

## Research Progress on a Cast-Forged Alloy for High-Performance Aviation Turbine Disks (Post-print)

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### Abstract

To address the deficiencies of advanced cast-and-wrought superalloys for aero-engine turbine disks, such as difficult hot workability, temperature capability below 680 °C (630 MPa, 1000 h creep rupture life), and inability to meet the design requirements of modern aero-engines, a design concept was proposed to optimize the composition of precipitation strengthening phases and introduce microtwins to enhance the strength in the intermediate temperature range (service temperature range of the alloy), while simultaneously reducing the high-temperature strengthening effect of precipitates and the solution temperature to decrease the strength in the high temperature range (processing temperature range of the alloy). Based on this philosophy, a series of Ni-Co based cast-and-wrought superalloys (TMW alloys) were designed and fabricated. The fabrication and property evaluation of designed alloy disk blanks (diameter 440 mm, thickness 65 mm) demonstrated that TMW alloys can be processed via hot working on conventional melting-casting and forging equipment, with a temperature capability exceeding 700 °C, representing an improvement of more than 50 °C over U720Li, currently the strongest commercial cast-and-wrought turbine disk alloy. This paper briefly reviews the main research progress on this high-performance cast-and-wrought superalloy, covering aspects including the design philosophy, compositional characteristics, forgeability, microstructure control, property features, and deformation strengthening mechanisms of TMW alloys.

### Full Text

#### Preamble

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## Research Progress in a High Performance Cast & Wrought Superalloy for Turbine Disc Applications

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### Abstract

The advanced cast & wrought (C&W) superalloys for turbine disc applications have low temperature bearing capacity and poor processing ability compared with the powder metallurgy (PM) superalloys. Recently, a new kind of C&W Ni-Co base disc superalloys, named as TMW alloy, has been developed on the basis of an innovative concept. Full-scale pancakes (440 mm in diameter and 65 mm in thickness) with an average grain size of about 10  $\mu$ m were successfully manufactured by a conventional C&W processing route. TMW alloys can provide 58~76 °C temperature advantages over alloy U720Li in 0.2% creep strain life under 630 MPa and are comparable with the advanced PM disc superalloys. In this work, the alloy design, preparation processing capacity, microstructural control, performance characteristics and deformation strengthening mechanisms were briefly introduced.

**KEY WORDS** cast & wrought superalloy, TMW alloy, alloy design, microstructure control, deformation mechanism

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### Introduction

The aero-engine turbine disc is a critical hot-section component that anchors turbine blades, harnesses energy from combustion gases through the blades, and transmits this energy via a shaft to drive the fan, compressor, and engine rotation at high speed. With continuous demands for higher thrust-to-weight ratio (TWR) and lower specific fuel consumption (SFC) in high-performance

engines, the turbine inlet temperature (TIT) has increased from approximately 800 °C in the 1940s to nearly 1700 °C today, and is expected to exceed 1900 °C in future engines with a TWR of 20 [1,2]. Consequently, the temperature capability of turbine disc alloys has risen from 550 °C to the current 700 °C, with future expectations of exceeding 800 °C (Table 1 [1,2]).

Modern aero-engines impose several fundamental requirements on turbine disc alloys: (1) high heat resistance and microstructural stability; (2) high specific strength; (3) high creep resistance and rupture life; (4) good low cycle fatigue performance; (5) low fatigue crack growth rate; and (6) lower manufacturing cost. Due to the complex structure of turbine discs and the high rim loads (centrifugal loads from turbine blades), complex stresses (mechanical and thermal stresses), and temperature gradients experienced in high-temperature service environments (relative temperature and stress distributions at rim and hub during service are shown in Figure 1 [Figure 1: see original paper][3]), alloy design must simultaneously consider diverse performance requirements while balancing the relationships among properties, microstructure, and processing. As a result, development of disc alloys has progressed more slowly than that of turbine blade alloys.

