation location is related to the cycle. Before the first inflection point, that is, the specimen in the high cycle fatigue interval between 10^5 and 10^7, the crack initiation occurs on the surface. Cracks were generated inside the specimens with ultra-high cycle fatigue life interval after the first inflection point. The research shows that the mechanism of fatigue crack source transfer is related to the competition between surface damage and internal damage of specimens, but there are different views on the explanation of this phenomenon. Shanyavskiy. et al.[20] believed that the internal stress was released due to stress concentration under high stress, and the surface crack played a major role at this time. The stress concentration in the material under low stress is the main factor of initiation. Some scholars believed that the internal grain was more likely to produce stress concentration under low stress resulting in plastic deformation, so the crack tended to be generated in the internal[21]. Kawagoishi et al.[22] argued that the surface oxide film inhibited the initiation of cracks, thus facilitating the initiation of internal fatigue cracks.

2.3 Fracture analysis

Fig.6 shows the fatigue fracture morphology with a stress amplitude of 558 MPa and a life of 1.6×10^6 cycles. From the macro fracture in Fig. 6(a), it can be seen that the crack originated on the surface and had two crack sources (c and d). Due to the high stress amplitude (558 MPa), with the increase of cycles, the phenomenon of extrusion furrows and extrusion ridges were formed between the residing slip bands, and they would be coupled with the oxides on the surface of the inclusions, resulting in the initiation of fatigue cracks on the surface, and it was easy to generate multiple crack sources. When the stress amplitude was low, only the optimally oriented slip system slips along a specific plane, and finally a single crack source was formed[23-25].

Fig.6(b) is an enlarged view of the crack source of "c" in Fig. 6(a). It can be seen that the fracture surface of the fatigue specimen is flat and there are clear stream-like stripes extending to the center of the specimen at the crack initiation point. According to the morphology, the fracture is divided into three regions: Ⅰ is the crack source region. Ⅱ is the crack initiation area, which has large fluctuation and relatively rough section. Ⅲ is a stable extension area with a flat section.

Fig.7 shows the fatigue fracture morphology with a stress amplitude of 508 MPa and a life of 1.2×10^7 cycles. The specimen life has entered the ultra-high cycle stage and the crack source has transferred from the surface to the subsurface. There is a competitive relationship between crack source initiation and surface initiation. For internal initiation, there is also a competition between the inclusion and the matrix. Therefore, the competition for the initiation location of ultrahigh cycle fatigue cracks in the metal is a key problem that needs to be further studied[26]. When the stress amplitude is reduced to a certain extent and the fatigue load is lower than the limit of resident slip zone, although each cycle will cause irreversible strain, it is not enough to open the cracking of resident slip zone on the surface. When the load is reduced to a certain extent, the possibility of surface crack initiation will be reduced[27].

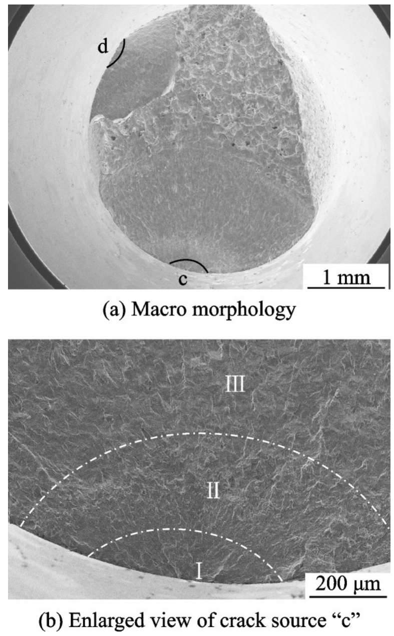
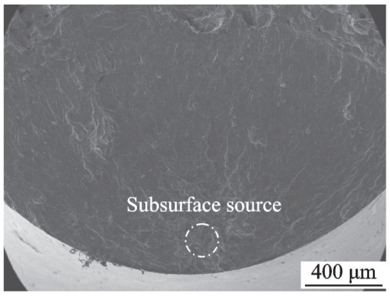
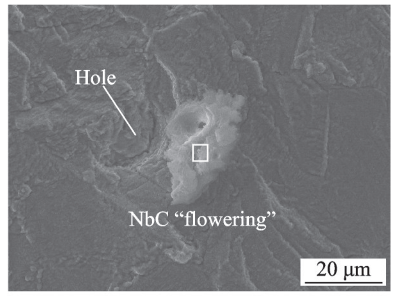


