Effect of samarium on the high temperature tensile properties and fracture behaviors of Al–Zn–Mg–Cu–Zr alloy

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Abstract
The Al–6.7Zn–2.6Mg–2.0Cu–0.1Zr alloy is modified by Sm for the purpose of improving its mechanical properties at high temperatures. According to the results, the as-cast structure can be refined through 0.3% Sm modification treatment, and finer precipitates can be obtained after aging. In the meantime, with the addition of Sm, a new high melting point Al\(_{10}\)Cu\(_3\)Sm\(_2\) phase is formed, the initial decomposition temperature of the alloy is raised, and the thermal stability is enhanced. As indicated by the tensile test results at room temperature and high temperatures, the strength of the modified sample was improved compared with the unmodified alloy. According to fracture surface analysis, ductile fracture conforms to the fracture characteristic exhibited by the modified alloy at room temperature and high temperatures. The improvement on high temperature strength of the modified alloy is attributed largely to a combination of grain refinement, precipitation strengthening and heat resistant phase to grain boundary pinning.

1. Introduction
Al–Zn–Mg–Cu–Zr alloy is regarded as one of the most significant materials in the aerospace industry. Due to such advantages as high modulus, low density and excellent stress corrosion resistance, it has been commonly used in the manufacturing of aircraft components [1–3]. As modern industry has placed increasingly higher requirements on materials, it is imperative to improve the overall properties of Al–Zn–Mg–Cu–Zr alloy. As a structural component of the fuselage, the temperature at which Al alloy is put into temporary service usually ranges from 356 to 450 K [4, 5]. In this case, it is of much practical significance to improve the high temperature performance of the alloy.

In Al–Zn–Mg–Cu–Zr, the high temperature hardening phase is dominated by Al\(_3\)Zr, which shows a consistent relationship with the matrix and is in a thermally stable phase [6, 7]. This phase can still hinder dislocation migration at high temperature and increase the strength [8, 9]. Therefore, in order to further improve the high temperature strength of Al–Zn–Mg–Cu–Zr alloys, it is essential to determine an appropriate microstructure, which may be comprised of precipitates with a better thermal stability and a stronger resistance to coarsening [10]. It can also improve high temperature performance by forming fine heat resistant phases. In recent years, there have been plenty of scholars conducting a study on the high temperature properties of Al alloys with rare earth elements contained. Research carried out by Colombo et al [11] demonstrated that the eutectic Si morphology of Al–Si–Mg alloy as modified by Er was effectively improved. In addition, needle-like Er-rich intermetallic compounds were observed in the interdendritic region, which is conducive to increasing the mechanical strength of the alloy at 423 K and 473 K. Khorshidi et al [12] found out that an appropriate amount of Gd is contributory to reducing the volume fraction of Mg\(_2\)Si particles and generating a layered MgGd intermetallic compound, thus improving the high temperature tensile properties of Al–15Mg-Si in situ composites. As demonstrated by Mondol et al [13], Sc element can lead to Al\(_3\)Sc or Al\(_3\)(Sc, Zr) dispersed phase with a higher level of thermal stability in 2219 alloy, thus enhancing the performance of the alloy in heat resistance.
In recent years, the effects of rare earths on the structure and mechanical properties of Al–Zn–Mg–Cu–Zr alloys have attracted extensive attention for study [14–17]. Rare earth is capable to refine grains and inhibit recrystallization, thus improving the overall performance of the alloy. As a relatively cheap rare earth element, Sm can also improve the microstructure and performance of Al alloys [18, 19]. Up to now, however, the effect of rare earth Sm modification on the microstructure and mechanical properties of Al–Zn–Mg–Cu–Zr alloys is rarely studied, and more attention is needed. The fracture behaviour at room temperature and high temperatures has yet to be studied, which is of much significance to the structural applications of Al–Zn–Mg–Cu–Zr alloys. Therefore, the primary purpose of this study is to figure out the effect of Sm on microstructure and the high temperature mechanical properties of Al–Zn–Mg–Cu–Zr alloy.

