Effects of quenching and tempering heat treatment on microstructure, mechanical properties, and fatigue crack growth behavior of 51CrV4 spring steel

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Keywords: spring steel, mechanical properties, fatigue crack growth behavior, heat treatment

Abstract
Herein, the microstructure, mechanical properties, and fatigue crack growth (FCG) behavior of 51CrV4 spring steels under different quenching and tempering heat treatment were studied. Results show that tempering temperature would not affect the prior austenite grain (PAG) size, while quenching temperature has a great influence on it; the volume fraction of carbide precipitates decreased with increasing quenching temperature. The width of martensite lath has little relationship with the prior austenite grain size. The mechanical properties of the investigated steels decreased with an increasing tempering temperature. Tempering temperature had a greater influence on the fatigue crack growth behavior compared with quenching temperature. The FCG of all the investigated steels were well interpreted by Paris model. When the quenching temperature was 840 °C, the stable expanding stage of the fatigue fracture was representative fatigue striations, fatigue steps and secondary cracks; multiple failure mechanisms coexisted in the rapid expanding stage; the fracture morphology of unstable failure stage was mainly dimples. When the quenching temperature increased to 880 °C and 920 °C, quasi-cleavage was observed in all the stages compared with the samples quenched at 840 °C. Secondary cracks, fatigue steps and crack closure could retard the FCG rate by consuming the driving force near the tip of fatigue crack.

1. Introduction

Spring steel is widely used in transportation and construction machinery due to its excellent comprehensive performance. 51CrV4, one of the most common used spring steels, shows promising mechanical properties and fatigue properties enhanced by the addition of micro alloying elements. Fatigue is a serious failure mode of spring steel, while the fatigue performance of steel is affected by many factors, such as microstructure, heat treatment, loading condition and so on. Therefore, it is necessary to clarify the fatigue crack growth (FCG) mechanism of 51CrV4 steel to prevent fatigue failure.

Many researchers [1–3] have studied the effects of heat treatment process on FCG behavior of steels. Liang et al [1] found that the coarsening martensite laths and precipitates in a low activation martensite steel would induce rough fracture surface, and then the crack closure induced effective driving force decreased. However, the FCG rate in the new threshold area after a long-time tempering was lower. Li et al [2] concluded that the FCG uniformity and roughness of fatigue fracture of AISI4340 steel increased with an increasing tempering temperature, which could effectively reduce the fatigue crack growth rate. In addition, some scholars explained the growth of fatigue crack by microscopic observation and analysis. Kevin et al [3] studied the high cycle fatigue damage of tempered martensite steel by in situ crack observation, and found that the early crack growth was...
driven by different damage mechanisms, which occurred in the interior and boundaries of martensite blocks. However, few studies focus on the influence of heat treatment on FCG behavior of 51CrV4 spring steel. In the present research, five different heat treatment processes were conducted on the experimental 51CrV4 steel. The microstructure, mechanical properties and fatigue crack growth rate under different conditions were tested and analyzed. The fatigue fracture morphology was observed by scanning electron microscopy (SEM). Paris model was adopted to describe the FCG rate.

2. Materials and methods

The investigated material is a 2 mm thick 51CrV4 spring steel produced by XinSteel Industrial Co., Ltd. The main chemical compositions are detailed in Table 1. The transformation temperature Ac3 measured by Gleeble3800 thermal simulation testing machine was 836 °C. Five different heat treatment processes (as shown in Table 2) were carried out according to Figure 1. Firstly, the 51CrV4 steel was austenitized at 840 °C ~ 920 °C for 20 min, then quenched by water, and then re-heated to 380 °C ~ 460 °C for 90 min. Finally, the samples were quenched to room temperature by water. In order to ensure the accuracy of the heating parameters, the box-type resistance furnace was firstly heated to a temperature lower than the set temperature by 20 °C, then put the sample into the heating furnace, and started the time when the temperature raised to the set temperature. The entire heat treatment process was filled with nitrogen for protection.

The heat-treated samples after grinding and polishing were submerged in constant temperature water solution with a certain proportion of supersaturated picric acid at 70 °C for 7 ~ 10 min, then the corroded surface of the samples was washed, and the prior austenite grain was observed by a metallographic microscope and calculated by Imagetool software.