Fig.6 Fractography arising from surface with stress amplitude of 558 MPa and life of 1.6×10^6 cycles

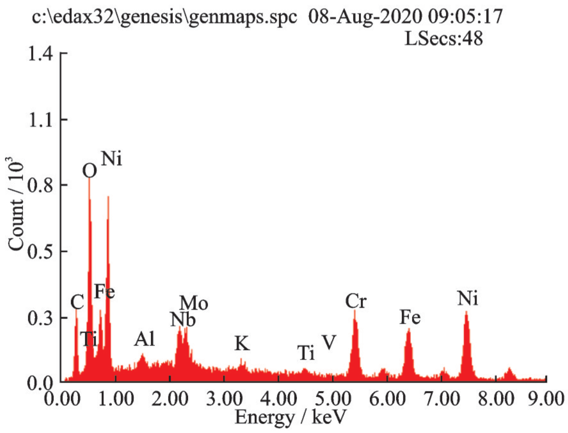
It can be seen from Fig.7(b) that the NbC phase in the extension area has been greatly expanded in volume and seriously cracked itself, leading to the appearance of "flowering" phenomenon. Energy spectrum analysis of the red marked area shows that the oxygen content is very high. The carbides can even fall off, leaving holes where they used to be. Hou and Lyu et al.[28-29] also found the same phenomenon in the study on high temperature fatigue of GH4169 alloy. When these carbides are subjected to cyclic loading under high temperature environment, the deformation cannot be coordinated because the elastic modulus and thermal expansion coefficient are different from that of the matrix, and the dislocation movement will result in blocking and plugging. This will result in stress concentration between carbide and matrix and lead to secondary cracks. Therefore, in order to improve the fatigue strength of the alloy, the size and number of particles of the second phase should be controlled in the matrix[30-31].



(a) Crack source



(b) Carbide



Element	Weight fraction/%	Atom fraction/%
CK	11.09	29.07
OK	15.81	31.10
AlK	00.93	01.08
NbL	04.34	01.47
MoL	03.18	01.04
KK	00.76	00.61
TiK	00.84	00.55
VK	00.09	00.06
CrK	12.13	07.34
FeK	14.73	08.30
NiK	36.11	19.36
Matrix	Correction	ZAF

(c) Energy spectrum analysis of carbide in Fig.7(b)

Fig.7 Fractography arising from inside with stress amplitude of 508 MPa and life of 1.2×10^7 cycles

Fig.8 shows the fatigue fracture morphology with a stress amplitude of 452 MPa and a life of 4.2×10^8 cycles. It can be seen that the crack initiation is in the interior and the whole growth area is relatively flat. A similar "fish-eye" characteristic morphology is found. Because the source is close to the surface, "fish-eye" features are incomplete. Fig.8(b) is the corresponding complete "fish-eye" feature diagram. Ⅰ (Fig.8(c)) is the source of the crack. It is the intergranular fracture of the matrix and located in the center of the "fish-eye". Ⅱ (Fig.8 (d)) is the crack initiation area (FGA area). This area has a rough cross-section with large fluctuation and is bright under scanning electron microscope. Therefore, it is also known as the granular bright area (GBF area). Ⅲ (Fig.8(e)) is the initial extended area (FiE area) with "fish-eye" characteristics. The section of this area is relatively flat. According to measurement by Liu et al.[32], the roughness of FGA region in "fish-eye" feature area was one order of magnitude greater than that of FiE region. The region in Fig.8(a) is the steady-state crack propagation region outside the "fish-eye". In the ultrahigh cycle fatigue failure of metals, "fish-eye" characteristic morphologies generated by nonmetal and porosity defects as the source are relatively common, while "fish-eye" characteristic morphologies generated by intergranular fracture as the source are typical characteristics of ultrahigh cycle fatigue at high temperature.

Fig.8(c) shows the morphology of the fracture source region, and it can be seen that the fracture is basically intergranular, that is, the main crack is generated intergranularly from the beginning. The surface of the fracture is covered with a layer of granular material and uneven, which is actually the oxidation product formed after severe oxidation. Meanwhile, secondary cracks along the grain boundary have also been observed. Fig.8(d) shows the morphology of the crack initiation zone at 100 μm away from the crack source, and it can be seen that the crack initiation zone is a mixed propagation form of intergranular and quasi-cleavage. Fig.8(e) shows the initial growth zone morphology at a distance of 300 μm from the crack source. It can be seen that the section is relatively flat with a large number of stream-like stripes along the crack growth direction, which is a typical cleavage fracture feature. Observation of the instantaneous fault zone of the crack (Fig.8(f)) showed that there were a large number of small dimples and some large holes with the size of 10—20 μm formed by NbC phase, indicating that the instantaneous fault zone was ductile.