2. Materials and methods

Pure Al, Mg, Zn and master alloy Al-50%Cu, Al-10%Zr were used as raw materials to prepare Al-6.7Zn-2.6Mg-2.0Cu-0.1Zr ingots (the element content is wt%). Sm element was added in the form of Al-20%Sm. The alloy melting was carried out in a box-type resistance furnace at 1053 K, with C2Cl6 used to degas the melt. Afterwards, samples were ground and mechanically polished, before being etched using Keller’s reagent. The morphology and microstructure of the as-cast, hot extruded and heat treatment alloys were analyzed using optical microscopy (OM, XD30M), scanning electron microscopy (SEM, Apereo C) and scanning transmission electron microscopy (STEM, F200X, FEI) with an operating voltage of 200 kV. Differential scanning calorimetry and thermogravimetric analysis (DSC, TG, SDT Q600) were conducted to determine the decomposition temperature of the second phase and the weight loss of the alloy, with the heating rate set to 10 °C min⁻¹. Philips X’Pert x-ray diffractometer was applied to obtain x-ray diffraction pattern (XRD), and the test was conducted with Cu Kα radiation at a 4 ° min⁻¹ scan rate. The alloy phase composition was determined according to the JCPDS card. The grain size of the as-cast alloy is measured by linear intercept. At least 10 measurements on each OM micrographs. The final average grain size is obtained by measuring five OM micrographs. The MST-200 tensile testing machine is used to conduct high temperature instantaneous tensile experiments (in accordance with GB/T228.1-2010), with the sample intercept size set to Φ5 × 35 mm. The tensile fixture and the sample are heated in the CEC1200 type experiment box, and the K type thermocouple is used to test the experimental temperature. After each sample reached the experimental temperature, it was maintained for 10 min, before being stretched at a speed of 1 mm min⁻¹.

3. Results

3.1. Microstructural evolution

Figures 1(a)–(b) shows an optical micrograph of the cast alloy. It can be seen from the figure that the structure of the unmodified alloy is uneven (figure 1(a)). Besides, it can be known that the grain size reaches an average of 188 μm. The addition of 0.3% Sm is effective in improving the microstructure of the alloy and refining the grains, with an average grain size of 108 μm (figure 1(h)). According to the XRD results (figure 1(c)), the unmodified alloy is comprised of Al matrix and MgZn₂ phase. When 0.3% Sm is added, a new diffraction peak appears in the alloy. Compared with the JCPDS card, it is more consistent with the diffraction peak of Al10Cu7Sm2, suggesting that the alloy is comprised of Al matrix, MgZn₂ and Al10Cu7Sm2. In order to further determine the phase composition and changes of the alloy, an analysis was performed of SEM pictures and energy dispersive spectrometer (EDS) (figures 1(d)–(f)). As indicated by the results, the addition of Sm is effective in refining the eutectic phase (η phase) at the grain boundary [21]. Despite a small amount of Sm that dissolves into the η phases (in figure 1(e),(B)), the remaining Sm in a large majority develops into a rare earth phase at the grain boundary. EDS reveals that the atomic ratio of Al, Cu and Sm is 10:7:2. Combined with XRD, the rare earth phase is determined as Al10Cu7Sm2 phase (in figure 1(e),(C)). After homogenization and hot extrusion, the η phase disappeared from the alloy, but there remains an undissolved phase in the alloy (figures 1(g)–(h)). The remaining phase of the unmodified alloy is dominated by the Fe-rich phase [22, 23]. In comparison with the unmodified alloy, there are more remaining phases in the modified alloy. From the local element surface distribution, it can be seen that in addition to the Fe-rich phases as indicated by yellow circle in figure 1(i), there are plenty of rare earth phases (circled in purple in figure 1(i)) in the modified alloy, suggesting that the Al10Cu7Sm2 phase performs well in heat resistance.
The heat resistant particles near the grain boundary have an important influence on the high temperature performance of the alloy [12]. From the bright field image (BFI) and selected area electron diffraction (SAED) diagrams, it can be seen clearly that the phase with a larger size at the grain boundary of the modified alloy after aging is Al\textsubscript{10}Cu\textsubscript{7}Sm\textsubscript{2} phase with a size of 0.8–2.5 $\mu$m (figure 2). These are the remaining phases after solid solution.

In addition to the grain size and heat resistant particles that plays a significant role in the properties of the alloy, secondary precipitation is equally crucial to the properties of the alloy. Figure 3 shows the transmission electron microscope (TEM) and high resolution transmission electron microscope (HRTEM) images of the two alloys after aging. According to this figure, there are a large number of fine precipitates uniformly distributed in the crystal grains. The size of the precipitates in the Sm modified alloy is smaller than that in the unmodified alloy.

