Hot deformation behavior of hard-to-deform Ni-based Alloy

X D Lu¹², Y W Zhang¹², S Y Shi², B Wen², X Su² and J H Du²

¹Beijing Key Laboratory of Advanced High Temperature Materials, Central Iron and Steel Research Institute, Beijing 100081, China
²Beijing Gaona Aero Material CO., LTD., Beijing 100081, China
Email: lxdong0700@hotmail.com

Abstract. The very poor hot workability of hard-to-deform GH4742 alloy is represented in the aspects of very narrow available deformation temperature range, high deformation-resistant force and low thermoplasticity. In this paper, the hot deformation behavior of a Ni-based superalloy was studied in the temperature range of 1060-1160°C and strain rate range of 0.05-0.5 s⁻¹ using hot compression tests. A hot deformation equation is given to characterize the dependence of peak stress on the temperature and strain rate. Results of hot compression test of the GH4742 alloy were indicated that the hot deformation is sensitive to temperature and strain rate. A special heat treatment to change the ingot microstructure is proposed to improve the hot workability of the GH4742 alloy. Modified microstructure induces that the hot deformation flow stress of GH4742 alloy was decreased effectively and hot deformation plasticity was increased obviously.

1. Introduction
Modern aerospace vehicles require faster, more efficient and more flexible control, which require the engine to provide more powerful power. The most effective way to improve engine power is to increase the turbine inlet temperature of the engine, which requires the main hot-end components to have higher operating temperature and high temperature performance. The turbine disk is one of the most critical hot-end components. Ni-based superalloy is an ideal turbine disk material. In order to improve the high temperature strength and service temperature, a large number of alloying elements are often added.

The hard-to-deform GH4742 superalloy has a high content of strengthening phase (volume fraction up to 35%), poor hot workability which is mainly reflected in the three areas of narrow deformation temperature range, large deformation resistant force and low thermoplasticity [1-3]. The most important alloying elements of hard-to-deform alloys are the γ' phase forming elements Al, Ti, Nb, etc., and precipitate as a strengthening phase at medium temperatures, resulting in obvious aging effect. At high temperature, it is generally dissolved in the matrix in the form of solute atoms, which plays a solid solution strengthening role. Another kind of alloying elements is W, Mo, Co, Cr, etc., both at room temperature and high temperature, they are generally in the form of solute atoms, so reinforcement is very obvious, but to be stable at high temperatures [3]. However, increasing of strengthening elements will lead to decreasing of melt temperature and increasing of recrystallization temperature. Very narrow available deformation temperature range, high deformation-resistant force and low thermoplasticity are the main obstacles that restrict hot workability of the GH4742 alloy [3-5].
The poor hot workability of high-performance, hard-to-deform alloys is caused by the strong aging strengthening of the \( \gamma' \) phase coherent with the matrix and the solid solution strengthening effect of various alloy elements. From the law of aging strengthening, it can be known that the precipitation phase has the largest strengthening effect at a suitable size. When the size is further increased, the strengthening effect will rapidly weaken. In general, two methods can be used to roughen the \( \gamma' \) phase: one is to slowly cool in the \( \gamma + \gamma' \) two-phase region, and the other is to hold it for a long time at a suitable temperature in the \( \gamma + \gamma' \) two-phase region. The \( \gamma \) phase in the as-cast structure undergoes relatively large changes after the slow cooling process, mainly forming coarse spherical and irregular petal-shaped \( \gamma' \) phases and fan-shaped structures. These two factors play a similar role, that is, the slower the cooling rate, the more they tend to form coarse irregular \( \gamma' \) phases and fan-shaped structures; similarly, the higher the concentration of the \( \gamma' \) phase forming elements, such as Al, Ti, and Nb, the more it tends to form coarse irregular phase and fan structure. Due to the high energy at the grain boundary, the resistance to be overcome nucleation of the \( \gamma' \) phase is small, so a large semicircular \( \gamma' \) phase is often formed on the grain boundary. After the slow cooling treatment, the grain boundaries are bent to a certain degree, especially at the dendrites, this feature is more obvious. In summary, the ingot is slowly cooled after homogenization, forming coarse \( \gamma' \) phase and curved grain boundaries, preparing the structure for the hot deformation of hard-to-form alloys, which can effectively improve the hot working performance of such alloys [6,7]. Generally, scholars use thermal compression experiments to study the mechanical behavior and microstructure evolution of materials during plastic deformation [8-14]. In this paper, the stress-strain curve, thermal deformation constitutive relationship and dynamic recrystallization process of GH4742 alloy were studied under the experimental conditions of 1060~1160℃ and strain rate 0.05~0.5s\(^{-1}\). In this paper, microstructure evolution of GH4742 alloy under slow cooling treatment condition at cooling rate of 5℃/h and 50℃/h was studied, a method of improving hot workability of high-alloyed superalloy was raised, micro-mechanism of the hot deformation process was analyzed.

