Original Article

Property optimization of low-cycle fatigue in Al-Si piston alloy at elevated temperatures by ultrasonic melt treatment

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A R T I C L E   I N F O

Article history:
Received 26 March 2019
Accepted 31 July 2019
Available online 22 August 2019

Keywords:
Al-Si alloys
Low cycle fatigue
Fatigue crack initiation
Temperature
Property optimization

A B S T R A C T

The cyclic deformation and damage behaviors of Al-Si piston alloys are comprehensively studied. It is found that the fatigue cracks mainly initiate from broken primary Si at low temperature and phase/matrix interface debonding at higher temperature. The transgranular crack propagation can be found for all the temperatures and the grain is obviously reduced after low-cycle fatigue at higher temperature. Furthermore, a hysteresis energy-based life prediction model was developed and utilized. Based on the model, the optimum fatigue life was found at intermediate temperature. A strategy for fatigue property optimization was proposed: increasing W0 at low temperature and increasing β at high temperature, to enhance the fatigue life during entire service temperatures with the ultrasonic melt treatment. In this way, the remarkable improvement of the fatigue properties may be achieved.

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1. Introduction

Nowadays, the multi-phase Al-Si alloys have been widely used in key component operating at high temperature such as diesel engine piston, because of their low thermal expansion, low density, outstanding formability, and excellent wear resistance [1–4]. Due to the high temperature strength and excellent thermal stability provided by Ni, Cu and Mg, the Al-12Si-4Cu-2Ni-Mg alloy turns out to be one of the excellent materials for piston [5–7]. For the alloy, the Si are commonly added to promote strength and castability, and the transition elements addition improves mechanical properties at elevated temperatures through forming the stable intermetallics (IMs) [1,8]. However, the pistons are subjected to cyclic loading at alternating temperatures, such as start-up and shut-down operating cycles [2,3]. The service conditions induce in the material a complex stress/strain in the hottest part facing the combustion chamber [2,9,10]. The long-term performance of pistons often depends on the fatigue property of the alloy, especially for the strain-controlled low-cycle fatigue (LCF) at elevated temperature.

In order to enhance the reliability of pistons, it is very important to improve the material’s fatigue property [11–14]. The significant achievements have been obtained to improve materials’ fatigue resistance, including reducing...
Where:

- Grain size, optimizing alloy composition and surface pretreatment [15–18]. As for the cast materials, there are mainly two approaches: (i) chemical refinement using a minor addition of an alloying element such as Na, P or Sr [11,19,20] and (ii) physical refinement with external influences such as the casting process and heat treatment. However, these methods are substantially effective for the LCF at room temperature [25,26], the results of fatigue optimization are often not satisfactory in LCF at elevated temperatures. The difficulty mainly comes from that it is lacking the understanding on what attributions of material to determine its LCF property at elevated temperatures. For the fatigue analysis, there are three main problems to be solved:

- Macroscopic properties (fatigue model): Traditional strain-life and stress-life approaches are used in most industries for fatigue life prediction [27,28]. However, many fatigue results indicate that the LCF life is not only affected by strain amplitude, but also by the cyclic stress. For the same experimental data, the opposite judgments can be obtained under the two criteria, which will result in the divergence of the optimization direction [15,25]. Recently, an energy-based LCF life prediction model was proposed and used successfully [29,31–33]:

\[ N_f = \left( \frac{W_0}{W_0'} \right)^\beta \]  

Where, \( N_f \) is fatigue life, \( W_0 \) is the saturation hysteresis energy, \( W_0' \) and \( \beta \) are material parameters.

- Microscopic deformation mechanisms (damage behavior): Essentially, fatigue is a process of damage accumulation, subjecting to external cyclic loading. In our previous studies [25,29,31], the LCF life may be controlled by two parameters, the fatigue damage exponent \( \beta \) and the intrinsic fatigue toughness \( W_0 \). The former term impacts the fatigue damage rate and microscopic deformation mechanism. The latter term helps retard the fatigue crack propagation (high ductility) and improve the fatigue damage capacity. Since the hysteresis energy is related to cyclic stress and strain amplitudes, the model provides a new perspective for LCF performance optimization.