Figure 2 [Figure 2: see original paper][4] summarizes the evolution of temperature capability (630 MPa, 1000 h rupture life) of superalloys for turbine discs since the 1950s [4]. Commercial cast & wrought (C&W) superalloys such as U720Li and In718Plus currently have temperature capabilities not exceeding 680 °C [5,6], while powder metallurgy (PM) superalloys such as ME3 (Rene104) and LSHR have capabilities up to 730 °C [7]. Notably, a new type of Ni-Co base cast & wrought superalloy (TMW alloy) recently developed by the National Institute for Materials Science (NIMS) in Japan achieves a temperature capability of 725 °C, approaching the level of the strongest commercial PM superalloys.

This paper briefly introduces the main research progress on this high-performance cast & wrought superalloy from the perspectives of design philosophy, compositional characteristics, forgeability, microstructural control, performance features, and deformation strengthening mechanisms.

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## 1. Alloy Design

Turbine disc alloys can be classified into two categories based on processing route: cast & wrought alloys and powder metallurgy alloys. Cast & wrought alloys offer simple processing and lower cost, but suffer from difficult deformation, poor microstructural uniformity, and limited comprehensive performance improvement. This is primarily because cast & wrought alloys require appropriate thermomechanical processing and subsequent heat treatment after ingot casting to achieve desirable microstructures and properties. However, excessively high volume fractions of precipitation strengthening phases (greater than 50%) or excessive solid solution strengthening elements lead to non-uniform

ingot microstructures, narrow processing windows, and difficult hot working. Consequently, the strengthening levels achievable in cast & wrought superalloy design are limited, and temperature capability cannot exceed 700 °C (630 MPa, 1000 h rupture life). Alloys with higher temperature capabilities currently require powder metallurgy processing.

To design and prepare cast & wrought superalloys with temperature capability exceeding 700 °C, the bottleneck of easy strengthening but difficult deformation must be addressed. Based on the different temperature characteristics of service and processing conditions for turbine disc alloys, an innovative alloy design concept was proposed: strengthening the “medium-temperature region (service temperature range)” while weakening the “high-temperature region (deformation processing range)” (Figure 3 [Figure 3: see original paper][8-11]), opening a new pathway for developing cast & wrought alloys surpassing 700 °C [8-11]. The designed alloy exhibits higher strength than conventional alloys at low and medium temperatures, but lower strength at high temperatures, as illustrated in Figure 3.

Currently, turbine disc alloys with relatively high temperature capabilities (U720Li, ME3, and LSHR, etc.) are all Ni-based alloys with a g/g' -Ni<sub>3</sub>Al two-phase microstructure, where the g phase is a face-centered cubic matrix and the g' phase is an ordered L1<sub>2</sub> precipitation strengthening phase. The g' phase contributes over 70% of the high-temperature strength. Therefore, optimizing the g' precipitation strengthening phase is an effective approach to realize this design concept. Since the g' phase exhibits anomalous yielding behavior in the 0~800 °C temperature range [12], many phases with A<sub>3</sub>B structure also show anomalous yielding, such as Co<sub>3</sub>Ti, Ni<sub>3</sub>Si, and Ni<sub>3</sub>Ge [13-15]. Among these, Co<sub>3</sub>Ti has an anomalous yield peak temperature approximately 100 °C higher than Ni<sub>3</sub>Al, but its strength is lower than Ni<sub>3</sub>Al above 1000 °C (Figure 4 [Figure 4: see original paper][13-15]), potentially achieving the alloy design target proposed herein.

The distribution of g/g' two-phase regions in the Ni-Co-Al-Ti quaternary phase diagram is shown in Figure 5 [Figure 5: see original paper][16]. Based on the composition characteristics of U720Li alloy (45% g' content), Ni-Co base superalloys with g/g' two-phase structure were simplified into binary alloys Ni-9.2%Al and Co-16.9%Ti (mass fraction), which were mixed in certain proportions to obtain a series of Ni-Al-Co-Ti quaternary alloys. X-ray diffraction and morphological analysis results indicate that the g/g' two-phase region can be continuously distributed from the Ni-9.2%Al side to the Co-16.9%Ti side [16], providing a theoretical basis for subsequent design of cast & wrought Ni-Co base alloys.