The microstructure of steels after heat treatment was observed by SEM after etched with a 4% nitric acid alcohol solution. The martensite lath and precipitate were observed by transmission electron microscope.

Table 1. The main chemical compositions of 51CrV4 steel (wt. %).

| C   | Si  | Mn  | P    | Cr  | Ni  | Mo  | V   | Ti  |
|-----|-----|-----|------|-----|-----|-----|-----|-----|
| 0.52| 0.22| 0.92| 0.001| 1.01| 0.02| 0.019| 0.121| 0.005|

Table 2. The five designed heat treatment processes for the 51CrV4 steels.

| Sample | Austenitizing temperature, °C | Austenitizing time, min | Tempering temperature, °C | Tempering time, min |
|--------|-------------------------------|------------------------|--------------------------|---------------------|
| 840–380| 840                           | 20                     | 380                      | 90                  |
| 840–420| 840                           | 20                     | 420                      | 90                  |
| 840–460| 840                           | 20                     | 460                      | 90                  |
| 880–380| 880                           | 20                     | 380                      | 90                  |
| 920–380| 920                           | 20                     | 380                      | 90                  |
TEM samples were firstly mechanically polished to 50 mm thick, and then ion polished in a 5 kV Gatan691 precision ion polishing system (PIPS).

The hardness of the samples was obtained by Rockwell hardness tester, and each sample was measured at least five times. Samples for uniaxial tensile test were prepared according to GB/T 228.1–2010, and the gauge is 50 mm, and the tensile tests were performed on a universal tensile testing machine.

The fatigue crack growth rate tests of the heat-treated 51CrV4 steels were conducted on an MTS810 machine at room temperature. The compact tensile (CT) samples were designed according to GB/T 6398–2017 standard, as shown in figure 2. The load was constant amplitude sinusoidal wave under a frequency of 10 Hz, the maximum load was 1 KN, the stress ratio was 0.1. After the samples were broken, the morphology of the fatigue fracture was analyzed by SEM.

Figure 2. Sample size for fatigue crack growth rate test. (Unit: mm).

Figure 3. Optical microstructure of prior austenite grain after different heat treatments: (a) 840–380, (b) 840–420, (c) 840–460, (d) 880–380, (e) 920–380.
3. Results and discussions

3.1. Microstructure

Typical optical microstructure of prior austenite grain (PAG) after different heat treatment processes is shown in figure 3. Figure 4 summarized the effects of heat treatment processes on prior austenite grain size, which was measured by linear intercept measurement. At least 200 grains of each sample were measured on the optical micrographs and the average value was obtained by statistics. It can be seen that the PAG size did not change obviously with an increasing tempering temperature when the quenching temperature was fixed at 840 °C. However, the PAG size increased rapidly with an increasing quenching temperature when the tempering temperature was fixed at 880 °C. Figure 5. SEM micrograph of the 51CrV4 steels at different magnification: (a), (b) 840–380; (c), (d) 840–420; (e), (f) 840–460; (g), (h) 880–380; (i), (j) 920–380; (k) EDS spectra of encircled region in (f).

![Figure 4. Effects of heat treatment processes on prior austenite grain size.](image)

![Figure 5. SEM micrograph of the 51CrV4 steels at different magnification: (a), (b) 840–380; (c), (d) 840–420; (e), (f) 840–460; (g), (h) 880–380; (i), (j) 920–380; (k) EDS spectra of encircled region in (f).](image)
temperature was the same. The PAG size increased about three times when the quenching temperature increased from 840°C to 920°C. It is concluded that tempering temperature would not affect the prior austenite grain size, while quenching temperature has a great influence on it.

Figure 5 shows the SEM images of the investigated steels. As can be seen, many fine precipitates and coarse carbides precipitates (confirmed by EDS in figure 5(k)) with ellipsoidal and irregular shapes mainly distributed in the prior austenite matrix and grain boundaries. The major microstructure evolution during the tempering process is the decomposition of martensite, the precipitation and growth of carbides, and the recrystallization of α solid solution. If the tempering time is not long enough, less carbides will be precipitated. With the extension of tempering time, the martensite gradually decomposed, and the supersaturated carbon atoms were gradually

