Original Article

Microstructure evolution and mechanical behaviors of alumina-based ceramic shell for directional solidification of turbine blades

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ABSTRACT

The high temperature mechanical behaviors of Al₂O₃-based ceramic shell for the directional solidification of turbine blades were investigated. Isothermal uniaxial compression tests of ceramic shell samples were conducted on a Gleeble-1500D mechanical simulator with an innovative auxiliary thermal system. The microstructures of sintered, heat-treated, and tested samples were characterized using scanning electron microscope and X-ray powder diffraction. The experimental stress–strain results of heat-treated samples were obtained. The eutectic mixture region composed of ZrO₂ and SiO₂ that regenerated by the decomposition reaction of excess zirconium silicate appears in the samples after heat treatment at 1500. The fracture type of CST25 and CST700 is brittle fracture, but CST1100 and CST1400 have thermo-viscoelastic and viscoplastic properties under stress conditions at high temperatures (>1100). The evolution of zirconium silicate decomposition reaction during the directional solidification process is analyzed. The SiO₂ particles provide the major viscosity source of the ceramic shell at high temperatures. The thermo-viscodamage constitutive model of Al₂O₃-based ceramic shell for the directional solidification of single crystal superalloy is established.

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1. Introduction

Single-crystal (SX) nickel-based superalloys have been employed in both aircraft engines and land-based power-generation applications for many years owing to their terrific high temperature mechanical properties and excellent anti-fatigue performance. In order to achieve better performance to meet the needs for advanced manufacturing, many researchers have paid attention to the optimization of alloy composition and the improvement of material processing technologies, which includes the directional solidification technology. The turbine blades are manufactured by pouring the molten metal into the ceramic shell mold, and the ceramic
shell mold is subjected to high temperatures. Therefore, the quality of the ceramic core and shell mold used in the directional solidification process would immediately affect the performance, dimensional integrity, product yields and production costs of the single crystal superalloy [1–4]. In order to improve the refractoriness and strength of the shell, fine alumina is added to the filler [5–7].

In the study of the ceramic shell for directional solidification of single crystal superalloy, the current research focuses on the thermal resistance, pre-heating temperature, physical property, interface reaction, fabrication process, and composition, etc. [5–9]. Szeliga et al. presented the value of interface thermal resistance (ITR) between the IN 713C nickel superalloy casting plate and the ceramic shell mold surface [5]. The influence of internal thermal baffle between casting and central rod of the ceramic mold on the shape of liquids isolus was studied. A modification of standard ceramic shell mold for the manufacturing of single crystal castings with Bridgman method was proposed [6]. Venkat et al. discussed the improvement in the shell refractoriness and dimensional stability of the castings of gas turbine components made from Ni-base superalloys by directional solidification process [7]. Wu et al. developed an integral fabrication process of ceramic casting mold for complex turbine blades by combining stereolithography technology with gelcasting [10].

Chen et al. reported that the ceramic shells were made by the standard procedure for preparation of ceramic molds in the investment casting. The influence of active elements C and Hf on the interface reactions and wettability between the ceramic mold material and a Ni3Al-based superalloy was examined using a sessile drop experiment [9]. Konrad et al. investigated investment casting molds with different numbers of shells and pre-heating temperatures. The primary layer consisted of colloidal silica bound ZrSiO4 with additions of CoAl2O4 to achieve fine grains and to reach a good surface quality, whereas the following layers consisted of mullite bound by colloidal silica. Physical properties of the investment shell mold were analyzed by Laserflash methods and differential scanning calorimetry (DSC) for temperatures up to 1300 °C [8].