- Improving the LCF property (fatigue optimization): From the aspect of fatigue optimization, the increasing of \( W_0 \) and \( \beta \) are beneficial for the improvement of LCF life. Improving the static toughness (product of strength & elongation) of materials can be an effective method to improve \( W_0 \). Improving deformation reversibility and damage uniformity is beneficial to the increase of \( \beta \) value. However, the \( W_0 \) and \( \beta \) show an obvious inverse relation for most alloys (Fig. 1a), which will restrict the improvement of LCF life [25,34]. A reasonable distribution of the parameters is possibly an effective way to improve the overall fatigue performance when the two parameters cannot be increased simultaneously [29].

Given this, a new material optimization strategy including such as increasing \( W_0 \) at low temperature and increasing \( \beta \) at high temperature for the alloy was introduced (Fig. 1b). Ultrasonic melt treatment is one of the more promising means of improving the mechanical properties of Al alloys, as it effectively reduces the porosity while simultaneously refining the microstructure through cavitation-induced dendrite fragmentation and/or cavitation-induced heterogeneous nucleation [35]. The corresponding LCF properties of the Al-Si alloys with diverse microstructures prepared by different casting processes were compared and investigated. And then the deformation mechanisms and damage behaviors were carefully researched focusing on fatigue crack initiation and propagation, as well as microstructure evolution.

2. Experimental materials and processes

The two high-performance Al-Si piston alloys with similar chemical composition of Al-12Si-3Cu-2Ni-1Mg (wt%) were studied. The as-cast piston ingots were prepared by different processing routes: gravity casting without (AC) and with ultrasonic melt treatment (UT). The fatigue samples cut from the casted piston ingots have gauge dimensions of 8 mm (diameter) × 15 mm (length) after T6 heat treatment. LCF tests controlled with total strain amplitude were carried out at 280 °C, 350 °C and 425 °C with an Instron 8862 testing machine under a constant strain rate \( 5 \times 10^{-4} \text{ s}^{-1} \). The grain size distributions were measured by the EBSD (electron backscatter

Fig. 1 – Schematic diagram of improving LCF property: (a) summarization of the LCF parameters for metallic materials [25,34]; (b) suggestion of improving LCF life based on fatigue parameters with the increase of temperature [29].
diffraction) integrated in a ZEISS SUPRA35 scanning electron microscope (SEM). Transmission electron microscopy (TEM) samples were cut from the Al-Si alloy samples, ground to a thickness of 0.05 mm and then twin-jet electro-polished at −20 °C using a solution of 20% perchloric acid and 80% methanol by volume, and the TEM foils were examined using a FEI Tecnai F20 microscope.

3. Results

3.1. Microstructure

The typical microstructures including the primary Si and grain size distributions in the two alloys are shown in Fig. 2. The alloy contains α-Al, eutectic/primary Si and several intermetallic compounds (including AlCuNi, AlFeMnNi, Al3(Fe, Mn, Co) and β-Al6FeSi phases). And the atomic percent was measured by energy dispersive spectroscopy (EDS), the similar intermetallic compounds can be found in the studies [8,36,37]. The primary Si particles are distributed homogeneously in the UT alloy (Fig. 2b). The UT reduces not only the average size greatly (from 33.1 μm to 24.3 μm) but also the maximum size (from 65 μm to 55 μm) of primary Si particles. The EBSD microstructures of the α-Al matrix show that the average grain size (from 516 μm to 439 μm) and the maximum size (from 1273 μm to 1164 μm) are also reduced by UT (Figs. 2c and d).

3.2. The tensile and fatigue properties

Fig. 3a displays the engineering stress-strain curves of Al-Si piston alloys at various temperatures. The ultimate tensile strength (UTS) and elongation to fracture of the two alloys can be found in Fig. 3b. With increasing temperature, the UTS decreases, and the elongation to fracture increases for both alloys. While, the UT alloy exhibits higher tensile strength and elongation for all the temperatures (Fig. 3b). The improvement of the strength of UT alloy is mainly attributed to the refinement of microstructures, including primary Si, intermetallic phases and grain.