The new cast & wrought alloy selected U720Li, which has similar microstructure and higher temperature capability, as the optimization target. The alloy design concept is shown in Figure 6 [Figure 6: see original paper][10]. An appropriate amount of Co-Co<sub>3</sub>Ti alloy with g/g' structure was added to Ni-based turbine disc alloys (U720Li, RR1000, ME3) with g/g' structure. Increasing Co content can improve alloy microstructural stability and hot workability, while

increasing Ti content raises the precipitation phase content, thereby enhancing alloy strength. The optimized  $g'$  phase composition is  $(Ni, Co)_3(Al, Ti)$ , which is beneficial for extending the strengthening effect in the medium-temperature service region while potentially causing rapid decay of strengthening effects in the high-temperature processing region, ensuring the hot processing window is not narrower than that of U720Li alloy.

The alloy design method is shown in Figure 7 [Figure 7: see original paper][18]. Thermo-Calc software and the nickel-based database (NI-DATA) were used to calculate the compositions of  $g$  and  $g'$  phases in U720Li alloy and Co-16.9%Ti alloy. U720Li alloy was then mixed with 10%, 15%, and 20% of Co-Ti alloy, and the compositions of  $g$  and  $g'$  phases in the designed alloys were calculated. Different TMW alloys were obtained according to the designed  $g'$  phase volume fractions. The alloys designed in Figure 7 can be used to study the strength variation trend when  $g'$  phase composition is constant but volume fraction changes (TMW-5, TMW-6, and TMW-7), or when  $g'$  phase volume fraction is constant but composition changes (e.g., TMW-1 (U720Li), TMW-4, and TMW-6). The compositions of TMW alloys and some comparative alloys are listed in Table 2 [17]. TMW alloys are characterized by Co content of 20%~31% and Ti content of 5.1%~7.4%, which is higher than current commercial turbine disc Ni-base superalloys [18]. Co and Ti can substitute for Ni and Al positions in  $g'$  to form  $(Ni, Co)_3(Al, Ti)$ , optimizing the  $g'$  phase. Additionally, increasing Co and Ti content while decreasing Cr content can effectively suppress the formation tendency of topologically close-packed (TCP) phases during long-term service [11].

In the new TMW alloys, the  $g'$  composition can be expressed as  $(Ni, Co)_3(Al, Ti)$ . When Ti content is less than 6%, only  $L1_2$ -structured  $(Ni, Co)_3(Al, Ti)$  phase exists in the alloy. When  $6\% < Ti < 10\%$ , blocky or needle-like phase ( $Ni_3Ti$ ) forms, as shown in Figure 8a [Figure 8: see original paper]. Figure 8b shows the time-temperature-transformation (TTT) curves for  $g'$  phase ( $Ni_3Ti$ ) formation in different alloys. The formation of  $g'$  phase is influenced by Ti content, Ti/Al ratio, and alloy composition. Additionally, heat treatment temperature also affects  $g'$  phase ( $Ni_3Ti$ ) formation [19,20].

When Ti content exceeds 10%, a large number of plate-shaped phases are observed, as shown in Figure 9a [Figure 9: see original paper][21,22]. Selected area electron diffraction and HRTEM observations indicate this phase has an hcp structure (Figure 9b) with lattice constants  $a = 0.51$  nm and  $c = 1.25$  nm. Based on three-dimensional atom probe (3DAP) composition analysis results (Figure 9c), the plate-shaped phase can be expressed as  $(Ni, Co, Cr)_3(Ti, Al)$ . The difference between  $g'$  phase and  $(Ni, Co, Cr)_3(Ti, Al)$  phase is that  $g'$  phase has  $L1_2$  structure with low Ti and high Al content, while  $(Ni, Co, Cr)_3(Ti, Al)$  phase has hcp structure with high Ti and low Al content. The difference between  $g'$  phase and  $(Ni, Co, Cr)_3(Ti, Al)$  phase is that both have hcp structure, but  $g'$  phase has  $a = 0.51$  nm,  $c = 0.83$  nm, while  $(Ni, Co, Cr)_3(Ti, Al)$  phase has  $a = 0.51$  nm,  $c = 1.25$  nm.