Figure 6. (a) Bright field and (b) dark field images of TEM for sample 840–380; (c), (d) TEM micrographs of sample 840–460 at different magnification; (e) TEM images and corresponding diffraction pattern of sample 920–380; (f) EDS spectra of encircled region in (e). PAGB: prior austenite grain boundary.
de-dissolved and precipitated in the form of carbides, and distributed in the matrix. The supersaturated carbon atoms in the martensite gradually de-dissolve to form carbides with the increase of the tempering temperature, and the supersaturated carbon concentration in the matrix was greatly reduced, the matrix gradually recovered, and the lamellar cementite was spherized, the content of spherical cementite increased significantly. It is also found that as the quenching temperature increased, the volume fraction of carbide precipitates decreased, resulting in the decrease of the grain boundary migration resistance. Therefore, the PAG size increased with an increasing quenching temperature.

The TEM images in figures 6(a)–(e) show that the main substructure in the experimental steel was high density dislocation, the EDS and SAED results in figures 6(e)–(f) indicated that the carbide precipitate in the steel was Cr$_7$C$_3$. As shown in figure 6(a), the width of martensite lath was about 72 nm, and the lath width in other samples were about the same. Therefore, the width of the lath has little relationship with the PAG size. Many other studies also confirmed this phenomenon [4] because the lath width is dominated by the martensite nucleation rate, which is affected by the chemical composition, hence the lath width is steady for a certain steel [5]. Coarse precipitates were observed both in the boundaries (figure 6(a)) and interior (figure 6(c)) of the prior grain. Dark field micrographs of the carbides in sample 840–380 (figure 6(b)) show that many fine carbides precipitated in the matrix, and some carbides precipitated along the lath interfaces. Figure 6(d) revealed that the precipitates were surrounded by high density dislocations, indicating that the precipitates are effectively to hinder the movement of dislocations and strengthen the matrix.

### 3.2. Mechanical properties

Figure 7 shows the yield strength (YS), ultimate tensile strength (UTS) and elongation (EL) of the samples under different heat treatment processes. When the quenching temperature was the same, the UTS and YS of the experimental steel decreased with the tempering temperature changed from 380 °C to 460 °C, but the EL did not change much. This is because the activity of saturated carbon atoms in martensite increased with the tempering temperature, making martensite easier to decompose. So, the solid solution strengthening effect of C atoms was gradually weaken, resulting in the decrease in strength. The dislocation density was also reduced during tempering, which will cause the above changes in mechanical properties. When the tempering temperature was the same, the UTS, YS and EL of all the investigated steels show a trend of first increasing and then decreasing.
when the quenching temperature increased from 840 °C to 920 °C. As the quenching temperature increased, the carbides dissolved and the solid solution strengthening effect increased, resulting in the strength increased accordingly. However, excessive heating temperature (920 °C) would cause the coarsening and decarburization of austenite grains and martensite packets, which had an adverse effect on the strength.

Generally, the relationship between yield strength and the strengthening mechanism of microstructure can be expressed as follows [6]:

\[ \sigma_y = \sigma_{Fe} + \sigma_{ss} + \sigma_p + K_L \frac{L}{d} + K_D \frac{D^{1/2}}{d} \]  \hspace{1cm} (1)

\[ \sigma_p = \frac{0.538 G b f^{1/2}}{d} \ln \left( \frac{d}{2b} \right) \]  \hspace{1cm} (2)

\[ \sigma_{ss} = 1723 \sqrt{\frac{f}{C}} + 5.8[Cr] + 18[Mo] + 105[Si] + 45[Mn] + 37[Ni] + 4.8[V] \]  \hspace{1cm} (3)

Where, \( \sigma_y \) is yield strength, \( i_p \), \( \sigma_{ss} \), \( K_L (L)^{-1} \), and \( K_D \frac{D^{1/2}}{d} \) is the strengthening effect caused by solid solution, second-phase precipitation, lath width and dislocation, respectively. \( \sigma_{Fe} \) is the strength of annealed pure iron, \( K_L \) is the coefficient of lath width, and \( L \) is the width of lath [7, 8].