2. Experimental procedure

The chemical composition of GH4742 superalloy used in this investigation is listed in Table.1. The Gleeble test material was selected from 2/3 R of the GH4742 alloy φ508mm ingot slice. After low magnification analysis, the grain structure was uniform and fine. Cut a number of φ13.5 × 20mm samples along a 2/3 R circumferential line there, and homogenize them (1135℃ for 40 hours, cool to 1000℃ at a cooling rate of 50℃ / h for 0.5 hours, then Furnace cooled to 600℃ (air-cooled) and processed into a φ12×18mm thermal compression sample. In the thermal simulation test, the deformation temperatures were selected as 1060℃, 1080℃, 1100℃, 1120℃, 1140℃, and 1160℃, and the deformation amounts were 20%, 30%, 40%, 50%, and 60%, and the deformation rates were 0.05, 0.1, and 0.5 (s\(^{-1}\)), as resumed in Table 2.

| Table 1. Chemical composition of GH4742 superalloy (wt.%)|  |
|----------------------------------------------------------|---|
| Al | Co | Cr | W | Mo | Ti | Nb | Ni |
| 2.54 | 10.22 | 14.15 | 0.2 | 5.12 | 2.52 | 2.73 | balance |

| Table 2. Experimental procedure and parameters |
|------------------------------------------------|
| Homogenization treatment | 1135℃ * 40 h →Cooled at 50℃ / h cooling rate→1000℃ |
| Temperature | 1060℃, 1080℃, 1100℃, 1120℃, 1140℃, 1160℃ |
| Strain rate | 0.05s\(^{-1}\), 0.1s\(^{-1}\), 0.5s\(^{-1}\) |
| Deformation | 20%, 30%, 40%, 50%, 60% |

The effect of hot deformation conditions on the recrystallization of the sample was achieved through OM observation. The metallographic sample was cut along the compression axial line of the
thermally compressed sample. After the sample was mechanically ground and polished, 50 ml of water + 40 ml of hydrochloric acid + 10 ml of hydrogen were used. The chemical solution of a mixed solution of hydrofluoric acid and 2 drops of hydrogen peroxide was observed by a LEICA MEF4A optical metallographic microscope.

Cylindrical specimens, 8 mm in diameter and 12 mm in height, were machined by EDM (electro discharge machining) and for slow cooling treatment. The specimens were solutioned at 1135°C for 10 min in order to dissolve all γ' precipitates into the matrix. Then, they were cooled down with a rate 5°C/h and 50°C/h to 1000°C, furnace cooling (FC) to 600°C and then air cooling (AC) and directly air cooling from 1135°C to room-temperature, respectively.

The effect of slow cooling treatment on the precipitation behavior of the phase of GH4742 alloy was achieved by SEM observation. After mechanically grinding and polishing the sample after hot compression, it was electrolytically corroded with a mixed solution of 12 ml phosphoric acid+40 ml nitric acid+48 ml sulfuric acid. By JEOL JSM -6480LV scanning electron microscope for observation.

3. Results and discussion

3.1. True stress-true strain curves

The common feature of the true stress-true strain curves is that, with the increase of strain, the stress increases rapidly at the beginning and the deformation intensifies significantly. When the stress reaches a certain value, the deformation resistance reaches the maximum, which is called the peak stress. Later, with the increase of strain, the true stress gradually decreases and finally reaches a relatively stable value, which is called the steady-state rheological stress. Obviously, at the beginning of the deformation of the sample, mainly deformation strengthening, leading to the increase of the rheological resistance, with the increase of deformation, dynamic recovery and dynamic recrystallization gradually increased, at a certain deformation, the dynamic balance of the two changes resulting in the basic unchanged rheological resistance. Figure 1a shows the true stress-true strain curves of the thermal compression test of GH4742 alloy under different heating temperatures with the same deformation rate (both 0.5 s⁻¹). It can be seen that under the same engineering strain, the deformation resistance increases with the decrease of deformation temperature, which is especially significant when the temperature is low (such as 1100,1080 and 1060°C). Figure 1b shows the true stress-true strain curve of the thermal compression test under the same heating temperature (both 1120°C) and different strain rates. It can be seen that, under the same engineering strain, the deformation resistance increases with the increase of deformation rate.