Figure 1. Microstructure and morphology of the alloy: (a), (b) as-cast optical micrographs of unmodified and alloy containing 0.3% Sm; (c) XRD curves of the as-cast alloys; (d), (e) and (f) SEM and EDS analysis of the alloy; (g) and (h) SEM images of unmodified and alloy containing 0.3% Sm after modification, (i) partial surface distribution map of (h).

Figure 2. BFI and SAED analysis chart of residual phase: (a) BFI; (b) SAED.
The unmodified alloy is comprised primarily of G.P. zones, \( \eta \) and \( \eta' \) phase. The precipitated phase after Sm modification is mainly composed of G.P. zones and \( \eta' \) phase. Analysis of the HRTEM diagram and the corresponding FFT and IFFT diagrams proved this point. The orientation relationship of \( \eta' \) phase in the unmodified alloy is \([12\bar{1}0]_{\eta'/\alpha}[1011]_{\alpha}\) (in figure 3(a)), while the \( \eta' \) phase orientation relationship is \([1000]_{\eta'/\alpha}[1011]_{\alpha}\) (in figure 3(b)). In the modified alloy, apart from the G.P. zones consistent with the aluminum matrix (in figure 3(d)), another \( \eta' \) phase was also observed. The orientation relationship is \([2\bar{4}10]_{\eta'/\alpha}[1011]_{\alpha}\) (in figure 3(d)).

In order to understand the precipitation behavior of Al-6.7Zn-2.6Mg-2.0Cu-0.3Sm alloy, the high-angle annular dark field (HAADF) diagram and distribution of elements of the precipitated phase was analyzed, as shown in figure 4. The low magnification HAADF photos show that there are intermittent precipitates at the grain boundaries (figures 4(a)–(b)). The main elements are Zn and Mg in \( \eta \) phases, as indicated by the yellow area in figures 4(a)–(b). Besides, the spherical Al\(_2\)Zr precipitated during the homogenization process (green area in figures 4(a)–(b)) and the S(Al\(_2\)CuMg) phase transformed from the eutectic phase during this process remain (white area in figures 4(a)–(b)). These are the high temperature strengthening phases of the alloy [22, 23]. Sm element did not form precipitation phases similar to Al\(_2\)Sc and Al\(_2\)Er, which are different from the influence of Sc and Er on the precipitation behavior of Al–Zn–Mg–Cu–Zr alloy [13]. Figures 4(c)–(d) is a high magnification HAADF photo of the precipitated phase inside the grain. The size of the precipitate is 5–10 nm, mainly containing Zn, Cu and Mg elements. The Zr element is evenly distributed inside the crystal grains in the field of view. The content of Sm solid-dissolved in the crystal grains is small, and most of it exists in the precipitated phase (marked in red in figures 4(c)–(d)).

### 3.2. Thermal analysis

Figure 5 shows the differential scanning calorimetry (DSC) and thermogravimetric analysis (TG) curves of the as-cast alloys in an argon atmosphere. As indicated by the DSC curve, the two alloys exhibit a single endothermic peak below 900 K, to be precise, 748.9 K, which represents the decomposition temperature of \( \eta \) phases. The melting temperature of the unmodified alloy aluminum matrix is 901.28 K, while the melting temperature of Sm modified alloy aluminum matrix is 901.90 K, which means a slight increase of 0.62 K. In addition to the aforementioned two endothermic peaks, the modified alloy shows an endothermic peak at 937.67 K. According to XRD, EDS and TEM analysis, it can be judged as the endothermic peak of Al\(_{10}\)Cu\(_7\)Sm\(_2\) phase. According to the TG curve, the initial temperature at which weight reduction occurs in the unmodified alloy is 897.75 K, while the initial decomposition temperature is 923.34 K for the Sm modified alloy, which means an increase of 25.59 K. Moreover, the temperature ranges between 897.75 K and 966.09 K, while the weight loss of Sm modified alloy is
less significant compared with the unmodified alloy. In other words, the addition of Sm element forms Al$_{10}$Cu$_7$Sm$_2$ phase with better heat resistance, which enhances the high temperature stability of the alloy.