The transformation process from  $g'$  phase to  $\beta$  phase can be described as follows [23]: Both  $g'$  and  $\beta$  phases have similar  $AB_3$ -type structures, with each unit cell containing 3 A atoms and 1 B atom, and share the same close-packed structure on close-packed planes. In fcc structure, the atomic stacking sequence on close-packed (111) planes is ABCABCABC, while in hcp  $\beta$  phase, the stacking sequence along close-packed (0001) planes is ABACABAC, as shown in Figure 10 [Figure 10: see original paper][23]. When extrinsic stacking faults form in  $g'$  phase, the stacking sequence along close-packed planes is equivalent to adding an A atom layer, changing the normal sequence from ABCABC to ABACABC. The underlined portion has the same atomic arrangement as  $\beta$  phase, which becomes the preferred nucleation site for  $\beta$  phase. The formed  $\beta$  phase core then grows along the [0001] direction on (111) $g'$  planes via superdislocations.

Additionally, Co content in the alloy affects stacking fault energy, influencing microstructure and deformation behavior [24,25]. Figure 11a [Figure 11: see original paper][24] shows a filtered HRTEM image of dislocation decomposition. Combined with other alloy physical parameters, the stacking fault energy can be calculated. Figure 11b [Figure 11: see original paper][25] shows the variation curve of stacking fault energy with Co content in TMW alloys. When Co content is less than 24%, stacking fault energy decreases with increasing Co content. At approximately 24% Co, stacking fault energy reaches its minimum value. When Co content exceeds 24%, stacking fault energy increases with Co content. Therefore, Co content in TMW alloys is generally selected between 22% and 26%.

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## 2. Alloy Processing Properties

Based on the above compositional design, four TMW alloys (TMW-2, TMW-24, TMW-4, and TMW-4M3) were selected for pilot-scale preparation verification through deformation capability evaluation and mechanical property testing of 0.5 kg and 100 kg ingots.

Pilot-scale verification employed a vacuum triple melting process (vacuum induction melting (VIM) + electroslag remelting (ESR) + vacuum arc remelting (VAR)) to ensure metallurgical quality of large-size ingots. VAR ingots with 508 mm diameter were processed through multiple hot forging steps to produce 200 mm diameter bars. Finally, using hot die forging technology, several TMW alloy turbine disc blanks with 440 mm diameter and 65 mm thickness were successfully prepared (Figure 12 [Figure 12: see original paper][26]). All subsequent microstructure and property study specimens were extracted from such disc blanks.

### 3. Microstructural Control and Heat Treatment

The properties of wrought superalloys with g/g' two-phase structure are primarily determined by grain size and g' phase distribution characteristics. The relationship between main properties (tensile strength, creep strength, low cycle fatigue, and fatigue crack growth) and grain size in U720Li cast & wrought disc alloy is shown in Figure 13 [Figure 13: see original paper][27].

Grain size in TMW alloys is determined by solution treatment temperature and time. With increasing solution temperature, the volume fraction of primary g' phase decreases and grain size increases (Figure 14a [Figure 14: see original paper][28]). Meanwhile, primary g' phase size, volume fraction, and grain size are correlated (Figure 14b [Figure 14: see original paper][29]). Additionally, the distribution characteristics (size and volume fraction) of aging and cooling g' phases (secondary and tertiary g' phases) are determined by both solution and aging heat treatment regimes, which affect alloy properties. Table 3 shows the changes in g' phase distribution characteristics and grain size in U720Li and TMW-4M3 alloys after different solution treatments (4M3-1 and U720Li solution-treated at 1100 °C for 4 h followed by oil quenching, 4M3-2 solution-treated at 1120 °C for 4 h followed by oil quenching) [30]. The results indicate that TMW-4M3 alloy can achieve better comprehensive properties using higher solution temperatures. However, for comparison with U720Li, TMW alloy property test specimens were all processed using the standard U720Li heat treatment regime (with solution temperature lower than the optimal temperature for TMW).

