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# 論文内容要旨

The development of the high-efficiency technology for gas turbine, jet engine and thermal power plant is required to reduce greenhouse gas emission and to minimize operational costs. Increase in the gas stream and steam temperatures of gas turbine, jet engine and thermal power plant are effective ways to reduce greenhouse gas emission and to enhance the efficiency of power generation.

The selection of the turbine entry temperature (TET), which is the temperature of the hot gases entering into the turbine arrangement, is considered to be important for concerning the design of a gas turbine and jet engine. This is the reason why the gas stream temperature falls since mechanical work is extracted from the gas stream. Therefore, the temperature condition at turbine entry is considered to be the most demanding matter on the turbomachinery. Additionally, the TET of jet engine and gas turbine has been increased year by year. At present, the Japanese National Project is under way to promote the development of component technology for the next generation 1700°C class gas turbine for the power plant. And the 1600°C class gas turbine utilizing some of the technologies developed in the National Project is now under the demonstrative operation. Long-term high reliability of structural component of gas turbine for the base load power plant is required due to the long duration of service, for example $10^5$ h, under high temperature and high stress conditions. At the same time, the turbine inlet peak temperature of jet engine has already reached 1600°C when the airplane takeoff and 1300°C during the cruise flight, thus the high reliability of the component materials of jet engine is also required.

For the coal-fired fuel power plant, the development of the most efficient plant technology is required to minimize carbon dioxide emission, since coal-fired fuel power plants emit in operation much more carbon dioxide than that of natural gas fired combined power plants. In Japan, Electric Power Development Co. (EPDC), who coordinates the development program of all Japanese utilities and manufactures, started the development of technologies on advanced steam condition (Ultra Super Critical: USC) in 1981. And in 1993 the USC power plant was developed in Japan before the rest of the world. For the coal-fired fuel power plant, the conditions of steam temperature and steam pressure have already reached 600°C, 24.5 MPa. As in the United States and Europe, due to the prices and availability of oil and marked slowdown in the construction of nuclear power plants, attention is focused on the coal-fired fuel power plants with high availability and efficiency as new generating capacity addition.

For the structural components for gas turbine, jet engine and thermal power plant, the various heat resistant materials are used. In the gas turbine and jet engine, nickel-base superalloys which have strengthening microstructure such as γ' matrix, are mainly used for the turbine blades. The improvement of creep strength of nickel-base superalloys for the component material of turbine blades, as well as the advances of turbine cooling efficiency and Thermal Barrier Coating, significantly contributes to enhance the operational temperature of gas turbine and jet engine. The advance of nickel-base superalloys which are the superior heat resistant materials has been enabled the long-term operation of gas turbine and jet engine. The nickel-base superalloy is a multiphase alloy, which consists of $\gamma$ matrix, $\gamma$' matrix, MC carbides and so on. The nominal composition of the $\gamma$' matrix is $Ni_3(Al, Ti)$ which occupies 60-70 volume % of $\gamma$ matrix.

The nickel-base superalloy was developed in Britain in 1941 at first. Though the first nickel-base superalloy was produced in wrought form, after that, the conventional cast alloy, directionally solidified alloy, and single crystal alloy were developed in sequence due to the progression of cooling process for the casting. Nowadays, many researches on single crystal nickel-base superalloys are conducted, while the researches on polycrystalline nickel-base superalloys are few. Most of the first stage turbine blades are manufactured by the directionally solidified alloy and single crystal alloy, however the subsequent stage turbine rotor blades and stator vanes are still made of polycrystalline nickel-base superalloys.

In the coal-fired fuel power plants, which rely upon superheated steam at 565°C, high-strength creep-resistant ferritic steels are preferred on account of their lower cost. The Ultra Super Critical power plant has become feasible using conventional technology and modified 9-12%Cr heat resistant steels. The boiler components in the coal-fired fuel power plants operated under a temperature of 600°C are manufactured by W-strengthened 9%Cr ferritic steel which has high creep strength due to the strengthening microstructure such as martensitic lath. The improvement of creep strength of creep-resistant ferritic steel enabled the high operational conditions of steam temperature and steam pressure, 600°C, 24.5 MPa in coal-fired fuel power plant.

In the turbine of the jet engine, of which tolerable rotation speed is approximately 20000 rpm and the peak flame temperature reaches 1600°C, large centrifugal force occurs in turbine blades which have the complex configuration such as cooling holes that induces local creep damage and creep crack initiation by the stress concentration. The crack initiations of gas turbine blades of electric power plant were also reported. Although the conditions of material used are different from nickel-base superalloys in gas turbine and jet engine, the creep crack also initiates at boiler component in the coal-fired fuel power plant which is manufactured by W-strengthened 9%Cr ferritic steel.

Generally, under the high temperature and stress loading conditions, the grain boundaries tend to be the starting point of fracture due to the occurrence of creep damage, even though a material has high creep resistance. Therefore, the accurate evaluation method of creep fracture life for the polycrystalline materials which have many grain boundaries is required for the prediction of creep fracture life.