### 3.3. Tensile test

Due to the working environment, 7xxx Al alloy is required to demonstrate excellent high temperature mechanical properties. Figure 6 shows the dimensions of the tensile sample and typical engineering stress-strain curve. As shown in figure 6(b), the tensile stress-strain curve is comprised of three different stages: elastic stage, uniform deformation stage and local deformation stage. In the elastic phase, stress and strain show a linear
increase. Then, work hardening occurs due to the cumulative dislocation in the uniform deformation stage. With the strain further increase, necking occurs progressively after the stress reaches its maximum.

Figure 7 shows the engineering stress-strain curves of the two alloys at different temperatures (ranging between 298 K and 573 K). When the temperature varies, the stress-strain relationship of the samples show a significant difference. Uniform elongation (marked with a solid square on the curve) controls the start of local deformation, as determined by the Considère criterion [28]:

$$\left( \frac{\partial \sigma}{\partial \varepsilon} \right)_f = \sigma$$

(1)

where $\sigma$ represents the true stress and $\varepsilon$ refers to the true strain. In figure 7(a), the uniform elongation of the unmodified alloy sample at room temperature is 14.63%, while the uniform elongation of Sm is increased to 17.89% after modification. As the temperature increases, the uniform elongation of the two alloys continues to decrease. The difference in uniform elongation between unmodified and modified alloys is shown to be significant (figures 7(a)–(c)). When the stretching temperature reaches 473 K, the uniform elongation stands at 8.23% and 8.40%, respectively (figure 7(d)). Besides, the uniform elongation of the two alloys shows similarity. The stretching temperature continues to increase, while the uniform elongation shows no significant change (figures 7(e)–(f)).

The data on ultimate tensile strength (UTS) at all tensile temperatures is presented in figure 8(a). Under the tensile temperatures of 298 K, 373 K, 423 K, 473 K, 523 K and 573 K, the UTS of the unmodified alloy is 704, 609, 535, 447, 290 and 131 MPa, respectively. While the modified alloy is 732, 641, 573, 487, 316, 141 MPa, respectively. Compared with the unmodified alloy, the degrees of increase varies. In addition, it can be seen that the UTS of both alloys decreases with the increase of temperature, suggesting that the tensile properties of the alloy are highly sensitive to temperature. Moreover, elongation first shows increase with the rise of temperature, while it fluctuates after the temperature reaches 473 K. This is related to the re-dissolution of the precipitated phase [29].

The sensitivity of materials to temperature can be expressed by the value of strain hardening index ($n$), that is, Holloman formula [30]:

$$\sigma = K \varepsilon^n$$

(2)

Where $K$ is a constant, $n$ indicates the strain hardening ability of the alloy, and the strain hardening ability of the alloy improves with the rise in the value of $n$. The value of $n$ can be calculated using formula (2) [30]:

$$n = \frac{\partial \ln \sigma}{\partial \ln \varepsilon}$$

(3)

In figure 8(b), the value $n$ of the two alloys relative to the tensile test temperature is calculated on the basis of true stress-strain curve. It can be seen clearly that Sm is effective in increasing the value $n$ of the alloy at high temperatures (423 K–573 K) after modification. In addition, the test temperature rose from RT to 573 K, and the $n$ value dropped rapidly, which indicates that strain softening occurs at higher temperatures. The reductions of strain hardening rate and thermal softening is contributory to the decrease of value $n$ after the tensile test temperature of the two alloys increases [31].

3.4. Fractography

Figure 9 shows the morphology of tensile test fracture occurring to the two alloy at different temperatures. At room temperature (RT), the fracture surface of the unmodified alloy is shown to be in a mixed ductile-brittle
Figure 7. Stress-strain curves of alloys at different test temperatures: (a) 298 K; (b) 373 K; (c) 423 K; (d) 473 K; (e) 523 K; (f) 573 K.