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#### 4.1 Tensile Deformation

Figure 15 [Figure 15: see original paper][7,8] shows the variation of tensile strength with temperature for TMW alloys and other comparative alloys. In the 650~750 °C range, TMW-4M3 alloy exhibits higher tensile strength than U720Li and ME3 alloys, particularly at 750 °C where the yield strength is approximately 10% higher than U720Li and about 5% higher than ME3.

Figure 16 [Figure 16: see original paper][31] shows the microstructures of TMW-4M3 and U720Li alloys after tensile deformation at room temperature, 650 °C, and 725 °C. U720Li alloy deformation is dominated by antiphase boundary (APB) shearing of g' phase from room temperature to 725 °C. In contrast, TMW-4M3 alloy shows APB shearing at room temperature, stacking fault shearing and g phase deformation at 650 °C, and deformation microtwin shearing of g' phase and g phase at 725 °C [31].

Based on these observations, a model describing the deformation mechanisms in Ni-Co base wrought superalloys was proposed. The two deformation mechanisms in alloy tensile deformation are APB shearing and stacking fault shearing of g' phase. By comparing the surface APB energy and stacking fault energy per unit deformation area, the dominant deformation mechanism in a specific

temperature range can be determined [32]. The expressions for surface APB energy and stacking fault energy per unit deformation area are:

$$\theta f_{\gamma}(APB) \quad (1)$$

$$\frac{DE}{f_{\gamma'III} + f_{\gamma'II}} + \frac{(SF)}{f_{\gamma'III} + f_{\gamma'II}} \quad (2)$$

where DE is the increased surface energy per unit deformation area,  $\gamma\{APB\}$  is the APB energy of g' phase,  $f_{g' III}$  is the volume fraction of tertiary g' phase,  $\theta$  is a coefficient between 0 and 1,  $f_{g' II}$  is the volume fraction of secondary g' phase,  $\gamma\{SF\}$  refers to the stacking fault energy of g' phase,  $\gamma_m$  is the stacking fault energy of g phase, and  $f_m$  is the volume fraction of g phase. Based on known parameters, schematic diagrams of APB shearing and stacking fault shearing can be plotted, as shown in Figure 17 [Figure 17: see original paper][32]. There exist critical temperature and critical tertiary g' phase volume fraction between the two mechanisms. When deformation temperature or tertiary g' phase volume fraction is below the critical value, APB shearing dominates; otherwise, stacking fault shearing dominates. Since TMW-4M3 alloy has higher volume fraction of tertiary g' phase, higher APB energy, and lower stacking fault energy than U720Li alloy, the transition in tensile deformation mechanisms occurs at lower critical tertiary g' phase volume fraction and lower temperature, consistent with TEM observations.

During tensile testing in the 300~550 °C temperature range, TMW alloys exhibit both conventional dynamic strain aging and anomalous dynamic strain aging phenomena (Figure 18a [Figure 18: see original paper][33]). Conventional dynamic strain aging typically occurs at low temperatures or high strain rates, while anomalous dynamic strain aging appears at high temperatures or low strain rates. TEM observations confirm that the interaction between solute atoms (C atoms) and mobile dislocations is the main cause of conventional dynamic strain aging (Figure 18b [Figure 18: see original paper][34]), while anomalous dynamic strain aging is related to stacking faults (Figure 18c [Figure 18: see original paper][34]). This new phenomenon of anomalous dynamic strain aging involving stacking fault participation in plastic deformation has not been previously reported in dynamic strain aging studies. To further investigate anomalous dynamic strain aging behavior, different heat treatment regimes were selected to alter the volume fraction and distribution of tertiary g' phase [34~36]. The results show shortened stacking faults (Figure 18d [Figure 18: see original paper][34]), confirming the association between anomalous dynamic strain aging and stacking fault participation in plastic deformation.