There have been some studies about the stress problems of traditional investment casting processes, and thermal stresses occur in the cast parts which lead to different consequences during casting (e.g., distortion, crack, hot tear and residual stresses, etc.). These can have extremely detrimental effects in the quality of cast products. Shell cracking and damage occur mostly on the edges of components, such as trailing edges and sharp corners, where reduced shell thickness and high stress result in ceramic failure. The ceramic mold should have sufficient green and fired strength to withstand dewaxing and metal pouring stresses [11]. However, the high-temperature mechanical behavior and constitutive model of Al2O3-based ceramic shell for directional solidification of SX Superalloy is rarely reported in the literature. The study of high-temperature mechanical behaviors and constitutive model of the ceramic shell plays an important role on controlling the dimensional integrity of turbine blade, reducing the production costs of turbine blades, and improving the yield of casting, which also helps to investigate the reasons of crack and fracture in the ceramic casting mold. Particular attention is paid to figure out the high-temperature mechanical behaviors and microstructural evolution. In addition, further research should be conducted into the constitutive model of Al2O3-based ceramic shell for directional solidification of single crystal superalloy.

2. Experimental procedure

2.1. Directional solidification process

The Bridgman method is used for the directional solidification of SX blades. The whole process includes melting, pouring and solidification in the directional solidification furnace. The schematic of the furnace is shown in Fig. 1, which mainly includes a heating zone, a graphite baffle, a copper plate, a cooling zone and a withdrawal device [12]. After the superalloy is poured, the superalloy, core and shell with water-cooled copper plate are withdrawn from the heating zone to the cooling zone, and the ceramic core and shell undergo a rapid cooling process. During the process, the withdrawing rate is 4 mm/min, the heating zone temperature is set to 1500 °C, and the temperature of circulating cooling water is 40 °C [13].

2.2. Experimental methods and design

The prime coat consisted of a zircon filler and a colloidal silica binder. Alumina grit was applied as a primary stucco. A seven-coat mold backup was manufactured by use of ceramic slurries based on zircon powders, alumina powders and colloidal silica binder. Alumina grits were applied as backup stuccos. The shell molds, having an average shell thickness of 8.8 mm, were dewaxed in a steam autoclave and then sintered in a sintering furnace at 1000 °C for 60 min (CSTS).

In order to simulate the realistic directional solidification process, some of the sintered samples were subsequently subjected to heat treatment at 1500 °C for 30 min (CST25, CST700, CST1100, CST1400). The ceramic shell samples (CST25, CST700, CST1100, CST1400) were removed from the
furnace at 1000 °C to the atmosphere at 25 °C for simulating the realistic rapid cooling process during directional solidification from the heating zone to the cooling zone. The ceramic shell samples (CST25, CST700, CST1100, CST1400) were tested at 25 °C, 700 °C, 1100 °C and 1400 °C, respectively. The ceramic shell samples were held at the test temperatures for about 8 min. The average size of the ceramic shell was 15.2 mm in diameter and 8.8 mm in length. The heat treatments of ceramic shell samples to simulate the directional solidification process are summarized in Table 1.

Cylindrical samples were prepared and applied in isothermal uniaxial compression tests on a Gleeble-1500D mechanical simulator with the auxiliary thermal system. The strain-rate of 10⁻⁴/s was chosen. Taken into consideration the inhomogeneity of the ceramic shell, the tests of five samples (CSTS, CST25, CST700, CST1100, CST1400) were repeated 5–7 times. The experimental results of the ceramic shell mechanical properties are mainly processed by the arithmetic mean:

\[ \bar{M} = \frac{1}{N} \sum_{i=1}^{N} M_i \]  

(1)

where \( N \) is the number of sample tests, \( M \) is critical strain, compressive strength or elastic modulus. The microstructures of sintered, heat-treated, and tested samples were characterized using Zeiss Merlin VP Compact field emission gun scanning electron microscope (FEG-SEM) equipped with energy dispersive X-ray spectroscopy (EDS). The FEG-SEM was operated at an accelerating voltage of 15 kV and a working distance <7 mm in backscattered electrons (BSE) imaging mode. The X-ray diffraction patterns of the samples were performed on a diffractometer (XRD, Rint-2000V/PC, Rigaku, Japan) using Cu-Kα radiation.