The crack initiation and propagation along the grain at the beginning is mainly due to the fact that the grain boundary weakening at high temperature is more serious than the ingrain weakening

Osinkolu et al. conducted a 650 ℃ fatigue test on single-notch fatigue specimens and also found that GH4169 alloy expanded intergranularly in the air and mainly transgranularly in the vacuum. Therefore, it can be proved that oxygen plays a crucial role in grain boundary weakening at high temperature[28,33].

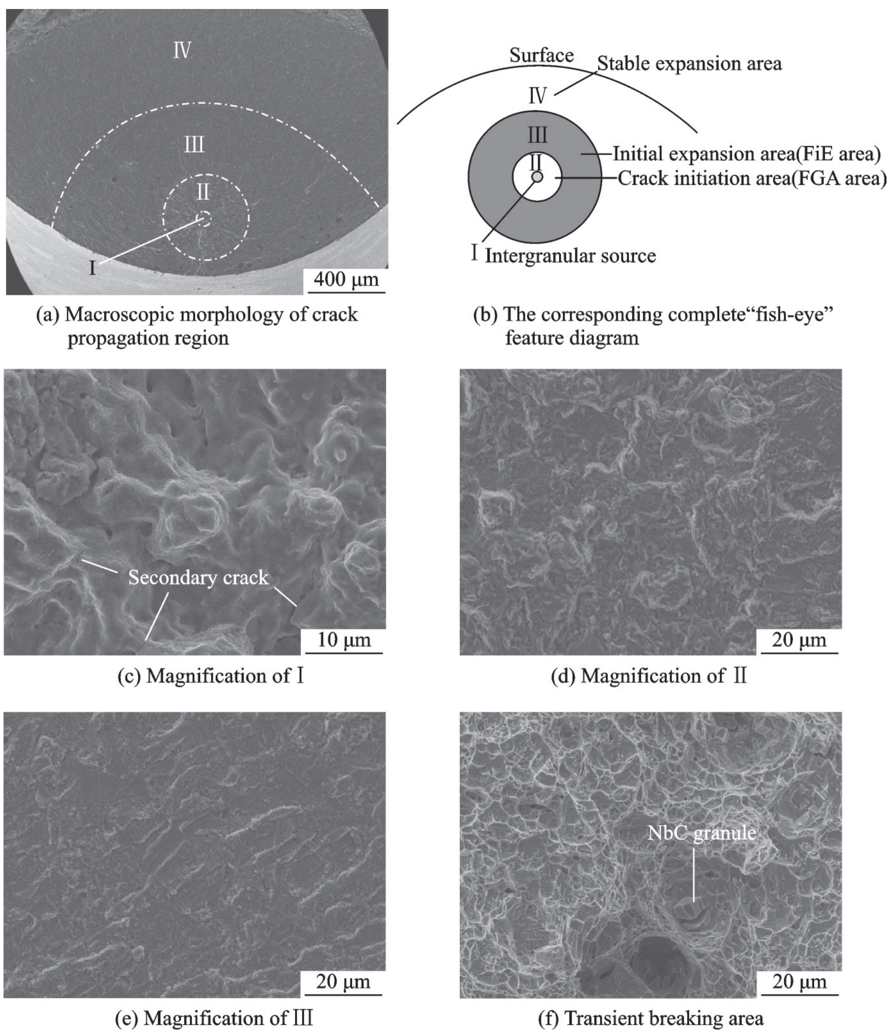
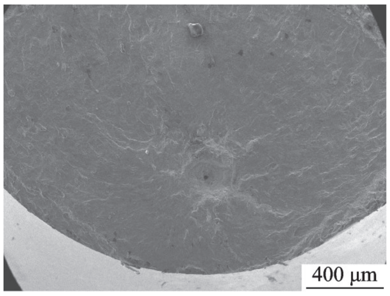
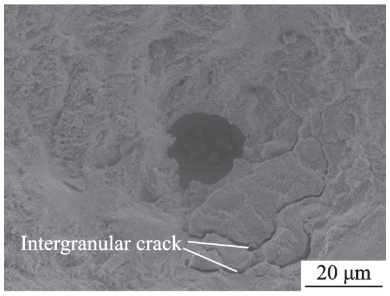