Figure 8. Statistical diagram of the UTS and strain hardening index n of the alloy: (a) UTS; (b) n.
mode, with a large number of flat cleavage surfaces and a few dimples observed. In comparison, the number of cleavage surfaces of the Sm modified alloy is smaller but that of the dimples is larger, with some small dimples observed on the cleavage surface. Besides, the alloy manifests ductile fracture from an overall perspective. At 423 K, the number of dimples on the tensile fracture surface increases significantly in comparison with RT. Cleavage surface of the unmodified alloy shows small dimples as well. The fracture surface of the metamorphic alloy consists mainly of dimples and tearing edges with small dimples. At this time, the level of plasticity is higher than that at RT. With a 473 K temperature rise, the dimples on the fracture surface of the two alloys and those on the tearing edge became larger, the dimples of the modified alloy become smaller than the unmodified alloy, and there are fine heat-resistant phases observed in the dimples. When the temperature reaches 573 K, the dimples in the alloy expand substantially, while those on the tear edges of the two alloys disappear. Compared with the unmodified alloy, more heat-resistant phases were found in the dimples of the Sm modified alloy, and their size is relatively small. As indicated by the EDS analysis, the atomic content of this phase is consistent with $\text{Al}_{10}\text{Cu}_{7}\text{Sm}_{2}$.

4. Discussion

4.1. Refinement mechanism of grains and precipitates

According to the aforementioned analysis, compared with the unmodified alloy, the 0.3% Sm modified alloy contains finer grains. Besides, after aging treatment, the size of the 0.3% Sm modified alloy in precipitated phase is smaller but the number is larger. At the same time, Sm inhibits the transition from $\eta$ phase to $\eta'$ phase. Next, our attention is shifted to these transformation mechanisms.

4.1.1. Analysis on the mechanism of Sm grain refinement

Since the equilibrium distribution coefficient $K$ of rare earth elements is less than 1, during the solidification process, a large number of Sm atoms are enriched at the solid/liquid interface, resulting in a significant increase in the Sm concentration in the liquid alloy at the front end of the interface. Consequently, constitutional supercooling would occur, thus improving the ability of grain nucleation and refining the grains. In the meantime, the $\text{Al}_{10}\text{Cu}_{7}\text{Sm}_{2}$ phases formed by the Sm atoms at the grain boundaries hinder the further growth of the crystal grains while refining the crystal grains. Moreover, the Sm atoms aggregating at the front end of the solid/liquid interface prevent Zn, Mg and Cu from entering the solid solution at the grain boundary, which is...
effective in inhibiting the growth of the eutectic phase [16]. Thus, the addition of Sm can refine the eutectic phase at the grain boundary.

4.1.2. Effect of Sm on the precipitation behavior of alloy
The mechanism by which Sm element enhance the precipitation of the \( \eta' \) phase involves two aspects. One is that the atomic diameter of Sm (0.180 nm) is considerably larger than that of Al (0.143 nm). Sm atoms cause lattice distortion and supersaturated vacancies to aggregate when entering the Al matrix [19]. Vacancy groups play the role as nucleation site for precipitation, thus accelerating the precipitation of the \( \eta' \) phase. The other is the interaction between Sm atoms and vacancies. Due to the intense interaction between the two, Sm atoms and vacancies are combined to generate Sm-vacancy pairs, which shows a lower fluidity than free vacancies and thus contributes to nucleation points in the matrix during the aging process [33]. The small amount of Sm in the precipitated phase in figure 4 also proves this.

Compared with Mg and Zn, vacancy binding energy of Sm is slightly reduced [16]. As a result of the Sm atoms are preferentially attached to the vacancy, thus isolating the vacancy from Mg and Zn. This preferential combination of vacancies and Sm produces a significant inhibitory effect on the movement of Mg and Zn atoms, thus causing delay to the growth of \( \eta' \) phase. Meanwhile, the transition from \( \eta' \) phase to \( \eta \) phase is slowed down. Therefore, no \( \eta \) phase can be observed in the modified alloy.