The dependences of fatigue life (Nf) on the plastic strain amplitude (Δεp/2) and stress amplitude (Δσ/2) of AC and UT
alloys at different temperatures are illustrated in Fig. 3. No matter for strain-life or stress-life, significant enhancement of LCF property is obtained in UT samples compared with AC for all test temperatures. It is interesting to note that the UT samples present a higher fatigue life than AC one at all the tested temperatures. The distinctive LCF performance of UT alloy can be related to its cyclic property evolutions and damage behaviors.

3.3. Cyclic deformation behaviors

The cyclic response curves of AC and UT alloys at different temperatures are displayed in Figs. 4a–b, respectively. Some common characteristics can be seen in the alloy: First, the fatigue life increases with decreasing strain amplitude, while the cyclic stress increases with increasing strain amplitude for all the tested temperatures. Second, the cyclic softening occurs remarkably or slightly depending on the temperature and strain amplitude. Third, the half-life hysteresis loops (Fig. 4c and d) indicate that the cyclic stress decreases and cyclic plastic strain increases obviously at higher temperatures.

4. Discussion

4.1. Fatigue life prediction model

In general, the fatigue life prediction models are often based on cyclic strain (e.g. the Manson-Coffin relation for LCF) or cyclic stress (e.g., the Basquin equation for HCF). However, both of the plastic strain and cyclic stress are significant factors for fatigue behaviors. The hysteresis energy ($W_{h}$) involving both strain and stress amplitudes shows more stable than stress or strain solely and can be used more reasonably to evaluate the fatigue property. The $N_{f}$–$W_{h}$ relationships on log-log scales display the crossover of LCF life curves at elevated temperatures (the $W_{h}$ here were acquired from half-life time). At the same hysteresis energy, the highest fatigue life can be found in the intermediate temperature range, for example, the AC alloy at 350 °C and the UT alloy at 280 °C (Fig. 5a). The corresponding values of fatigue parameters based on different models can be seen in Table 1. In the model, the value of $\Delta S_{p} \cdot \Delta \sigma$ and imposed plastic work $W_{i}$ shows a linear relationship (Fig. 5b):

$$W_{i} = \int \sigma \cdot d e = k \cdot \Delta S_{p} \cdot \Delta \sigma$$  \hfill (2)

Here $k$ is designated as “shape factor”. From the formula, the $W_{i}$ can be achieved with a concise way.

The effects of temperature on fatigue life and damage behavior are also discussed and analyzed. The values of two parameters $W_{0}$ and $\beta$ were calculated from the LCF model. The $W_{0}$ and $\beta$ increase linearly with increasing temperature (Fig. 5c), it is worth noting that the $W_{0}$ and $\beta$ show an obvious inverse relation with temperature increasing for the present alloys), which can be expressed as $W_{0} = e T + f$ and $1/\beta = m T + n$, respectively. By replacing the parameters of $W_{0}$ and $\beta$ in Eq. (1), a new relation can be obtained:

$$W_{h} = (e T + f) \cdot N_{f}^{-(m T + n)}$$  \hfill (3)
Fig. 4 – Cyclic deformation behaviors of Al-Si alloys prepared by AC (a, c) and UT (b, d) at elevated temperatures: (a-b) cyclic stress response curves; (c-d) half-life hysteresis loops.

Where, the $e$, $f$, $m$ and $n$ are constants for the given testing condition and material (e.g., $e_1=0.18$, $f_1=50.5$, $m_1=2.1 \times 10^{-3}$, $n_1=0.3$ for the AC alloy; $e_2=0.13$, $f_2=-23.8$, $m_2=1.1 \times 10^{-3}$, $n_2=0.068$ for the UT alloy). The life prediction factor (LPF) is usually adopted to estimate the accuracy of prediction method [10]:

$$\text{LPF} = \max \left\{ \frac{N_{\text{cal}}}{N_{\text{exp}}}, \frac{N_{\text{exp}}}{N_{\text{cal}}} \right\}$$

(4)