Fig.8 Fractography arising from inside with stress amplitude of 452 MPa and life of 4.2×10^{8} cycles

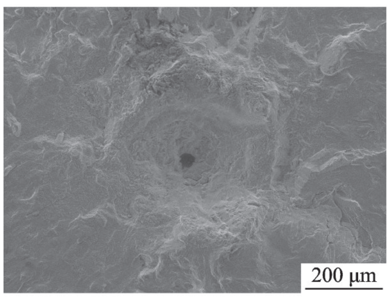
Fig.9 shows the fatigue fracture morphology with a stress amplitude of 452 MPa and a life of 1.85×10^{8} cycles. The macro morphology in Fig.9(a) shows that the crack originated from the inside and originated from the casting hole defect with a diameter of about 20 μm (Fig.9(b)). In the process of fatigue cyclic loading, the holes produced stress concentration and formed pits with a diameter of about 200 μm along the crystal growth as the crack initiation area (Fig.9(c)), and there were many secondary cracks along the crystal in the pits. Due to the large area of the crack initiation area, the characteristics of the initial expansion area are not obvious, so there is no "fish-eye" formation feature on the macro fracture morphology.



(a) Macroscopic fracture morphology



(b) Hole defects



(c) Area near the crack source

Fig.9 Fractography arising from inside with stress amplitude of 452 MPa and life of 1.85×10^{8} cycles

3 Conclusions

(1) Ultrasonic fatigue test was carried out on GH4169 alloy commonly used in aircraft engines at 650 ℃. The fatigue S-N curve was obtained. The fatigue S-N curve presents a "step-like" shape, with the first inflection point near 1×10^{7} cycles and the second inflection point near 1×10^{8} cycles. There is no engineering fatigue limit, and it still shows a downward trend after 10^{7} or even 10^{9} cycles. Therefore, in the strength design of aircraft engine parts, the material can not be considered to have an infinite life because of its low stress.

(2) Analysis of S-N curve and fracture shows that the crack initiation location is related to its life. Cracks are generated on the surface below 10^{7} cycles, while it is inside above 10^{7} cycles. As the service life of most aircraft engine parts is more than 10^{7} cycles, it is necessary to strictly control the distribution of the types and sizes of internal defects in materials.

(3) The fracture analysis of crack initiation in the interior shows that the crack initiation source in the ultra-high cycle fatigue at 650 ℃ is mainly the local intergranular fracture and casting defect of the matrix. The characteristic morphology of "fish-eye" with the source of intergranular fracture was also found. In the phase of crack propagation, the mixed propagation of intergranular and cleavage is the main form. Therefore, the casting quality of high-temperature service materials of aircraft engine is strictly controlled to avoid casting defects, and the strength of grain boundary is strengthened through reasonable heat treatment process.

References

[1] YUAN Xiaoling. Nickel-based deformed superalloy [S]. China Aviation Materials Manual. Beijing: China Standard Press, 2001. (in Chinese)

[2] MORRISSEY R, NICHOLAS T. Staircase testing of a titanium alloy in the gigacycle regime[J]. International Journal of Fatigue, 2006, 28(11): 1577-1582.

[3] XU Haifeng, KONG Jian. Effects of Sr and Ce on the high-temperature mechanics and thermal fatigue properties of Nickel-based superalloys[J]. Journal of Shandong Agricultural University (Natural Science Edition), 2020, 51(2): 320-322. (in Chinese)

[4] LI Piao, YAO Weixing. Review on low cycle mechanical fatigue life models of Nickel based single crystal alloys[J]. Material Review, 2020, 34(9): 9124-9131, 9151. (in Chinese)

[5] WANG Zhen, LIANG Jiecun, LI Xide. Effect of load holding time on low cycle fatigue performance of Nickel-based single crystal superalloys[C]//Proceedings of the 26th Annual Academic Conference of Beijing Mechanics Society. [S. l.]: Beijing Mechanics Society, 2020: 1750-1751. (in Chinese)

[6] DONG Jianmin, LI Jiarong, HAN Mei. Examine the effect of corrosion on the high cycle fatigue performance of nickel-based single crystal high temperature alloy[J]. Materials Engineering, 2020, 48(1): 77-83. (in Chinese)

[7] DENG Wenkai, XU Qinghao, JIANG Liang. Thermal and mechanical fatigue properties of IN718 Nickel base superalloy[J]. Chinese Journal of Nonferrous Met-

als, 2019,29(5): 983-989.(in Chinese)