According to the solubility product equation (4) of VC in austenite [9], VC will be completely dissolved in the matrix when the austenitizing temperature is above 926 °C. When the dislocation slip follows Orowan mechanism, equation (5) [10] shows that the strengthening effect of second phase is the largest at 920 °C and the smallest at 840 °C for the investigated steel. However, with the decrease of the size and volume fraction of carbide precipitates and the increase of austenitizing temperature, the second phase particles cannot control the PAG size in the matrix, resulting in the increase of martensite block size, and the strength of martensite decreased. The weakening of martensite strength was caused by the dislocation movement promoted by phase interface.

\[ \ln \left( \frac{f}{C} \right) = 6.72 - \frac{9500}{T} \]  \hspace{1cm} (4)

\[ \Delta \sigma_p = 8.995 \times 10^3 \frac{f^{1/2}}{d} \ln (2.417d) \]  \hspace{1cm} (5)

Where, \( f \) is volume fraction (%) of the precipitated phase, and \( d \) is the average size (nm) of precipitates.

Figure 8 shows the Vickers hardness of the 51CrV4 steels under different heat treatment processes. When the tempering temperature changed from 380 °C to 460 °C, the hardness of the steels was 47.28 HRC, 43.86 HRC and 41.46 HRC, respectively. It is concluded that when the quenching temperature is the same, the hardness of the steels decreased with an increasing tempering temperature. The hardness of tempered martensite steel is related to various microstructure characteristics, such as grain size, precipitates, and solid solution strengthening [11]. Results show that the volume fraction of carbide precipitates in the investigated steels decreased with an increasing tempering temperature while the quenching temperature was fixed, indicating that the hardness of the steels is related to the second phase particles. When the quenching temperature increased from 840 °C to 920 °C, the hardness was 47.28 HRC, 48.08 HRC and 47.62 HRC, respectively. The results show that the hardness variation of the steel was the same as the strength, which increased first and then decreased with the increase of quenching temperature.
3.3. Fatigue crack growth behavior

Figure 9 shows the curves of cyclic loading cycles \((N)\) and fatigue crack length \((a)\). It can be seen that the crack length increased with the increase of loading cycles, and the rate of increase was getting faster and faster. The fatigue crack growth rate \((\frac{da}{dN})\) increased with the increase of \(a\). When \(a\) reached near the critical crack size, \(\frac{da}{dN}\) increased sharply, the crack propagated unsteadily and result in final fracture of the samples. It is also found that \(\frac{da}{dN}\) is related to the heat treatment process of the sample because of the evolution of microstructural characteristics. The quenching temperature had little effect on the FCG behavior. As the quenching temperature increased, the fatigue life decreased slightly, and the FCG rate increased slightly. This is because the amount of the grain boundaries reduced for larger size of prior austenite grains at higher quenching temperature, resulting in the increase of grain boundary dislocations density and the decrease of crack propagation resistance. Therefore, it is more likely to lead to intergranular cracking. However, the tempering temperature had a greater influence on the fatigue crack growth behavior. As the tempering temperature increased from 380 °C to 460 °C, the fatigue life and critical crack size increased. Therefore, the fatigue crack growth life of sample 840–460 was much longer than other samples.

Generally speaking, the fatigue crack growth process consists of three stages: slow expanding stage, stable expanding stage (Paris region) and rapid expanding stage. In the present study, the experiment is basically aimed at the stable expanding stage, which accounts for the majority of FCG life. Figure 10 shows the FCG rate curves of 51CrV4 steels. In the stable expanding stage, the stress intensity factor \((\Delta K)\) and the FCG rate \((\frac{da}{dN})\) are approximately linear in the double logarithmic coordinate, which is in line with the Paris model \([12]\), as show in equation (6):

![Figure 9. The relationship between fatigue cycles and crack length of 51CrV4 steels under different heat treatment processes.](image)

![Figure 10. FCG curves of 51CrV4 steels.](image)
Where, \( C \) and \( m \) are constants, and different materials have different values. Table 3 shows the calculated values of \( C \) and \( m \), together with the correlation coefficient of the samples under different conditions. The values of \( C \) and \( m \) at different conditions are different, indicating that these parameters are greatly affected by the microstructure of steel.

Figure 11 shows the fitted FCG curves based on Paris model. It is obvious that the calculated results of Paris model are coincide with the experimental data, and all the correlation coefficient are above 0.95. The FCG rate under the same \( \Delta K \) decreased with the increase of tempering temperature because martensite would gradually decompose, so the carbon content of martensite decreased, resulting in the improvement of the deformation coordination ability between martensite and ferrite as the plasticity of martensite increased \[13\]. The fatigue crack growth rate tests proved that such phase interface can effectively improve the fatigue resistance.