2.3. High-temperature experimental system for mechanical behaviors

The Gleeble system has been widely employed in the study of constitutive laws for materials. It is mainly composed of a heating system, mechanical system and computer control system [14,15]. Fig. 2 shows ceramic shell samples and Gleeble uniaxial compression experiment, respectively. Because of the fact that the thickness of the samples used in the compression experiment is close to the actual thickness of the Al₂O₃-based ceramic shell, the uniaxial compression experiment machine takes unique advantage of the measurement for the stress–strain and other mechanical properties of the ceramic shell.

However, the uniaxial compression experimental machine and the loaded samples need to form a closed current loop so that the current can heat the samples. The machine can only be used to measure the high temperature mechanical properties of the conductive material. In order to measure the high temperature mechanical properties of non-conductive materials, such as the ceramic shell and core, on the uniaxial compression experiment machine, we designed and developed an auxiliary thermal device.

Fig. 3 shows the schematic diagram of the auxiliary thermal system of the Gleeble testing. The main components of the auxiliary thermal system are compression bars, silicon carbide screw tube, insulation fiber box, S-type thermocouple for temperature control, and temperature control cabinet [16]. The auxiliary thermal system utilizes the principle of current flowing through silicon carbide screw tube to generate a heating zone. Due to the high electrical conductivity and high temperature resistance of silicon carbide, the maximum temperature of the heating zone can be up to 1600 °C, and the radiant

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Table 1 – Heat treatments of ceramic shell samples to simulate the directional solidification process and test conditions.

<table>
<thead>
<tr>
<th>Ceramic shell samples</th>
<th>Sintering</th>
<th>Heat treatment</th>
<th>Test temperature</th>
</tr>
</thead>
<tbody>
<tr>
<td>CSTS</td>
<td>60 min @1000 °C</td>
<td>–</td>
<td>25 °C</td>
</tr>
<tr>
<td>CST25</td>
<td>60 min @1000 °C</td>
<td>30 min @1500 °C</td>
<td>25 °C</td>
</tr>
<tr>
<td>CST700</td>
<td>60 min @1000 °C</td>
<td>30 min @1500 °C</td>
<td>700 °C</td>
</tr>
<tr>
<td>CST1100</td>
<td>60 min @1000 °C</td>
<td>30 min @1500 °C</td>
<td>1100 °C</td>
</tr>
<tr>
<td>CST1400</td>
<td>60 min @1000 °C</td>
<td>30 min @1500 °C</td>
<td>1400 °C</td>
</tr>
</tbody>
</table>

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Fig. 2 – Ceramic shell samples and Gleeble uniaxial compression experiment.
heating section forms a uniform high temperature environment. During experiments, the temperature fluctuation and uniformity was maintained within ±4 °C. The heating section heats the tested sample by thermal radiation. The sample is centered in the heating section and loaded by two compression bars.

3. Results and discussion

3.1. Microstructural of the sintered and heat-treated samples

It can be seen in Fig. 4(a) and (c) that the size and distribution of macropores in the sintered samples (60 min @1000 °C) and heat-treated samples (60 min @1000 °C + 30 min @1500 °C) are similar, and their morphological characteristics are almost identical as well. Therefore, the permeability of directional solidification is not affected. Comparing Fig. 4(b) with Fig. 4(d), the microstructural differences between the sintered and heat-treated samples can be observed at higher magnification. There are a large number of gaps in the sintered samples. However, the fine gaps between the particles in heat-treated samples are filled and re-sintered to form a large particle. The composition in the gaps at high temperature is mainly β-cristobalite, which forms α-cristobalite at room temperature. Due to the re-sintering, the compressive strength of the ceramic shell samples is improved at room temperature.