4.2. Strengthening mechanism of tensile properties
The strengthening mechanism of Al–Zn–Mg–Cu–Zr alloy mainly involves: (1) solid solution hardening, (2) precipitation strengthening, (3) fine grain size or grain boundary strengthening, (4) dislocation strengthening [34–36]. In this study, firstly, due to the difference in atomic radius and Young’s modulus, Sm atoms fail to match the matrix. The strain around these atoms will impede the movement of dislocations, thus improving the properties of the alloy. Nevertheless, the solid solubility of Sm in Al alloys stays at a low level, as a result of which the contribution of solid solution hardening to the overall level of strength is insignificant [15, 16, 19]. Secondly, the G.P. zones and \( \eta' \) phase represent the main precipitation phases for strengthening the 7xxx alloys, while the \( \eta \) phase shows a relatively limited increase in strength. Additionally, the increasing density of precipitates contributes more significantly to the improvement of strength [1]. The addition of Sm to the G.P. zones and the size of the \( \eta' \) phase are smaller and more numerous. In the meantime, Sm suppresses the transition from \( \eta' \) phase to \( \eta \) phase. All of these lead to a significant improvement of the strength exhibited by the modified alloy. Thirdly, it can be known from the as-cast structure that the Sm modification is accompanied by a noticeable grain refinement. According to the Hall-Petch formula, the strength of the alloy is inversely proportional to its grain size [37]. Therefore, a significant grain refinement can increase the strength of the alloy considerably. Fourthly, with the addition of Sm element, the fine and dispersed G.P. zones and \( \eta' \) phase and the fine Al\(_{10}\)Cu\(_7\)Sm\(_2\) phase have a strong interaction with dislocations. This is also contributory to the improving the strength of alloy. In summary, the room temperature strength of the modified alloy is significantly improved due to the combined effect of precipitation strengthening, fine grain strengthening and dislocation strengthening.

The finer the grains at room temperature, the larger the grain boundary area, which is not conducive to crack propagation, so the higher the strength of the metal and the better the plasticity. At higher temperatures, the diffusion of grain boundaries may be significantly accelerated, which not only makes the effect of fine-grain strengthening disappear, but may also reduces the high temperature properties of the alloy [38, 39]. In this study, the grain size of the alloy containing 0.3% Sm decreases. The experimental temperature is between 293 K and 573 K, and its strength and plasticity are both increased compared with the unmodified alloy, as shown in figure 7. The reasons are as following:

First of all, it can be known from the thermal analysis (figure 5), the addition of Sm increases the initial decomposition temperature of the alloy, thereby improves the thermal stability of the matrix, which is conducive to obtaining better thermal strength of the alloy. Secondly, a significant factor in the improvement of high temperature strength lies in the precipitation strengthening of the modified alloy. At high temperatures, the \( \eta' \) phase begins to coarsen, thereby reducing its pinning ability to dislocations and grain boundaries [40]. From figure 3(a), it can be seen that the \( \eta' \) phase in the unmodified alloy after aging at 393 K for 24 h is larger, and even transformed into \( \eta \) phase. However, after 0.3% Sm modification, the alloy precipitated phase contains a small amount of Sm (figure 4). Compared with the unmodified alloy within the same aging time, these Sm-rich precipitated phases are smaller in size and larger in number (figure 3(b)). The addition of Sm improves the thermal stability of the precipitated phase. Besides, the increase in the number of precipitates may play a major role in hindering dislocations [41]. Furthermore, the improvement of high temperature strength is largely attributed to plenty of fine Al\(_{10}\)Cu\(_7\)Sm\(_2\) phase produced in the alloy after Sm modification (as shown in figure 2). This rare earth phase shows excellent heat resistance, and it is capable to pin grain boundaries at high temperatures, thus inhibiting the growth of crystal grains and the spread of cracks. Thence, it allows the
modified alloy to maintain a high level of strength at high temperatures. Sm improves the thermal stability of the matrix and the precipitated phase, as well as the high melting point Al10Cu7Sm2 for dislocation to grain boundary pinning. The combined effect of the two improves the high temperature strength of the alloy.

5. Conclusions

The effect of 0.3% Sm on the microstructure and high temperature tensile behavior of Al-Zn-Cu-Mg-Zr alloy was studied, which led to the results as summarized below.

1. After 0.3% Sm modification, the grains of the alloy were refined, thus forming a high melting point Al10Cu7Sm2 phase. The addition of Sm caused the initial decomposition temperature of the alloy to increase. After aging, the G.P. zone and η′ phase was refined.

2. With the increasing of tensile test temperature, the strain hardening index (n) value of unmodified and modified alloys declined due to thermal softening.

3. The tensile strength of unmodified and modified alloys decreased with temperature rise. The ultimate tensile strength of the modified alloy is higher than that of the unmodified alloy at every test temperature, which is because grain refinement, precipitation strengthening and Al10Cu7Sm2 phases are closely associated with grain boundary pinning.

4. η′ phase coarsening and thermal softening are mainly contributory to the strength of the alloy. As the temperature increasing, the fracture mode of the unmodified alloy shifts from ductile and brittle mixed with pure dimple fracture. Metamorphic alloys always maintain ductile fracture.

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