It is also observed that the FCG rate increased with an increasing of quenching temperature. Although the increase of quenching temperature can result in the coarsening of martensite size, it also contributed to the decrease of volume fraction of carbide precipitates, which is a crucial factor to control the FCG rate in the 51CrV4 steels. The frequency of crack-precipitate contact increased with the increase of precipitates volume fraction. Thus, the crack growth path would become more twisted and the FCG rate was slowed down \[14\]. In addition, the carbon content in the formed martensite is lower when the austenitizing temperature increased, so it would consume the driving force near the tip of fatigue crack through the passivation and deflection mechanism \[15, 16\].

### 3.4. Fatigue fractography analysis

Figure 12 shows the fatigue fractography of 51CrV4 steels under different heat treatment conditions observed by SEM. Figure 12(a1) shows the stable expanding stage of fatigue crack in sample 840–380, typical fatigue striations and many secondary cracks were observed. The crack tip opens and blunts during cyclic loading, and closes and sharpens when unloading \[17\]. Fatigue striations, which are usually perpendicular to the local FCG direction, are formed because of the above process, and it is repeated in the next cycle \[18, 19\]. Previous studies have shown that the striation spacing was related to the FCG rate, and the spacing increased as the distance from the starting point of the crack increased \[15\]. Dislocations irreversibly slipped and accumulated inside the grains under the action of cyclic stress, and many discontinuous micro-cracks were generated in the plastic zone near the crack tip. These secondary cracks grow up and connect to each other, and eventually merge with the main

![Fatigue crack growth rate curves described by Paris model](image)

**Figure 11.** Fatigue crack growth rate curves described by Paris model.

| Sample      | \( C \)     | \( m \) | Correlation coefficient, \( R^2 \) |
|-------------|-------------|---------|-----------------------------------|
| 840–380     | 3.52 \times 10^{-8} | 2.54    | 0.963                             |
| 840–420     | 4.96 \times 10^{-8} | 2.40    | 0.992                             |
| 840–460     | 1.95 \times 10^{-8} | 2.64    | 0.975                             |
| 880–380     | 5.19 \times 10^{-9} | 3.22    | 0.958                             |
| 920–380     | 3.32 \times 10^{-9} | 3.38    | 0.963                             |

\[
da / dN = C \cdot (\Delta K)^m \]  

(6)
crack, leading to transgranular fracture. In addition, the main crack may also split into several smaller cracks to form secondary cracks due to various obstacles during the crack growth process, which consume a large amount of energy and thereby reduce the fatigue growth rate. Fatigue steps were observed as well in figure 12(a1) due to the blocking effect of hard martensite and grain boundaries on fatigue cracks. Some scholars have proposed a crystallographic mechanism to interpret the occurrence of fatigue steps [20]. Grain misorientation can delay the propagation of cracks because the slip system of each grain is different, a considerable amount of energy is

| Samples  | stable expanding stage | rapid expanding stage | final unstable failure |
|----------|------------------------|-----------------------|-----------------------|
| 840-380  | (a1)                   | (a2)                  | (a3)                  |
| 840-420  | (b1)                   | (b2)                  | (b3)                  |
| 840-460  | (c1)                   | (c2)                  | (c3)                  |
| 880-380  | (d1)                   | (d2)                  | (d3)                  |
| 920-380  | (e1)                   | (e2)                  | (e3)                  |

Figure 12. SEM images of fatigue fractography for the samples under different heat treatment conditions.
required for the dislocations to move from one grain to another, resulting dislocations accumulation in front of the grain boundaries.

Figure 12(a2) indicates that the fatigue fractography of rapid expanding stage in sample 840–380 is different from that of Paris region. Multiple fracture mechanisms coexist, such as quasi-cleavage, secondary cracks, fatigue striations and dimples. Less secondary cracks were observed with the propagation of fatigue cracks. This is because $\Delta K$ is large enough to destroy the microstructure of the steel and grow along the main crack. Besides, the static fracture mechanism became more and more prominent as the stress field intensity factor increased. When the maximum $\Delta K$ reached the critical value, unstable fracture would eventually occur, and the fracture morphology was mainly dimples, as shown in figure 12(a3). The evolution of fracture morphology for 840–420 (figures 12(c1)–(c3)) and 840–460 (figures 12(d1)–(d3)) is basically the same as that of 840–380. In addition, crack closure phenomenon occurred in the rapid expanding stage of 840–420, as shown in figure 12(b2). Crack closure could retard the fatigue crack growth rate because it consumes the driving force when the two fracture surfaces of a crack contact with each other, resulting in seriously damaged but smooth fracture [21, 22].