3.2. High-temperature mechanical behaviors

Fig. 5 shows the stress-strain experimental data zone of CTS25, CST700, CST1100 and CST1400, respectively. The fracture type
of CST25 and CST700 is brittle fracture, while CST1100 and CST1400 show thermo-viscoelastic and viscoplastic behavior. When the strain of the samples (CST1100, CST1400) exceeds 0.15, they are still not pulverized. The compressive strength of CST700 (51.14 MPa) is higher than that of CST25 (20.22 MPa), which can be ascribed to the conversion of α-cristobalite to β-cristobalite. As the transition from α-cristobalite to β-cristobalite is observed to occur between 200 and 280 °C [17], the test temperature of 700 °C allowed testing of samples in the stability range of β-cristobalite. The densities of α-cristobalite and β-cristobalite are about 2.32 g/cm³ and 2.22 g/cm³, respectively, and the expansion of the β-cristobalite volume makes the samples more compact [17]. Furthermore, β-cristobalite has higher strength than α-cristobalite [18]. The stress-strain experimental data zones of CST1100 and CST1400 become flatter, which illustrates that the high temperature experimental data is reproducible, and the high temperature mechanical properties of the samples obviously decrease with increase in the test temperature.

Fig. 6 provides the macrostructure of uniaxial static compressive failure for Al₂O₃-based ceramic shell samples. And the differences of the fracture types between CST25, CST700 and CST1100, CST1400 can be seen more clearly. The fracture type of CST25 and CST700 is brittle fracture, and the average volume of the residual pieces from CST700 is smaller than that from CST25.

The deformation of CST1100 and CST1400 exceeds 0.15, but the samples almost remain intact without comminuted fracture. It is observed that the cracks occur on the sides of the samples, but the particles at the cracked position are almost not shed, indicating that there is some adhesion between the cracked fragments and the matrix material at high temperatures (~1100 °C). This viscous connection can be further observed on SEM images of the sample side view, as shown in Fig. 7 (b) and (d), which clearly shows the phenomenon of viscous separation at the cracks of CST1100 and CST1400. The crack location should be viscous semi-solid at high temperature and mainly distribute around the large Al₂O₃ particles.

Fig. 7 shows the microstructural evolution of the samples (CST1100, CST1400). The eutectic mixture composed of ZrO₂ and SiO₂ (the eutectic mixture will be detailed in the following Section 3.3) appears largely in the fracture of CST1100 and CST1400, which also adheres near the large Al₂O₃ particles. However, there are no cleavage and dimple fracture facets of the ZrO₂, SiO₂, ZrSiO₄, and Al₂O₃ particles. CST1100 and CST1400 have thermo-viscoelastic and viscoplastic properties under stress conditions at high temperatures. This thermo-viscoelastic and viscoplastic properties are mainly due to the SiO₂ in the eutectic mixture, and SiO₂ starts the viscous slip in the high temperature environment.

### 3.3. The reason for thermo-viscoelastic and viscoplastic properties

The partial area diagram of the eutectic mixture region shown in Fig. 7 is enlarged and analyzed using EDS point elemental analysis mode. According to the results of EDS spectra in Fig. 8, it is clearly understood that the major component of the white gray is ZrO₂ and the major component of the black gray is SiO₂, and both of them form a eutectic system. This eutectic structure image is consistent with the literature results [19–21]. The above-mentioned viscous separation at the crack locations of CST1100 and CST1400 is mainly resulted from the viscous slip and viscous separation of SiO₂.

During the heat treatment of ceramic shell samples to simulate the directional solidification process, the excess
Fig. 6 – Macrostructure of uniaxial static compressive failure for ceramic shell samples.