![Figure 1](image.png)

Figure 1. True stress-true strain curves of GH4742 alloy at (a) various deformation temperatures at train rate 0.5 s⁻¹, (b) various strain rates at 1120°C.

The phenomenon that the metal’s strength increases due to plastic deformation is called deformation strengthening, and deformation strengthening can make the metal produce even plastic
deformation. The essence of polycrystalline Ni-based alloy deformation is multi-system slip, so as to satisfy the coordination of deformation among various grains. In linear strengthening stage of Ni-based alloy, an edge dislocation sliding is blocked, the screw dislocation will change the sliding direction for avoid obstacles, a certain number of screw dislocations will disappear each other through climbing together, it provides convenience for dislocation movement. Therefore, for Ni-based alloys with low stacking-fault energy, the linear strengthening stage is very short, and its deformation curve is mainly parabolic strengthening stage. In Ni-based alloys, in order to realize the cross slip of screw dislocation, a series splits of dislocations line must be formed. The thermal activation of atoms under the stress action contributes to this process. In case increase the strain rate, the interaction of γ strengthening phase and carbides is aggravated leads to increasing of deformation strength effect.

3.2. Hot deformation constitutive relationship
The flow stress of a material during thermal deformation is the most direct response to various physical changes during the material’s thermal deformation process. It is the basis and key to study the metal flow law and determine the microstructure after deformation. Therefore, the selected constitutive relationship must reflect the main structural changes during the hot deformation process, include the work hardening and dynamic softening behavior during the deformation process. Flow resistant force of thermal deformation is generally closely related to the characteristics of the material itself and parameters such as deformation temperature, deformation rate and amount of deformation. Based on the true stress-true strain curves of the GH4742 alloy obtained from the isothermal hot compression test (see Figure 1, an exponential constitutive relationship is established. Since ln(σ) and ln (ε̇), ln(σ) and 1/T are linear relationship (see Figure 2 and Figure 3), the slope of the straight line is the value of n and Q / R. The least square method can be used to obtain the alloy Construct the equation as

\[ Z = 1.296 \times 10^{-4} \cdot \varepsilon^{0.142} \cdot \exp\left(\frac{42070}{T}\right) \]  

(1)

![Figure 2. Relationship of flow stress and strain rate of GH4742.](image1)

![Figure 3. Relationship of flow stress and temperature of GH4742.](image2)

The temperature sensitivity parameter and the strain rate sensitivity parameter are shown in Figure 4 and Figure 5. It can be seen that the change trend of the flow stress of GH4742 alloy with thermal parameters is as follows:

After the true strain exceeds the peak strain, as the strain increases, the temperature sensitivity and strain rate sensitivity of the flow stress are reduced to varying degrees. After the strain exceeds the steady state strain, η and ξ have nothing to do with the strain. According to related theories of metal
physics, the changing trend of flow stress is the result of the movement and development of internal dislocations during metal deformation. When the metal is deformed above the recrystallization temperature, two processes of strengthening and softening occur simultaneously inside the material. When the strain is small, the softening process is mainly achieved by dynamic recovery. In essence, dynamic recovery can be understood as the way to improve the integrity of deformed metal by redistributing point defects and reducing the dislocation rearrangement caused by conservative slip that does not form new grain boundaries. Therefore, in the dynamic recovery process, the elimination ability of dislocations and point defects inside the material is limited, but the propagation speed of dislocations are closely related to the deformation temperature and strain rate. In other words, changes in temperature and strain rate will cause significant changes in the number, type and distribution of dislocations inside the material, which will cause the flow stress to be sensitive to temperature and strain rate. As the degree of deformation increases, the softening process is continuously increased. When the deformation exceeds the peak strain, in addition to dynamic recovery, dynamic recrystallization also participates in the softening process, which greatly improves the movement and elimination of internal dislocations. Therefore, changes in temperature and strain rate change the dislocation proliferation rate and dislocation density, and also greatly change the dislocation elimination ability. In summary, the changes in the number and distribution of dislocations caused by changes in temperature and strain rate are not greatly in small deformations, so the temperature and strain rate sensitivity of flow stress decreases. When the strain exceeds the steady state strain, the strengthening and softening processes within the material have reached a dynamic equilibrium. At this time, the increase in strain has basically no effect on the number and distribution of dislocations. Therefore, the parameters \( \eta \) and \( \xi \) have nothing to do with strain.