Figures 12(d1)–(d3) shows the fatigue fractography of sample 880–380, quasi-cleavage was observed in all the stages compared with the samples quenched at 840 °C. And the quasi-cleavage mechanism became more and more prominent as $\Delta K$ increased. Figures 12(e1)–(e3) show the fractography evolution for sample 920–380, which was similar to that of sample 880–380, but the quasi-cleavage facet of 920–380 was larger. As mentioned in figure 4, PAG size increased about three times when the quenching temperature increased from 840 °C to 920 °C. Therefore, the resistance of the grain boundaries on fatigue cracks decreased with the increase of PAG size, and the volume fraction of carbide precipitates decreased with the increase of quenching temperature, so cracks were easier to rapidly propagate inside the grains to form a quasi-cleavage facet. It can also be seen from figure 12 that the size of the quasi-cleavage facet of 880–380 and 920–380 was approximate to the statistical grain size in figure 4.

Figure 13 shows the microstructure near the fatigue fracture in Paris region of the investigated steels. Secondary cracks were obviously observed near the fracture for all the samples. Since the initiation and propagation of secondary cracks would consume energy, the growth of main cracks can be slowed down. Besides, microstructure such as hard martensite and carbide precipitates would promote rough or serrated fracture due to the deflection of crack. Therefore, the crack closure mechanism induced by roughness may be caused by the premature contact of crack surface [23]. In addition, second-phase particles were observed along the crack path in different samples which could cause more obstacles to the crack growth. Many micro-cracks and voids were also observed near the fracture because of the plastic deformation near the crack tip.

4. Conclusions

The microstructure, mechanical properties, and fatigue crack growth behavior of 51CrV4 spring steels under different quenching and tempering heat treatment were studied. The main conclusions are as follows:

1. Tempering temperature would not affect the prior austenite grain (PAG) size, while quenching temperature has a great influence on it. The volume fraction of carbide precipitates decreased with the increase of quenching temperature.
(2) The strength and hardness of the investigated steels decreased as the tempering temperature increased, while the strength and hardness showed a trend of first increasing and then decreasing as the quenching temperature increased.

(3) Tempering temperature had a greater influence on the fatigue crack growth behavior compared with quenching temperature. The fatigue crack growth (FCG) rate under the same $\Delta K$ decreased with the increase of tempering temperature. The FCG of all the investigated steels can be well described by Paris model.

(4) When the quenching temperature was 840 °C, the stable expanding stage of the fatigue fracture was typical fatigue striations, fatigue steps and secondary cracks; multiple fracture mechanisms coexisted in the rapid expanding stage; the static fracture mechanism became more and more prominent as the stress field intensity factor increased, and the fracture morphology of unstable failure stage was mainly dimples. When the quenching temperature increased to 880 °C and 920 °C, quasi-cleavage was observed in all the stages.

Acknowledgments

The authors gratefully acknowledge the financial support from National Natural Science Foundation of China (No. 51804137), China Postdoctoral Science Foundation (No. 2020T130556), Natural Science Foundation of Jiangxi Province (No. 20202BAB214004), Postdoctoral Selection Funding Project of Jiangxi Province (No. 2019KY55) and Program of Qingjiang Excellent Young Talents, Jiangxi University of Science and Technology (JXUSTQJYX2020011).

Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Author contributions

Conceptualization: Chaobin Lai and Ruiguo Lv. Data curation: Chaobin Lai and Shengci Li. Formal acquisition: Ruiguo Lv and Jianlin Yu. Investigation: Wei Huang and Tao Liu. Methodology: Wei Huang and Xiaoyong Tang. Project administration: Shengci Li and Ruiguo Lv. Writing—original draft: Wei Huang and Shengci Li. Writing—review and editing: Chaobin Lai.

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