However, creep crack growth behavior of the polycrystalline materials under the high temperature condition is shown to be complex due to the occurrence of the microstructural strengthening mechanism. Therefore, in order to construct the prediction law of creep fracture life for maintaining operational safety and minimizing operational costs of gas turbine, jet engine and thermal power plant, it is necessary to clarify the mechanisms of creep damage progression and creep crack growth for heat resistant materials with strengthening microstructure.

Previously, various evaluations of creep fracture life have been conducted for the management of safety for using heat resistant materials. There are some studies evaluating the creep fracture life by estimating the microstructural change such as rafting which is the coarsening phenomenon of $\gamma$' participates of nickel-base superalloys due to the high temperature and stress loading application. However, for the nickel-base superalloy, there is a report which shows that the rafting has ended during the early stage of the creep fracture life. Besides, the evaluation of creep damage accumulation for boiler piping in the thermal power plant is conducted by the measurement of void area fraction and hardness of the heat resistant steels. However, it is difficult to apply this method to the nickel-base superalloy because of the voids formation which is the preliminary stage of creep crack initiation was rarely observed.

Since the initiation of creep cracks is tolerable from the view point of current design of gas turbine, the investigation of the crack initiation and growth behaviors is considered to be important. To predict the creep fracture life after the crack initiation, the creep crack growth rate has been evaluated by many parameters such as the stress intensity factor $K$, $C^*$ parameter, load line displacement rate $d\delta/dt$, and $Q^*$ parameter. By the way, the characteristics of creep crack growth rate for nickel-base superalloy and W-strengthened 9%Cr ferritic steel are lower than that for Cr-Mo-V steel so called creep ductile material. Additionally, nickel-base superalloy and W-strengthened 9%Cr steel immediately rupture at the stage of the short crack length as compared with the creep ductile material, so called the creep brittle crack growth behavior. The characteristic of creep crack growth rate for the creep brittle materials estimated by $C^*$ parameters, which is calculated by using creep deformation rate, shows typical dual behavior that takes major portion of total crack growth life, approximately 80 ~ 90% of total life. That is the reason why the evaluation of creep crack growth rate by $C^*$ parameter is difficult to apply for creep brittle materials.

The equation of creep crack growth rate for creep brittle materials obtained from experimental results and the derivation method of creep crack growth life have been proposed by using $Q^*$ concept, previously. On the other hand, with increase in creep brittleness, the difficulty of application of these parameters to the creep brittle material has been indicated in the previous paper. Particularly, the creep crack growth length of the polycrystalline nickel-base superalloy such as IN100 is relatively so short that it seems to be necessary to predict the creep crack initiation life rather than the life of creep crack growth.

As mentioned above, for the nickel-base superalloys and W-strengthened 9%Cr ferritic heat resistant steel which have been classified as creep brittle materials, the clarification of the creep fracture mechanism including creep crack initiation and growth behavior is important for the prediction of creep fracture life.

In this study, firstly the creep crack initiation and growth behaviors and the creep damage formation behaviors of nickel-base superalloys were investigated experimentally by in-situ observational creep crack growth tests and analytically by elastic-plastic creep finite element analyses. And the appropriate prediction method of creep fracture life for nickel-base superalloy was discussed. Furthermore, the difference of the creep damage behaviors and creep crack growth mechanisms of various heat resistant materials with different creep ductility was mentioned. And the appropriate prediction method of creep fracture life was indicated for the nickel-base superalloys and W-strengthened 9%Cr ferritic heat resistant steel.

In chapter 1, review and background of the nickel-base superalloys and W-strengthened 9%Cr ferritic heat resistant steel were discussed.

In chapter 2, firstly in-situ observational creep crack growth tests were conducted to clarify the creep crack initiation and growth behaviors for the polycrystalline nickel-base superalloy IN100. As a results, for the creep brittle IN100, it was clarified that the unstable creep crack growth occurred immediately after the creep crack initiation. And the creep damage formation behavior was estimated by using elastic-plastic creep finite element analysis. Additionally, to give a comprehensive review of microstructural distribution effects on creep damage, the creep damage formation behavior and the creep crack initiation and growth behaviors of polycrystalline nickel-base superalloy IN100 and directionally solidified nickel-base superalloy CM247LC were compared with each other. As a result, the occurrence mechanism of creep damage was found to be depending on the microstructural distribution around the notch tip.

In chapter 3, the prediction method of creep fracture life of polycrystalline nickel-base superalloy IN100 was discussed. From the results shown in chapter 2, the occurrence mechanism of creep damage was found to depend on the microstructural distribution around a notch tip, which causes the scattering of creep crack growth behavior. This leads to the difficulties in the determination of creep crack growth life. Additionally, it is suggested that the short-term creep crack growth life for the polycrystalline nickel-base superalloy, in which the rafting of γ' precipitates does not occur, is difficult to be described by the $Q^*$ parameter due to the scattering of experimental data. Therefore, it is considered that the prediction of creep crack initiation life is important for the life prediction of polycrystalline nickel-base superalloy. The prediction method of creep crack initiation life for polycrystalline nickel-base superalloy was considered to be reasonable by using the creep deformation curve on the basis of the hybrid approximation of linear and non-linear creep deformation curves. It was shown that the creep crack initiation life could be predicted by this method accurately.