Statistical results of stacking fault density and stacking fault energy in TMW alloys under different deformation conditions (Figure 19 [Figure 19: see original paper][37]) indicate: stacking fault density is lowest in room-temperature

deformed samples; from room temperature to 350 °C (623 K), stacking fault density increases slowly with temperature; from 350~450 °C (623~723 K), stacking fault density increases sharply, reaching a maximum at 450 °C (723 K); from 450~725 °C (723~998 K), stacking fault density gradually decreases. Stacking fault energy increases with temperature from room temperature to 623 K, then decreases sharply with further temperature increase (623~723 K), reaching a minimum at 723 K, and subsequently increases again with temperature (723~998 K). Observations of samples deformed to different strain levels at 450 °C with a strain rate of  $8 \times 10^{-5} \text{ s}^{-1}$  indicate that the initial stage of dynamic strain aging is caused by dislocation unpinning, manifested primarily as full dislocations dissociating into stacking faults. With increasing deformation, interactions between stacking faults and between stacking faults and dislocations become the main factors affecting dynamic strain aging.

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## 4.2 Creep Deformation

Figure 20 [Figure 20: see original paper][9,38] shows the Larson-Miller parameter relationship between time and stress for TMW and comparative alloys at 0.2% creep strain. At the same creep time, the load-bearing capacity of TMW-4M3 alloy is about double that of U720Li. At a load of 630 MPa, the temperature capability is 76 °C higher than U720Li (estimated rupture life improvement of over 10 times), approaching the level of PM alloy ME3 [9,38].

Figure 21 [Figure 21: see original paper] compares the deformation microstructures of U720Li and TMW-4M3 alloys at different creep stages under 725 °C and 630 MPa conditions [38~40]. Both alloys have grain sizes of approximately 10 mm. In U720Li alloy, the primary creep stage is dominated by Orowan looping and dislocation slip, the secondary stage by stacking fault shearing of g' phase, and the tertiary stage exhibits grain boundary sliding. In TMW-4M3 alloy, the primary stage is dominated by stacking fault shearing, while the secondary and tertiary stages are dominated by microtwin deformation, with microtwin thicknesses of  $(109 \pm 15) \text{ nm}$  and  $(76 \pm 12) \text{ nm}$  in the second and third stages, respectively. These results demonstrate that stacking fault energy significantly influences creep deformation mechanisms.

Since stacking fault energy has a substantial effect on creep mechanisms in TMW alloys, several Ni-Co base wrought superalloys with different stacking fault energies were designed to study the influence of stacking fault energy variation on creep deformation mechanisms [41,42]. The results show that during the primary creep stage or at small creep deformations, creep mechanisms are mainly related to Orowan looping and APB shearing of smaller g' phases, with stacking fault energy variation having little effect. During the secondary creep stage or at large creep deformations, stacking fault energy variation significantly affects creep behavior: steady-state creep rate decreases with decreasing stacking fault energy, and creep life increases. In alloys with higher stacking fault energy,

$a/2\langle 110 \rangle$  full dislocations decompose at  $g'/g$  interfaces to form Shockley partials that shear larger  $g'$  phases and create stacking faults within them. In alloys with lower stacking fault energy,  $a/2\langle 110 \rangle$  full dislocations decompose in the  $g$  matrix, and the resulting  $a/6\langle 112 \rangle$  partials sweep through both  $g$  matrix and  $g'$  phases, forming extended stacking faults and deformation twins.