![Figure 4. Relationship of temperature sensitivity parameter and strain rate.](image)

![Figure 5. Relationship of strain rate sensitivity parameter and temperature.](image)

When the deformation temperature is low, the flow stress is more sensitive to the strain rate. As the temperature increases, the sensitivity of the strain rate decreases. Since both recovery and recrystallization are typical thermal activation processes, these processes cannot be fully performed at lower temperatures. When the deformation temperature is lower, dislocation mobility is low, the dislocation density is small, and dislocation cross slip and diffusion climbing pairs are blocked due to the existence of second phase particles in the matrix. Therefore, when the temperature is low, the softening mechanism is not enough to balance the change in the number of dislocations caused by the change in strain rate, which causes the material flow stress to fluctuate greatly with the change in strain rate. As the deformation temperature increases, the thermal activation process becomes more active, and the second phase in the matrix gradually dissolves. When the temperature is enough higher,
the pinning effect of the second phase particles on the dislocations is released. Therefore, under this deformation condition that the dislocations are easy to move when the temperature is higher. At this time, the number and morphological changes of dislocations caused by the change of strain rate can be easily eliminated through different mechanisms, which reduces the strain rate sensitivity of flow stress.

When the strain rate is high, the flow stress is more sensitive to temperature. With the increase of the strain rate, the deformation process is shortened, and the dislocation cross slip and climbing processes are too late to fully proceed. The increasing of temperature must be to increase the dislocation mobility or start a new dislocation motion method. Therefore, the effect of temperature on the flow stress is very significant. When hot working is carried out using high strain rate methods such as forging on the hammer, the flow stress is more sensitive to temperature, it is beneficial to reduce the load under the premise conditions of ensuring the forging quality. When using low speed equipment such as hydraulic presses for forging, the optional deformation temperature range is wider.

3.3. Dynamic recrystallization and microstructure evolution
Dynamic recrystallization is a physical process, when temperature below the static recrystallization temperature, deformation can provide the driving force for recrystallization. It can be seen that thermal deformation process parameters (including deformation temperature, strain rate, strain) have an impact on the dynamic recrystallization process of materials. It can be seen that the thermal deformation process parameters (including deformation temperature, deformation rate, and deformation amount) all have an effect on the dynamic recrystallization process of the material. Both the increase in deformation temperature and the amount of deformation will lead to an increase in storage energy, thereby promoting the occurrence and completion of dynamic recrystallization. Therefore, as the deformation temperature increases, the amount of deformation required to start recrystallization will decrease.

Figure 6 shows the influence of deformation temperature on dynamic recrystallization of compressing samples of GH4742 alloy. It can be seen that under the same engineering strain, the higher of deformation temperature, the recrystallization of samples is more fully.

![Figure 6](image)

**Figure 6.** Photographs showing the dynamic recrystallization under engineering strain of 20%, various temperature of (a)1060, (b)1080, (c)1100, (d)1120, (e)1140, (f)1160℃ test conditions.
Recrystallization is a spontaneous process with reduced free energy, including two processes: nucleation and growth. The occurrence of dynamic recrystallization depends on whether the deformation stored energy in the compressed sample can provide the energy required for dislocation actuation. Obviously, the higher temperature accelerates the dislocation climbing and grain boundary migration, which is beneficial to the recrystallization nucleation and grain growth. The larger strain not only provides higher deformation energy, but also causes a temperature rise in the compressed sample, which is conducive to the occurrence of dynamic recrystallization. When the deformation rate is large, the softening is insufficient during compression, and the deformation strengthening is dominant, so that the flow stress in the compression process is greatly increased. The flow stress is beneficial to the activation of more dislocation sources and the proliferation of dislocations.

Through observing the microstructure of the compressed samples with different deformation amounts and different deformation temperatures, it can be seen that at a certain deformation temperature, as the amount of deformation increases, recrystallization easily occurs, and the recrystallized grain size is small. The amount of deformation is the same, as the deformation temperature increases, recrystallization will occur more easily, and the recrystallization grain size will become larger. In addition, it can also be seen that dynamic recrystallization always occurs first at the grain boundaries, which is because the energy of the grain boundaries is higher, which provides the energy that the recrystallization nucleation needs.