In chapter 4, the combined analysis which consisted of the elastic-plastic creep finite element analysis and the vacancy diffusion analysis was proposed to clarify the difference of creep damage formation behaviors between different creep brittle materials. In the previous work, using the in-situ observational testing machine of creep crack growth, it was shown that creep damage formation and crack growth behaviors show variant characteristics depending on the creep ductility among different creep brittle materials. From these results, the creep damage formation behaviors, which consist of voids and micro cracks, were considered to be affected by the behavior of the vacancy diffusion and concentration. On the other hand, the creep damage formation behaviors of typical brittle materials were considered to be caused at the site of vacancy origination, that is, the site of plastic deformation caused by high equivalent stress.

In chapter 5, the creep crack growth mechanism of W-strengthened 9%Cr ferritic heat resistant steel was clarified by the elastic-plastic creep finite element analysis and the results were compared with that of IN100. The W-strengthened 9%Cr ferritic heat resistant steel is a creep brittle material that the creep damage formation behaviors were considered to be affected by the behavior of the vacancy diffusion and concentration, which results in typical periodic convexo-concave fracture unit cracking and it is well estimated by $Q^*$ parameter.

In chapter 6, the concluding remarks were conducted.

# 論文審査結果の要旨

高効率発電機器や航空機ジェットエンジンに用いられるクリープ脆性材料の損傷形成挙動の解明とクリープ破壊寿命則の構築は，工学的観点から重要であるにもかかわらず明確ではない。これはクリープ損傷形成挙動が材料組織に依存し，材料組織と微視・巨視力学的要因との関連性が未だ明らかにされていないことによる。著者は，その場観察試験，EBSD 解析と有限要素弾塑性・クリープ解析，空孔拡散解析という実験と理論解析両面から巨視および微視損傷形成挙動を明らかにし，クリープ脆性材料のクリープ寿命則構築に関する基礎基盤的知見を得ている。本論文は，これらの研究成果をまとめたものであり，全編 6 章からなる。

第 1 章は緒論であり，本研究の背景および目的と意義について述べている。

第 2 章では，多結晶および一方向凝固 Ni 基超合金切欠き材における巨視クリープ損傷形成挙動を，その場観察試験により調べ，さらに EBSD 解析を行って結晶粒界に沿って発生する微視的な結晶高方位差域（高 KAM 域）が巨視損傷域形成前に生じていることを示している。また，試作改良した有限要素弾塑性・クリープ解析ソフトを用いて高 KAM 域が，ひずみエネルギー集中域に対応しており，この領域が，微視クリープ損傷域であることを見出している。これらの結果は，微視クリープ損傷評価の観点から有効かつ重要な知見である。

第 3 章では，クリープ寿命が微視損傷に敏感に影響される場合について，線形・非線形融合変形構成則を用いて，クリープ変位を予測し，き裂発生寿命を精度よく予測する方法を提案している。この結果は，き裂発生後，急速に最終破壊に至る多結晶 Ni 基超合金 IN100 のクリープ寿命予測にとって有効かつ重要な知見である。

第 4 章では，Ni 基超合金から 9Cr 鋼に至るまでのクリープ脆性度の相違による多様なクリープ損傷形成発現機構を明らかにするために，有限要素弾塑性・クリープ解析と空孔拡散解析を融合した巨視・微視損傷解析法を提案し，その場観察により得られる多様なクリープ損傷形成挙動を理論的に再現することに成功している。この結果は，損傷形成挙動の相違に応じてクリープ寿命則を構築する場合に重要となる基礎基盤的な知見である。

第 5 章では，若干の塑性変形を許容するフェライト系耐熱鋼である W 添加 9Cr 鋼に顕著に見られる周期的な凹凸を有するクリープき裂の発現機構が塑性変形により，すべり方向にボイドが先行して発生する機構であることを実験および力学解析により明らかにし，き裂成長速度から寿命を予測することの妥当性を示している。この結果はフェライト系耐熱鋼のクリープ寿命則構築に重要な知見である。

第 6 章は結論である。

以上要するに本論文は，クリープ脆性材料の寿命則構築のために，微視および巨視損傷力学的研究を実験および力学解析の両面から行い，クリープ脆性度に応じて適切な寿命則を提示したものである。ここで得られた成果は，微視および巨視クリープ損傷形成挙動の解明と寿命則構築に有益なものであり，ナノメカニクスおよび機械工学の発展に寄与するところが少なくない。

よって，本論文は博士(工学)の学位論文として合格と認める。