Where, $N_{\text{exp}}$ and $N_{\text{cal}}$ are LCF life obtained from the experiment and calculation respectively. Based on Eqs. (3) and (4), the experimental and calculated results are shown in Fig. 5d, which indicates that the proposed model exhibits a fine adaptation for the experimental results within the double of the predicted life (LPF = 2), indicating that the model suits for the experimental data well. Furthermore, the evolution of the fatigue life with the temperature at a constant $W_a$ can be seen in Fig. 5e. The $N_i$ increases at first and then decreases with increasing temperature for both materials. A critical temperature ($T_c$) can be found at maximum fatigue life. The experimental results in Fig. 5a generally show the characteristics predicted above. For example, the predicted critical temperatures for AC and UT alloys are about 325 °C and 275 °C, respectively (Fig. 5e), and the experimental results show that at 350 °C and 280 °C, the fatigue lives are the longest for AC and UT alloys, correspondingly. At a constant temperature, if the $W_a$ increases, the $N_i$ decreases as indicted by Eq. (3). However, the $T_c$ slightly increases with $W_a$ increasing (Fig. 5f). On the other hand, the $T_c$ decreases for UT alloy, which means that the refined phases and prolonged fatigue life will increase the proportion of time-dependent damage.

4.2. Fatigue damage mechanisms

In general, the LCF damage mechanisms are mainly dominated by the plastic strain, which are characterized by microstructure damage such as fatigue crack initiation and propagation for the Al-Si alloys.

(a) Fatigue crack initiation: In order to obtain more accurate damage behavior at different temperatures, the corresponding features of fatigue cracking at $\Delta e/2$ of 0.2% are shown in Fig. 6. At 280 °C, some micro-cracks caused by the broken IMs and primary Si around the main crack can be found, which dominate the fatigue cracking behavior for both materials (Figs. 6a and d). At 350 °C, some debonded or broken Si phases along crack path can be found (Figs. 6b and e). When the temperature increases 425 °C, the well-developed micro-cracks in the secondary phases (Fig. 6c) and lots of debonded primary Si particles (Fig. 6f) occur in the matrix. Careful observation indicates that particle cracking dominates the surface damage, and the main cracks may expand along the broken phases. Considering the influence of microstructure on deformation damage behaviors, there are some differences for the two alloys. For instance, the numerous secondary cracks are easily found on the surface of fatigued AC specimen (Figs. 6a–c). As for the UT specimen, secondary micro-cracks are seldom seen in the matrix. For the influence of temperature on fatigue damage behavior, the broken Si and IMs dominate the cracking behavior for both alloys at 280 °C; the deboned Si can be found around main crack for 350 °C and then dominate the damage behavior at 425 °C for UT alloy. But for AC alloy, the deboned phases can not be found until the temperature increases to 425 °C.
Continuously cyclic straining leads to deformation mismatch between the primary Si and matrix, inducing the local stress/strain concentration and the micro-cracks initiation. Two different crack behaviors can be found at different temperatures; i.e. the broken primary Si for lower temperature and debonded Si mainly for higher temperatures. At the lower temperatures (280 °C and 350 °C for AC alloy; 280 °C for UT alloy), the stress of primary Si particles is relatively high, the particles should be easy to crack and the micro-cracks initiate from there. With increasing temperatures (425 °C for AC alloy; 350 °C and 425 °C for UT alloy), the micro-plastic deformation around the interface between Al matrix and Si particles are serious. The fatigue cracks commonly initiate from the interface between Al matrix and primary Si. The continued mechanical loading leads to form the networks of micro-cracks, which result in fatigue crack propagation and coalescence. In order to find out the fatigue cracking behaviors explicitly for the two alloys at elevated temperatures, the cracking behaviors were measured and counted as shown in Figs. 8a and b. Both the damage range (the longest distance between secondary cracks and main crack) and the crack density (total length of cracks near fracture surface in unit area) increase with increasing temperature (Fig. 7a). This indicates that the fatigue cracking is easier at high temperature. However, the damage range and crack density for UT specimen are definitely smaller than those for AC specimen (Fig. 7a), which means that the UT specimen exhibits better resistance to fatigue crack initiation, especially at high temperature.

(b) Fatigue crack propagation: The EBSD images of fatigue crack propagation behaviors are shown in Fig. 8, in which the main crack is within the two dotted lines. The direction of main cracks propagation is indicated by the arrows and it is about perpendicular to the loading direction. It is worth noting that the fracture feature is transgranular at all the test temperatures; the grain boundary is not the main damage source even at high temperatures, different from the current view [38–41]. The strong crack path deflection in refined grains for UT specimen may endow a declined effective crack propa-
Fig. 6 – Microscopic damage characteristics around main fatigue crack at $\Delta \varepsilon/2$ of 0.2% and different temperatures after fracture: (a) AC-280°C ($N = 2246$); (b) AC-350°C ($N = 13,501$); (c) AC-425°C ($N = 6943$); (d) UT-280°C ($N = 13,915$); (e) UT-350°C ($N = 17,282$); (f) UT-425°C ($N = 15,966$).