Fig. 7 – Microstructural investigation of the samples (CST1100, CST1400). (a) SEM micrographs of CST1100, (b) partial enlargement of (a), (c) SEM micrographs of CST1400, (d) partial enlargement of (c).
zirconium silicate at the surface of $\text{Al}_2\text{O}_3$ large particles occurs decomposition reaction as follows:

$$\text{ZrSiO}_4^{1500\,^\circ\mathrm{C}} = \text{ZrO}_2 + \text{SiO}_2$$

The $\text{SiO}_2$ regenerated newly by the decomposition reaction and the original $\text{SiO}_2$ fill the gaps between the particles, as shown in Fig. 4(d), and they are re-sintered at a high temperature. Consequently, the compressive strength of the samples can be improved at room temperature. Under high-temperature (>1100 °C) and stress conditions, the viscous slip of $\text{SiO}_2$ becomes the weakest part of the ceramic shell, which weakens the strength of the ceramic shell. This is one of the main reasons for the high temperature creep of the ceramic shell during the directional solidification process.

Though the samples produce $\text{ZrO}_2$ and $\text{SiO}_2$ at 1500 °C, the $\text{SiO}_2$ remains brittle at temperatures below 700 °C and does not initiate slippage under stress. Therefore, the fracture type of samples CST25 and CST700 is brittle fracture. The brittle fracture expands inside the $\text{ZrO}_2$, $\text{SiO}_2$, and $\text{ZrSiO}_4$ particles close to the $\text{Al}_2\text{O}_3$ matrix particles. This further proves that the eutectic mixture region near the $\text{Al}_2\text{O}_3$ matrix particles is the weakness position of the shell both at room temperature or high temperature, due to the decomposition reaction of zirconium silicate.

Fig. 9 exhibits the microstructural investigation of the samples (CSTS, CST25, CST700, CST1100, CST1400) fracture surface. Fig. 9(a), (c), (e), (g), and (i) are the images of secondary electrons (SEM) and Fig. 9(b), (d), (f), (h), and (j) are the images of the backscattered electrons (BSE). In the BSE images, the white gray is an element with a large atomic mass. The major component of the white gray is $\text{ZrO}_2$, the major component of the black gray is $\text{SiO}_2$ or $\text{Al}_2\text{O}_3$, and the major component of the middle gray is $\text{ZrSiO}_4$. After sintering at 1000 °C for 60 min, the sample (CSTS) does not appear eutectic mixture region of $\text{ZrO}_2$ and $\text{SiO}_2$ (Fig. 9(a) and (b)), which proves that zirconium silicate does not decompose. In Fig. 9(c)–(j), the eutectic mixture region composed of $\text{ZrO}_2$ and $\text{SiO}_2$ appears in the samples after heat treatment at 1500 °C. It is also observed in Fig. 9(d), (f), (h), and (j) that the eutectic mixture composed of $\text{ZrO}_2$ and $\text{SiO}_2$ is due to the decomposition of $\text{ZrSiO}_4$ because there is a clear decomposition interface in the $\text{ZrSiO}_4$ particles. The X-ray powder diffraction (XRD) patterns of the samples (CSTS, CST25, CST700, CST1100, CST1400) are presented in Fig. 10. The XRD results also show that the decomposition reaction of $\text{ZrSiO}_4$ does not occur in CSTS, because there is no diffraction pattern of $\text{ZiO}_2$, while the diffraction patterns of $\text{ZiO}_2$ appear in CST25, CST700, CST1100, and CST1400, which indirectly indicates that the decomposition reaction of $\text{ZrSiO}_4$ occurs in the samples after heat treatment at 1500 °C.

The cleavage and dimple fracture facets inside the samples (CSTS, CST25, CST700) are present obviously on the fracture surface in Fig. 9(a), (c), and (e), which directly proves that the fracture type of CSTS, CST25, and CST700 is brittle fracture, and corroborates the data of stress-strain. The results show that the crack propagation takes place inside the $\text{ZrO}_2$, $\text{SiO}_2$, and $\text{ZrSiO}_4$ particles. The $\text{ZrO}_2$ particles, the $\text{SiO}_2$ particles, and the $\text{ZrSiO}_4$ particles are all located near the surface of the $\text{Al}_2\text{O}_3$ large matrix particles. There is no sign of sliding between the $\text{ZrO}_2$ particles and the $\text{SiO}_2$ particles.