The dynamic recrystallization process is accomplished by nucleation and growth of recrystallized grains. The driving force is the deformation energy stored in the thermal deformation process. The newly formed recrystallized grains release this storage energy. The movement of grain boundaries is the main mechanism for the growth of recrystallized grains. The influence of deformation temperature and deformation amount on recrystallization can be obtained from the kinetic analysis of recrystallization process. In summary, high deformation temperature, large amount of deformation, and high deformation rate are all conducive to the occurrence of dynamic recrystallization during the deformation process.

3.4. Morphology of $\gamma'$ phase after slow cooling

The high alloying degree of GH4742 makes the initial melting point decrease, and the total dissolution temperature and dynamic recrystallization temperature of GH4742 alloy increase, resulting in a narrow hot working temperature range (about 80 ℃). When the alloy deforms in the single phase region, it has low flow stress, and the microstructure evolution is controlled by the migration of recrystallization grain boundaries. It is easy to obtain equiaxed crystal structure with sufficient dynamic recrystallization. When the alloy is deformed in the $\gamma+\gamma'$ two-phase region, the flow stress of the alloy increases greatly with the decrease of deformation temperature, and the dynamic recrystallization is inhibited by the precipitation of $\gamma'$ phase. Only when the deformation amount is large enough, the dynamic recrystallization transformation can be realized. In actual production, the temperature drop of the material during forging is inevitable, so it is difficult to ensure the deformation in the single-phase region. Therefore, the precipitation behavior of $\gamma'$ phase during thermal deformation becomes the key factor determining forging quality.

According to reference [5], Wang L.T. (1999) indicated that the annealing treatment method can change the grain size and morphology of $\gamma'$ phase. The classical $\gamma'$ phase precipitation theory holds that the size and precipitation rate of $\gamma'$ phase are mainly affected by supersaturation and atomic diffusion rate during cooling. At a faster cooling rate, on the one hand, the supersaturation of the alloy will increase rapidly, which is conducive to the rapid precipitation of $\gamma'$ phase; On the other hand, faster cooling rate will inhibit atomic diffusion and make the primary precipitation smaller. However, at a slower cooling rate, the growth rate of supersaturation of the alloy will slow down, while the diffusion rate will increase, and the size of primary $\gamma'$ phase precipitated in the corresponding alloy will increase. During continuous cooling, the alloy may also precipitate fine secondary $\gamma'$ phase or even finer tertiary $\gamma'$ phase due to secondary saturation. During high temperature slow cooling process,
a large number of coarse $\gamma'$ phases are formed, which can reduce the content of solid solution strengthening elements in the matrix. At the same time, the aging strengthening effect of $\gamma'$ phases will be weakened with the increase of the $\gamma'$ phases size. The reduction of the solution strengthening effect of the matrix and the weakening of the aging strengthening effect reduce the thermal deformation resistance of the hard-to-deform Ni-based alloy.

The $\gamma'$ phase in the cast structure of GH4742 superalloy was obvious change by the slow cooling treatment in the $\gamma+\gamma'$ two phases region after solution treatment. Coarse sphere-type, non-rule petal-type and fan-type $\gamma'$ phases were precipitated. Under different cooling conditions, the morphologies of $\gamma'$ phases which control cooling treatment from solution temperature 1135°C to 1000°C were completely different. The average size of dendritic-type $\gamma'$ phase was 1 $\mu$m at cooling rate of 50°C/h and 5 $\mu$m at cooling rate of 5°C/h, as shown in Figure 7. Morphologies, size and distribution of the $\gamma'$ strengthening phase were key microstructural factors for controlling flow stress and hot deformation plasticity within superalloy. Conversely, the $\gamma'$ particles growth in slowly cooling could consume completely the $\gamma$ matrix supersaturation during the subsequent cooling. However, the additional nucleation was limited.

![Figure 7. SEM photographs showing the microstructure after slow cooling treatment of (a) 50°C/h, (b) 5°C/h of the GH4742 alloy.](image)