Fig. 7 – Microscopic damage characteristics and analysis: (a) range of damage and crack density; (b) grain size evolution; (c) size of $\theta$ phase after fatigue.
gation speed than AC specimen, and finally lead to a longer fatigue life. Meanwhile, a decrease in grain size with increasing fatigue test temperature for both alloys can be found as shown in Fig. 7b. In comparison with the initial grain sizes at room temperature (AC: 516 μm and UT: 439 μm), only a slight decrease (about 3%) can be found at 280 °C for both materials. For the 350 °C, the refined gain sizes decrease about 11%. At 425 °C, gain sizes decrease about 22% and 16% for AC and UT materials respectively (Fig. 7b). The decreasing in grain size indicates that more plastic strain energy can be absorbed and dissipated, and even lead to improvement of fatigue resistance at higher temperature. The grain refinement mechanism at high temperature will be discussed later. But there is no significant change in the secondary phase size (Si and IMs) at high temperature.

4.3. Microscopic deformation mechanism

The TEM micrographs of the AC and UT samples are shown in Figs. 9–11. The massive Guinter-Preston (G-P) zones and precipitated α′′ phases distributed in the original microstructures of Al matrix. The dislocations tangle around the precipitated phases cannot be observed without the cyclic loading (Fig. 9). However, due to the high surface area present, the mixture is not initially in thermodynamic equilibrium. According to the Gibbs-Thomson effect [42,43], the solute atoms concentration in the matrix adjacent to the particles varies depending on the radius of the particles; solute concentration around the large particles is lower than that of the small particles. The concentration gradient between these different particles causes the solute move from the small particles to the large particles, which is termed as Ostwald ripening or coarsening [43,44]. The spontaneous process will accelerate at higher temperature. After high-temperature LCF, the precipitated phases (G-P zones and α′′ phases) transform from long and massive strips to globular particles (α phases). Meanwhile, with increasing the test temperature, the precipitated phases are constantly coarsening and the number of precipitated α phases obviously reduces (Fig. 10).

The Fig. 11 shows the evolutions of dislocations and phases in the two alloys at different temperatures. At 280 °C, some high dislocation density clusters tangled with phase particles occur in the Al matrix under cyclic loading (Figs. 11a and d). Because of dynamic recrystallization and dynamic recovery are intensified at elevated temperatures (350 °C and 425 °C), the dislocation annihilation and substructure reconstruction increase significantly. The grain refinement mechanism of the Al matrix may be explained as follows: (1) the proliferation and annihilation of dislocations in the original grains; (2) the formation of dislocation tangles and dislocation cells due to pile-up of dislocations; (3) the evolution of dislocation tangles and dislocation cells into subgrains, and (4) the transformation of the continuous dynamic recrystallization in subgrain boundaries to refined grains [45]. There refinement of grain at high temperature can be attributed to the development of the dislocations and substructure, as seen in Fig. 11b, c, e, and f.

Based on the discussed above, the cyclic softening in the Al-Si alloys may be attributed to two possible mechanisms:

Fig. 8 – EBSD image of fatigue crack propagation at Δε/2 of 0.2% and different temperatures: (a) AC-280 °C; (b) AC-350 °C; (c) AC-425 °C; (d) UT-280 °C; (e) UT-350 °C; (f) UT-425 °C.
one is macro-phases cracking (such as primary Si and IMs), which is mainly influenced by mechanical strain or stress; the other is evolution of dislocation configurations and micro-phases coarsening (evolution from $\theta^\prime$ or $\theta$ to $\theta'$), which is mainly influenced by temperature and cyclic time.

4.4. Mechanisms of LCF improvement

Essentially, fatigue is a process of damage accumulation. As mentioned above, the LCF life may be controlled by fatigue damage exponent $\beta$ and intrinsic fatigue toughness $W_0$. The following discussion will focus on the influence of $W_0$ and $\beta$ on the damage behaviors, as well as the optimization of the fatigue life and fatigue performance of the alloy.