Figure 22 [Figure 22: see original paper][43] schematically illustrates the variation of creep deformation mechanisms with creep conditions in large-grain-size ( $>100$  nm) Ni-Co base wrought superalloys. Due to the low stacking fault energy of TMW alloys, stacking faults form more readily during deformation. According to Zhang et al. [44], stacking fault width is stress-dependent; when stress increases, stacking fault width broadens, allowing stacking faults in different  $g'$  phases to connect and form continuous stacking faults. Therefore, at low temperatures and low stresses, deformation occurs primarily through partial dislocations shearing  $g'$  phases to form numerous stacking faults. During medium-temperature deformation, when loading stress is high, the stress dragging dislocation motion also increases, resulting in numerous stacking faults penetrating both  $g$  matrix and  $g'$  phases. When creep temperature rises to  $725$  °C or loading stress is high, dislocation motion accelerates in the alloy, full dislocations in the matrix easily decompose, forming deformation twins. Alloy deformation then proceeds mainly through stacking faults and microtwins shearing  $g'$  phases, creating stacking faults and microtwins penetrating both  $g$  matrix and  $g'$  phases. Simultaneously,  $g'$  phases begin to coarsen; during coarsening of larger  $g'$  phases, some smaller  $g'$  phases dissolve, widening the matrix channels between  $g'$  phases and facilitating dislocation bypassing. When creep temperature further increases to  $815$  °C, some small  $g'$  phases dissolve, the matrix channels between large  $g'$  phases widen, and dislocations bypass  $g'$  phases.

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### 4.3 Low Cycle Fatigue Behavior

The low cycle fatigue performance of TMW alloys at  $400\sim 725$  °C with strain amplitude range of  $0.79\%\sim 1.22\%$  is shown in Figure 23 [Figure 23: see original paper][8,9,45]. At  $400\sim 650$  °C (Figures 23a and 23b), TMW alloys exhibit better low cycle fatigue performance than U720Li alloy. At  $725$  °C with low strain amplitude (Figure 23c), TMW alloys also show superior low cycle fatigue performance to U720Li, but at high strain amplitude, TMW alloy performance is inferior to U720Li. TMW alloys show cyclic hardening at  $400$  °C, but pronounced cyclic softening at  $650\sim 725$  °C (Figure 23d), with both hardening and softening degrees increasing with strain amplitude. Fatigue life decreases with increasing temperature and strain amplitude. TEM observations (Figure 24 [Figure 24: see original paper][45]) indicate that at  $400$  °C, alloy cyclic hardening is caused by APB shearing of  $g'$  phase, while stacking fault shearing of  $g'$  phase may be responsible for cyclic softening.

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#### 4.4 Fatigue Crack Growth Behavior

Fatigue crack growth rates of TMW alloys at 400–725 °C with a stress ratio of 0.05 are shown in Figure 25 [Figure 25: see original paper][8,9,46,47]. At 400 °C (Figure 25a), TMW alloys have lower fatigue crack growth rates than U720Li alloy. At 650 °C (Figure 25b), fatigue crack growth rates are essentially equivalent to U720Li. At 725 °C (Figure 25c), fatigue crack growth rates are slightly higher than U720Li. Fracture morphology analysis indicates [47] that fracture at 400 °C is primarily transgranular, while at 650 °C and 725 °C, mixed intergranular and transgranular fracture modes occur.

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#### 4.5 Alloy Strengthening Mechanisms

The main strengthening mechanisms in TMW alloys include solid solution strengthening, g' phase precipitation strengthening, and grain boundary strengthening. To evaluate the contribution of different strengthening methods to alloy strength, TMW-4M3 alloy was subjected to different solution treatments, and the contributions of several strengthening mechanisms were quantitatively analyzed through hardness and tensile tests [48,49]. Figure 26 [Figure 26: see original paper][49] shows the microstructures of TMW-4M3 alloy after different solution treatments. When heat-treated below the g' solvus temperature, three sizes of g' phases exist in the alloy (primary, secondary, and tertiary g' phases). When heat-treated above the g' solvus temperature, only two sizes of g' phases exist (secondary and tertiary g' phases), with larger grain sizes. The yield strength of the alloy can be expressed as:

$$\sigma_{0.2} = \sigma_{Ni} + \Delta\sigma_{sol} + \Delta\sigma_{\gamma'I} + \Delta\sigma_{GB} + \Delta\sigma_{\gamma'III} + \Delta\sigma_{\gamma'II}$$