Therefore, diffuse rate of the aging strengthening elements such as Al, Ti was the main factor for controlling growth of $\gamma'$ particles. Obviously, the further growth of $\gamma'$ particles could be accelerated by slow cooling treatment. Big size dendritic-type $\gamma'$ phase effectively decreases strength of GH4742 superalloy. This conclusion was documented by micro-hardness test results that the HV 397 at cooling rate of 50°C/h and HV 315 at cooling rate of 5°C/h. At reference [6], Lu X. D. (2009) indicated that the slow cooling treatment after solution can be obviously decrease the strengthening effect of precipitates. Obviously, the large-curvature curved grain boundary is caused by a fan-shaped structure, so the formation mechanism of this large-curvature curved grain boundary can be obtained by studying the formation mechanism of the fan-shaped structure. This fan-shaped structure is characterized in that the inside is composed of a large number of parallel rod-shaped $\gamma$phases, and the ends are grain boundaries. Zener [15] (1946) and H.L. Danful [16] (1996) believe that this characteristic organization is the product of a discontinuous phase change. As we all know, discontinuity refers to discontinuous mutations in the chemical composition and crystallographic orientation on both sides of the interface between the supersaturated substrate and the dissolution zone, and this interface is often incoherent. Discontinuous phase transition always originates at the grain boundary, because the grain boundary is a disordered, high-energy thin layer, and the energy that needs to be overcome discontinuous desolvation is relatively low. Therefore, this fan-shaped structure is easy to appear when the grain boundary nucleation probability is relatively large, and the grain boundary diffusion coefficient is relatively large. The fan-shaped region generally grows only to one side of an adjacent grain and forms a coherent interface with another adjacent grain. This discontinuous phase transition occurs in a
The sub-solution temperature range, and its obviously feature is that the phase transition diffusion path is only the grain boundary layer. The increase of the fan-shaped region is controlled by grain boundary diffusion rather than lattice diffusion.

Figure 8 shows that the comparison of limited deformation under air cooling and slow cooling at 50℃/h two test conditions. As shown in Figure 8, max permissive deformation of air cooling example is 40%, max permissive deformation of slow cooling example is 50%. Therefore, effect of slow cooling treatment on increasing the hot deformation plasticity for GH4742 superalloy is notable.

4. Conclusions
1) In this paper, the hot deformation behavior of a Ni-based superalloy were studied in the temperature range of 1060-1160℃ and strain rate range of 0.05-0.5 s⁻¹ using hot compression tests. Results of hot compression tests were indicated that the hot deformation is sensitive to temperature and strain rate.

2) In order to reveal the effects of various hot deformation process parameters on the forging process of hard-to-deform alloys and to effectively control the quality of forgings, the influence of the microstructure evolution process must be fully considered. Through systematic study of the hot deformation behavior of the hard-to-deform GH4742 alloy, established the microstructure evolution model of GH4742 alloy.

3) By slow cooling pre-processing after homogenizing, a structure composed of coarse γ' phase and an γ'-matrix with a low degree of solid solution can be obtained, and the hot workability of the alloy is effectively improved. The slow cooling treatment can promote the occurrence of recrystallization during thermal deformation, and the formation of curved grain boundaries can reduce the cracking tendency during deformation and improve the thermoplasticity.

Acknowledgments
The research was supported by National important base research and development program of China (Contract No. 2010CB631203).

References
[1] Zhong Z Y and Zhuang J Y 2003 Iron & Steel Research. 15 1-9
[2] Long Z D, Zhuang J Y, Deng B and Zhong Z Y 1999 Acta Metall. Sin. 35 114-6
[3] Zhang B J, Zhao G P, Xu G H and Feng D 2005 Acta Metall. Sin. 41 1207-14
[4] Wang L T, Bian F, Long Z D, Deng Q, Zhuang J Y and Zhong Z Y 1999 Acta Metall. Sin. 35 103-5
[5] Long Z D 1999 Doctor's academic degree thesis of Central Iron & Steel Research Institute, Beijing
[6] Lu X D, Deng Q, Du J H, Qu J L, Zhuang J Y and Zhong Z Y 2009 J. Alloys Comp. 477 100-3
[7] Lu X D, Du J H, Deng Q and Zhong Z Y 2009 J. Alloys Comp. 486 195-8
[8] Wang L, Xie G, Lou L H 2013 Mater. Let. 109 154-7
[9] Azarbarmas M, Aghaie K M, Cabrera J M and Calvo J 2016 Mater. Sci. Eng. A678 137-52
[10] Huang J and Zhou X 2006 Mater. Let. 60 1854-8
[11] Wang Y, Shao W Z, Zhen L and Zhang XM 2008 Mater. Sci. Eng. A486 321-32
[12] Liu W C and Radhakrishnan B 2010 Mater. Let. 64 1829-32
[13] Stewart G R, Elwazri A M, Yue S and Jones JI 2006 Mater. Sci. Tec. 22 519-24
[14] Karamis M B and Halici 2007 Materials Let. 61 944-48
[15] Zener C 1946 Trans. Am. Inst. Min. Eng. 167 550-72
[16] Danflou H L, Macia M, Sanders T H and Khan T 1996 Superalloys 1996 119-25