[8] WANG Xin, CHEN Xing, HU Rengao, et al. Fatigue properties and fracture analysis of GH4169 alloy hole structure under cold extrusion[J].Materials Engineering, 2020, 48(6): 156-162.(in Chinese)

[9] LI Hongyu, LIU Yang, XING Jian, et al. Low cycle fatigue property of GH4169 alloy bar for nuclear power plant at 315 ℃[J].Casting Technique,2020,41(1): 6-8.(in Chinese)

[10] ZHONG Liqiong. Study on the influence of surface deformation intensification residual stress field on high cycle fatigue performance of Inconel718 high temperature alloy[D].Guiyang: Guizhou University,2019.(in Chinese)

[11] KOBAYASHI K. High cycle fatigue properties of nickel-base alloy 718[J]. Acta Metall Sinica,2004,17(4): 345-349.

[12] PÉDRON J P, PINEAU A. The effect of microstructure and environment on the crack growth behaviour of Inconel 718 alloy at 650 ℃ under fatigue, creep and combined loading[J]. Materials Science & Engineering, 1982, 56(2): 143-156.

[13] LUNDSTR E, SIMONSSON K, GUSTAFSSON D, et al. A load history dependent model for fatigue crack propagation in Inconel 718 under hold time conditions[J]. Engineering Fracture Mechanics, 2014, 118(2): 17-30.

[14] XIAO L, CHATURVEDI M C, CHEN D L. Low-cycle fatigue behavior of INCONEL 718 superalloy with different concentrations of boron at room temperature[J]. Metallurgical & Materials Transactions A, 2005, 36(10): 2671-2684.

[15] SAKAI T, SAKAI T, OKADA K, et al. Statistical fatigue properties of SCM435 steel in ultra-long-life regime based on JSMS database on fatigue strength of metallic materials[J]. International Journal of Fatigue, 2006, 28(11): 1486-1492.

[16] DONG Xin, LI Peiyuan, WANG Xu, et al. Effect of corrosion on high cycle fatigue performance of TC17 titanium alloy[J]. Material of Mechanical Engineering, 2014, 38(11): 76-79.(in Chinese)

[17] YAN N, KAWAGOISHI N, CHEN Q, et al. Fatigue properties of inconel 718 in long life region at elevated temperature[J]. Key Engineering Materials, 2003, 243/244: 321-326.

[18] YI Hongfei, LU Liantao, ZHANG Jiwang, et al.The influence of loading frequency on the fatigue performance of steel material with long life[J]. Engineering Material Application, 2011, 38(9): 69-72.(in Chinese)

[19] SONG Zongxian, WANG Dongpo, WU Zhisheng, et al. Ultra-high cycle fatigue properties of GH4169 alloy by laser selective melting[J]. Material of Mechanical Engineering,2020,44(11): 72-77.(in Chinese)

[20] SHANYAVSKIY A A, BANOV M D, ZAKHAROVA T P. Principles of physical mesomechanics of narostructural fatigue of metals. Part II. Subsurface fracture of EP741 heat-resistant alloy[J]. Physical Mesomechanics, 2010, 13: 143-151.

[21] BATHIAS C. There is no infinite fatigue life in metallic materials[J]. Fatigue & Fracture of Engineering Materials & Structures, 1999, 22(7): 559-565.

[22] KAWAGOISHI N, CHEN Q, NISITANI H. Fatigue strength of inconel 718 at elevated temperatures[J]. Fatigue & Fracture of Engineering Materials & Structures,2001,23: 209-216.

[23] YAO Liangliang, ZHANG Xiancheng, LIU Feng, et al. Low cycle fatigue property of GH4169 Nickel-based superalloy at high temperature[J]. Material of Mechanical Engineering, 2016, 40(4): 25-29, 64.(in Chinese)

[24] GAO Ren. Ultrahigh cyclic fatigue performance of aluminum alloy friction stir welded joints under different stress ratios[D]. Tianjin: Tianjin University, 2018.(in Chinese)

[25] WEIDNER A, AMBERGER D, PYCZAK F, et al. Fatigue damage in copper polycrystals subjected to ultrahigh-cycle fatigue below the PSB threshold[J]. International Journal of Fatigue,2010,32(6): 872-878.

[26] HONG Youshi, SUN Chengqi, LIU Xiaolong. A review on the mechanism and model of ultrahigh cycle fatigue of alloy materials[J]. Advances in Mechanics, 2018,48: 1-65.(in Chinese)

[27] MUGHRABI H. On “multi-stage” fatigue life diagrams and the relevant life-controlling mechanisms in ultra high-cycle fatigue[J]. Fatigue & Fracture of Engineering Materials & Structures,2002,25(8/9): 755-764.