Analysis shows that yield strength  $\sigma_{0.2} = 2.46$  HV, so the equation can be rewritten to quantitatively express the contribution of different strengthening mechanisms:

$$\sigma_{0.2} = 2.46(H_{Ni} + \Delta H_{sol} + \Delta H_{\gamma'I} + \Delta H_{GB} + \Delta H_{\gamma'III} + \Delta H_{\gamma'II} + \Delta H_{TB})$$

where  $\sigma\{Ni\}$  is the strength of Ni matrix,  $\Delta\sigma\{sol\}$  is the strength contribution from solid solution strengthening,  $\Delta\sigma_{g'I}$ ,  $\Delta\sigma_{g'II}$ ,  $\Delta\sigma_{g'III}$  are strength contributions from primary, secondary, and tertiary g' phases,  $\Delta\sigma\{GB\}$  is the strength difference contributed by grain boundaries,  $\Delta\sigma\{TB\}$  is the strength difference contributed by twin boundaries, and H with corresponding subscripts represent the Vickers hardness values of the respective components.

Figure 27 [Figure 27: see original paper][49] shows the effect of different solution treatment temperatures on tensile yield strength and the contributions of various strengthening mechanisms. From the equation and Figure 27, the tensile yield strength consists of three main parts: Region 2 in the figure represents solid solution strengthening and tertiary  $g'$  precipitation strengthening; Region 3 represents secondary  $g'$  precipitation strengthening; and Region 4 represents primary  $g'$  precipitation strengthening, grain boundary strengthening, and twin boundary strengthening. As shown in Figure 26, grain size increases with solution temperature. According to the Hall-Petch relationship, grain boundary strengthening capability decreases. Meanwhile, due to the low volume fraction of primary  $g'$ , its strengthening contribution is not significant. When solution-treated above the  $g'$  solvus temperature, this contribution almost reduces to zero. The contribution from  $\Delta\sigma_{\text{sol}} + \Delta\sigma_{g' \text{ III}}$  gradually increases with solution temperature, reaching a maximum near the  $g'$  solvus temperature, equivalent to the strength of single-crystal superalloy TMS-4M3. The contribution from  $\Delta\sigma_{g' \text{ II}}$  also increases with solution temperature. Since  $\Delta\sigma_{\text{sol}} + \Delta\sigma_{g' \text{ III}}$  does not change with temperature and tertiary  $g'$  phase volume fraction increases,  $\Delta\sigma_{\text{sol}} + \Delta\sigma_{g' \text{ III}}$  increases. Therefore, the optimal solution treatment temperature for TMW-4M3 alloy should be between 1130~1140 °C. Different aging treatment regimes can be applied according to performance requirements to obtain optimally distributed secondary and tertiary  $g'$  phases. These results reveal the main strengthening factors in wrought superalloys and provide guidance for selecting solution temperatures and subsequent aging treatments.

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## Conclusions

1. The new cast & wrought TMW alloy uses the same alloying elements as U720Li alloy, but with Co content of 20%~31% and Ti content of 5.1%~7.4%, which is higher than U720Li and current commercial turbine disc Ni-base superalloys.
2. Compared with U720Li alloy, the new cast & wrought TMW alloy exhibits higher tensile strength, creep strength, low cycle fatigue resistance, and crack growth resistance. The temperature capability is more than 50 °C higher than U720Li, approaching the level of powder metallurgy superalloy ME3 (Rene104).
3. During tensile deformation, TMW alloy exhibits two mechanisms—antiphase boundary (APB) shearing and stacking fault shearing—determined by critical temperature and critical tertiary  $g'$  phase volume fraction. When deformation temperature or tertiary  $g'$  phase volume fraction is below the critical value, APB shearing dominates; otherwise, stacking fault shearing dominates.
4. Creep deformation mechanisms in TMW alloy include: In the primary creep stage, deformation is related to Orowan looping and APB shearing

of smaller  $g'$  phases, with stacking fault energy variation having minimal effect. In the secondary creep stage or at large creep deformations, stacking fault energy variation significantly affects creep behavior: steady-state creep rate decreases with decreasing stacking fault energy, and creep life increases.

5. TMW alloy exhibits both conventional dynamic strain aging and anomalous dynamic strain aging phenomena in the 300~550 °C temperature range, with anomalous dynamic strain aging being related to stacking faults (SF).

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