Increasing $W_0$: From the aspect of fatigue life optimization, the increase of $W_0$ is beneficial for the improvement of LCF life (Fig. 12a sketched base on the experiment results of Fig. 5a and Table 1). The enlarged $W_0$ originates from increasing the tolerance of defects, including the critical energy for the micro-cracks initiation, crack propagation and coalescence. Based on the cyclic deformation behaviors, the enhancement of fatigue life at higher temperatures should be mainly attributed to the relatively higher ductility as well as the refinement of the microstructure. Considering the lower bound of $W_0$ at low temperature, the crack rapid propagation in the matrix is usually caused by the deficiency of crack propagation resistance, and then dominates fatigue damage after occurrence of the fatigue crack. The change of fatigue damage at higher temperature is influenced by the deformation.
Fig. 11 – TEM observation of the LCF samples at Δεt/2 of 0.2% at different temperatures: (a) AC-280 °C; (b) AC-350 °C; (c) AC-425 °C; (d) UT-280 °C; (e) UT-350 °C; (f) UT-425 °C.

Fig. 12 – Summary of the LCF behaviors and mechanism of the two alloys: (a) the influence of fatigue parameters evolution on fatigue life; (b) the change of fatigue damage mechanisms with temperature; (c) fatigue parameters evolution with increasing temperature; (d) schematic diagram of improving LCF property.
homogeneity and reversibility. Specifically, there is a general correlation between fatigue toughness \( W_0 \) and static toughness \( U \) (the product of tensile strength and elongation), which is the main reason for the enlarged \( W_0 \) at high temperature. Therefore, improving the comprehensive toughness of materials can be an effective method to improve \( W_0 \). The UT alloy increases \( W_0 \) with decreasing the grain and second phase size (primary Si and IMCs); a significant improvement of fatigue life can be found at lower temperature.

Increasing \( \beta \): With a certain fatigue toughness \( W_0 \), the improvement of \( \beta \) leads to a longer fatigue life \( N_f \), implying that the alloy possesses a better fatigue damage resistance (Fig. 12a). The fatigue damage exponent \( \beta \) is closely related to the plastic deformation mechanism of the materials. Improving the plastic deformation reversibility and damage uniformity of materials (by changing alloy composition or microstructure) are beneficial to the improvement of \( \beta \) value [31]. However, the reduced \( \beta \) can be found with increasing temperature as mentioned above, which indicates the enhanced damage of the material at high temperature (Fig. 12b, also see Fig. 5c and Table 1). Cyclic deformation leads to the formation of many micro-cracks by debonded or broken Si and then decreases the LCF life at high temperature. For the fatigue damage, the micro-cracks mainly initiate from primary Si cracking induced by piling-up of dislocations at low temperature and phase/matrix interface debonding induced by vacancy accumulation at higher temperature (Fig. 12b). The differences of crack initiation mechanisms at particles are mainly affected by two aspects: the change of dislocations slip mode from planar slip to wavy slip with increasing temperature (influenced by temperature) and the critical stress/strain for the crack initiation on the particles (influenced by microstructure). In the first aspect, the three-dimensional wavy slip occurs frequently to cause the annihilation of edge dislocations and leaving numerical point defects (mainly vacancies), which will concentrate to form micro-void due to their tendency to cluster around the interface of primary Si. Increasing fatigue cracks nucleated from the debonded primary Si dominate the damage behavior at high temperature. In the second aspect, the refinement of Si increases the critical stress/strain for the crack initiation in the UT alloy. The refined microstructure in UT alloy also enhances deformation reversibility and diminishes the fatigue damage and further delays the cracking during cyclic loading (increasing of \( \beta \) as shown in Fig. 12c). The homogeneity of fatigue damage increases at higher temperature with the decrease of phase size and then the fatigue life increases.