[28] HOU Jie, DONG Jianxin, YAO Zhihao. Micro-damage mechanism of high temperature fatigue crack propagation in GH4169 alloy[J]. Chinese Journal of Engineering,2018,40(7): 822-832.(in Chinese)

[29] LYU Xudong, DENG Qun, DU Jinhui, et al. In situ observation of high temperature fatigue behavior of GH4169 alloy[C]//Proceedings of the 13th China Superalloy Conference. [S.l.]: High Temperature Materials Branch of Chinese Metal Society, 2015: 79-82.(in Chinese)

[30] GOPIDRIDHNA D, JHA S N. Influence of micro-

structure on fatigue properties of alloy 718[C]//Proceedings of Superalloys. [S.l.]: The Minerals, Metals and Materials Society, 1997: 567-573.

[31] PIERAGGI B, UJINET J F. Fatigue and creep properties in relation with alloy 718 microstructure [C]// Proceedings of Superalloys. [S. l.]: The Minerals, Metals and Materials Society, 1994: 535-544.

[32] LIU Xiaolong, SUN Chengqi, HONG Youshi. Faceted crack initiation characteristics for high-cycle and very-high-cycle fatigue of a titanium alloy under different stress ratios[J]. International Journal of Fatigue, 2016, 92: 434-441.

[33] OSINKOLU G A, ONOFRIO G, MARCHIONNI M. Fatigue crack growth in polycrystalline IN718 superalloy[J]. Materials Science and Engineering: A, 2003, 356 (1/2): 425-433.

Acknowledgements This work was supported by the Tianjin Technical Expert Project (No.19JCTPJC43800); the Tianjin Science and Technology Planning Project (No. 19YFFCYS00090); the Key Project of Science and Technology Cultivation of Tianjin Sino-German University of Applied Technology (No. zdkt2017-006) and the Priority Academic Program Development of Jiangsu Higher Education Institutions.

Authors Dr. **SONG Zongxian** received the B.S. degree in material forming and control engineering from Taiyuan University of Science and Technology, Taiyuan, China, in 2009, where he is now pursuing the Ph.D. degree. His current research interests include fatigue fracture of metal materials and advanced joint technology of aviation.

Prof. **WU Zhisheng** received the Ph. D. degree in material processing engineering from Tianjin University, China, in 2002. He is currently a full professor in Taiyuan University. His research interests include heavy equipment welding and automation and metal 3D printing materials and processes.

Author contributions Dr. **SONG Zongxian** and Prof. **WU Zhisheng** conceived and designed the experiments. Dr. **SONG Zongxian** performed the experiments. Dr. **QI Hao** and Mr. **CHEN Xin** analyzed the data. Mr. **ZHAO Zhenjun**, Mr. **WU Ruizhe** and Mr. **LIAN Zheng** contributed to reagents/materials/analysis tools. Dr. **SONG Zongxian** wrote the paper. All authors commented on the manuscript draft and approved the submission.

Competing interests The authors declare no competing inrests.

(Production Editor: SUN Jing)

GH4169镍基合金650 ℃超高周疲劳性能研究

宋宗贤[1,2]，齐　浩[3]，吴志生[1]，陈　鑫[2]，赵振钧[2]，吴睿哲[2]，廉　政[2]

(1.太原科技大学材料科学与工程学院,太原 030024,中国; 2.天津中德应用技术大学航空航天学院,天津 300350,中国; 3.南京航空航天大学机械结构力学及控制国家重点实验室,南京 210016,中国)

摘要: 采用超声疲劳试验设备,对航空发动机常用的GH4169合金在650 ℃下进行了10^5～10^9周次的疲劳试验。通过对疲劳S-N曲线及疲劳断口形貌的分析,得出以下结论:S-N曲线呈现"阶梯状",第1个拐点出现在10^7周次附近，第2个拐点出现在10^8周次附近。曲线不存在疲劳极限,经过10^7次甚至10^9次循环后仍呈现下降趋势。裂纹萌生的位置与寿命有关,在10^7周次以下裂纹均萌生于表面,10^7周次之后裂纹均萌生于内部。内部起裂源头主要为局部的沿晶断裂和铸造缺陷,裂纹扩展以沿晶和解理的混合断裂为主。

关键词: 镍基高温合金;超高周疲劳;裂纹萌生