As shown in Fig. 9(g) and (i), the viscous sliding topographies are clearly observed, and the viscous sliding topographies spread over the entire surface of the $\text{Al}_2\text{O}_3$ large matrix particles. Combined Fig. 9(h) and (j), it can be seen that the composition of this droplet-like viscous material is $\text{SiO}_2$ particles wrapped $\text{ZrO}_2$ particles or $\text{ZrSiO}_4$ particles. The $\text{SiO}_2$ particles provide the major viscosity source of the ceramic shell at high temperatures.

Schematic diagram of the process of zirconium silicate decomposition reaction are shown in Fig. 11. In 1500 °C high temperature environment, the excessive $\text{ZrSiO}_4$ near the surface of $\text{Al}_2\text{O}_3$ particles precipitates $\text{ZrO}_2$ and $\text{SiO}_2$, and they gradually nucleate and grow into $\text{ZrO}_2$ grains and $\text{SiO}_2$ grains. The $\text{SiO}_2$ grid grains would wrap $\text{ZrO}_2$ grain in its interior.

At the high-temperature (>1100 °C) and stress conditions, the $\text{SiO}_2$ grains initiate a viscous flow with $\text{ZrO}_2$ grains (i.e., high temperature creep), causing the strength of the $\text{Al}_2\text{O}_3$-based ceramic shell to decrease. The reaction surface initially appears at the interface of $\text{Al}_2\text{O}_3$ grains and $\text{ZrSiO}_4$ grains, which diffuses from the inside to the outside. The silicon dioxide gradually becomes enriched and nucleates, and the lines of silicon dioxide grain appear gradually in the particles. Between the reticular grains of silicon dioxide, a $\text{ZrO}_2$-rich zone is formed and nucleated, resulting in the formation of $\text{ZrO}_2$ grains. As the grain size of the silica increases, the reticular grains gradually merge to form large $\text{SiO}_2$ grains, but the total amount of the $\text{SiO}_2$ grains is less than that of the $\text{ZrO}_2$ grains. In the whole particle, the $\text{SiO}_2$ mainly exists in the reticular structure, and the $\text{SiO}_2$ grains wrap $\text{ZrO}_2$ grains through this reticular structure. In the case of high-temperature (>1100 °C) and stress, the $\text{SiO}_2$ grains will occur viscous flow with the $\text{ZrO}_2$ grains.
3.4. Thermo-viscodamage constitutive model

The damage evolution $\dot{D}$ can be a function of hydrostatic stress $\sigma_{kk}$, stress, $\sigma_{ij}$, strain $\varepsilon_{ij}$, strain rate $\dot{\varepsilon}_{ij}$, temperature $T$, and damage history $D$, such that [22]:

$$\dot{D} = F \left[ \sigma_{kk} (t), \sigma_{ij} (t), \varepsilon_{ij} (t), \dot{\varepsilon}_{ij} (t), T (t), D (t) \right]$$  \hspace{1cm} (2)

The first approximation of the damage evolution law for $\text{Al}_2\text{O}_3$-based ceramic shell is proposed as an exponential form of the total effective strain in this study:

$$\dot{D} = \Gamma^\varepsilon \exp (k \varepsilon_{\text{Tot}}^*)$$  \hspace{1cm} (3)

where $\Gamma^\varepsilon$ is a damage viscosity parameter, $\varepsilon_{\text{Tot}}^* = \sqrt{\varepsilon_{ij}^* \varepsilon_{ij}^*}$ is the effective total strain in the effective configuration. $k$ is a
material parameter and $\epsilon^v_{ij}$ includes both viscoelastic and viscoplastic components. The dependence of the damage density evolution equation on the total strain implicitly couples the damage model to the viscoelasticity and viscoplasticity models. Therefore, the changes in loading temperature, time, and rate implicitly affect the damage evolution through changes in viscoelastic and viscoplastic strains. However, the time of rupture in tests and peak point in the stress–strain diagram for the constant strain rate tests are highly correlated to the stress. Consequently, it can be assumed that the damage viscosity variable (in Eq. (3)) is a function of the stress. The stress dependency of the damage viscosity parameter is hypothetically governed by a power law function, which can be described as follows:

$$\Gamma^v = \Gamma^v_0 \left( \frac{Y}{Y_0} \right)^q$$

(4)

where $q$ is the stress dependency parameter; $Y_0$ and $\Gamma^v_0$ are the reference damage force and the reference damage viscosity parameter calculated at a reference stress, respectively; and $Y$ is the damage driving force in the nominal (damaged) configuration. Moreover, assuming the damage viscosity parameter to be a function of the damage force $Y$, in the nominal (damaged) configuration instead of the effective (undamaged) configuration allows one to contain damage history effects, such that by using the effective stress concept one can rewrite $Y$ as follows [23,24]:

$$Y = Y^* (1 - D)^2$$

(5)

Additionally, the damage density evolution highly depends on temperature. Therefore, the following damage evolution
Fig. 12 – Experiment results of typical compression stress–strain and simulation results based on thermo-viscodamage model.

law for Al2O3-based ceramic shell in uniaxial compression experiment can be deduced using Eqs. (3)–(5):

\[
\begin{align*}
\sigma^e_1 (t) &= \int_{-\infty}^{t} E \cdot (t-x) \cdot \frac{d\varepsilon (x)}{dx} \cdot dx \\
\sigma^e_2 (t) &= \sigma^e_1 / (1 - D) \\
D &= \int_0^\infty \left[ Y (1 - D) \right]^2 \exp \left( k T \right) \\
\end{align*}
\]

The experiment of results of typical compression stress–strain of Al2O3-based ceramic shell and the simulation results based on thermo-viscodamage model are presented in Fig. 12. It can be seen that the thermo-viscodamage model can effectively reflect the uniaxial compressive mechanical properties of Al2O3-based ceramic shell for directional solidification of SX blades, including the strength, elastic modulus, viscoelastic stage, and viscoplastic stage.

Comparing Fig. 12(a) with (b), it can be observed that the elastic stage of CST1100 (strain 0–0.088) is much larger than that of CST1400 (strain 0–0.025), while the strength of CST1100 is much greater than that of CST1400. This shows that in the early stage of directional solidification, if there is a small stress, the ceramic shell will enter the viscoplastic stage after a short period of elastic phase, and the ceramic shell will continue to shrink, making the blade dimensional integrity difficult to achieve.

4. Conclusions

(1) The high temperature mechanical behaviors of Al2O3-based ceramic shell for the directional solidification of single crystal superalloy were investigated on a Gleeble-1500D mechanical simulator with an innovative auxiliary thermal system. The maximum temperature of the heating zone can be up to 1600 °C, and the temperature fluctuation and uniformity was maintained within ±4 °C.

(2) There are a large number of gaps in the sintered samples. However, the fine gaps between the particles in heat-treated samples are filled and re-sintered to form a large particle. The compressive strength of the ceramic shell heat-treated samples is improved at room temperature.

(3) The fracture type of Al2O3-based ceramic shell samples is brittle fracture, but when the temperature exceeds 1100 °C, it will show thermo-viscoelastic and viscoplastic behavior. The high temperature mechanical properties decrease obviously with the increase of the test temperature.

(4) The evolution of the eutectic mixture of ZrO2 and SiO2 in the Al2O3-based ceramic shells was analyzed. Under high-temperature (>1100 °C) and stress conditions, the strength of the ceramic shell was weakened attributed to the viscous slip of SiO2. This was a major reason for the high-temperature creep of the ceramic shell during DS.

(5) Thermo-viscodamage constitutive model is established. The thermo-viscodamage model can effectively reflect the uniaxial compressive mechanical properties of Al2O3-based ceramic shell for directional solidification of SX blades.

Conflicts of interest

The authors declare no conflicts of interest.

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References