As discussed above, the variation of model parameters are derived from the damage mechanisms, leading to the different performances (fatigue crack initiation and propagation). The increase of \( W_0 \) and \( \beta \) is beneficial for fatigue performance optimization. However, the two parameters show an obvious inverse relation for the most alloys, which will restrict the improvement of LCF life (Fig. 1a). The inverse relation also can be found in the evolution of \( W_0 \) and \( \beta \) with the increasing temperature (Fig. 12b, also see Table 1). Due to the inverse relationship of fatigue parameters and corresponding damage behavior, the complex fatigue life evolution with \( T \) will be formed (Fig. 12d). Fortunately, a reasonable distribution of two parameters (such as increasing \( W_0 \) at low temperature and increasing \( \beta \) at high temperature for the alloy) is definitely a way to improve the overall fatigue performance when the two parameters cannot be increased synchronously. In fact, in this study, the fatigue life is improved by increasing \( W_0 \) at low temperature (at 280 °C, increasing \( W_0 \) from 1.8 MJ/m³ of AC to 12.88 MJ/m³ of UT, Table 1) and increasing \( \beta \) at high temperature (at 420 °C increasing \( \beta \) from 1.66 of AC to 1.78 of UT). It can be seen that the comprehensive optimization of LCF properties can be achieved by adjusting the parameters \( W_0 \) and \( \beta \) based on hysteretic energy model (Fig. 12d). Superior thermo-mechanical fatigue performance also can be expected by enhancing the LCF resistance in the all temperatures.

### 5. Conclusions

An energy-based LCF life prediction model was used to evaluate the LCF life of the cast Al-Si piston alloys at elevated temperatures. Comprehensive investigations focusing on the damage behaviors including fatigue life prediction and damage mechanisms (fatigue crack, propagation) of the Al-Si piston alloys at different temperatures were researched. Based on the damage behavior and model analysis, the appropriate method to enhance the LCF properties of the Al-Si alloy was reported and several conclusions can be given below:

1. Based on the hysteresis energy-based fatigue life prediction model, the relationship between fatigue life and temperature has been proposed. Based on the model, the increase of fatigue toughness \( W_0 \) and fatigue damage exponent \( \beta \) is beneficial for the improvement of LCF life. However, the \( W_0 \) and \( \beta \) show an obvious inverse relation for the alloy with temperature varying, which originates from the transition of microscopic damage mechanisms (fatigue crack initiation and propagation for LCF).

2. For the fatigue damage, the cracks initiate mainly from primary Si cracking induced by piling-up of dislocations at low temperature and phase/matrix interface debonding.
induced by vacancy accumulation at higher temperature. The different crack initiation mechanisms at particles are mainly affected by two factors: dislocations slip mode for damage accumulation (influenced by temperature) and the critical stress/strain for the crack initiation (influenced by microstructure).

(3) Due to the development of dislocation configurations and microstructures, the AI matrix is obviously refined after LCF with increasing temperature. The fracture appearance is transgranular for all the tested temperatures; the grain boundary is not the main damage source even at high temperature. The microstructure refinement enhances the plastic deformation homogeneity and resistance to crack propagation, the damage tolerance is improved (reflected by increasing $W_0$). Due to the inverse relationship of the two fatigue parameters caused by corresponding damage behavior, the complex fatigue life evolution appears, however, at a critical fatigue testing temperature ($T_{cr}$), the maximum fatigue life can be found.

(4) A reasonable distribution of two parameters (increasing $W_0$ at low temperature and increasing $\beta$ at high temperature) is an appropriate way to improve fatigue life in the whole service temperatures. The method (ultrasonic melt treatment) to enhance the LCF life of UT alloy with increasing the intrinsic fatigue toughness $W_0$ (e.g., optimizing the grain and phase sizes to increase the fatigue crack propagation resistance) at low temperature and increasing the fatigue damage exponent $\beta$ (e.g., refinement of primary Si particles to inhibit the micro-crack initiation) at high temperature was introduced. The improved LCF life of UT alloy was achieved.

Conflicts of interest
The authors declare no conflicts of interest.

Acknowledgments
The authors would like to thank Mr. J. L. Wen, Dr. C. H. Li for their helps of the fatigue experiments and SEM observations. The authors would like to thank Prof. J. P. Li and Prof. Y. C. Guo and Mr. F. Xian for ingot preparation. This work is supported by National Natural Science Foundation of China (NSFC) under Grant Nos. 51871224 and 51331007.

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