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Abstract

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Keywords

properties, tensile, toughness, fracture, processes, treatment, heat, post, forging, hot, impact, steel, enhancing, cast, microalloyed

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Enhancing impact fracture toughness and tensile properties of a microalloyed cast steel by hot forging and post-forging heat treatment processes

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Abstract

In the present work, the effects of hot forging and post-forging heat treatment on the impact fracture toughness and tensile properties of a microalloyed cast steel were investigated. Mechanical tests were used to evaluate the room temperature impact fracture toughness and tensile properties of the steel. The resulting microstructures were analysed by optical microscope (OM), scanning electron microscope (SEM) and transmission electron microscope (TEM). The extraction replica technology was used to investigate the characterisation of complex precipitates formed during heat treatment. The obtained results showed that the coarse-grained microstructure of the forged specimen was significantly refined after post-forging heat treatment. Presence of complex precipitates had a favourable effect on the formation of refined austenite grains, and consequently refined final microstructure. Hot forging was beneficial to enhance the impact fracture toughness and tensile properties of the microalloyed cast steel. After 920°C-treatment followed by air cooling, the impact energy of the forged specimen was significantly increased from 19.3 to 208.3 J, and further enhancement in tensile properties was obtained. The enhanced impact fracture toughness and tensile properties of the microalloyed cast steel after hot forging and post-forging heat treatment were closely related to the refined and homogenised ferritic-pearlitic microstructure.

Keywords: Microalloyed steel; Forging; Heat treatment; Fracture toughness; Tensile properties

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1. Introduction

The demand for new materials that provide excellent structural performances while reducing weight and being cost-effectively manufactured is increasing. For applications with good mechanical properties requirements, microalloyed steels have been widely used [1,2]. Microalloyed steels are designed to provide better combinations of excellent fracture toughness and high strength than conventional carbon steels. Microalloyed steels trace their origin to the weldable unalloyed constructional steels containing up to 0.2% C and 1.8% Mn [3]. Small amounts of titanium (Ti), niobium (Nb) and vanadium (V) are introduced into these steels in the amount up to about 0.1 mass percent. The dispersive particles of the MX (M-Nb, Ti, and V; X-N and C) interstitial phases formed in the properly selected hot working conditions offer an important cost-effective approach to austenite grain size controlling and strengthening [4].

The work of Najafi [5] and Rassizadehghani [6] indicated that good combinations of strength and ductility of low-carbon cast steels can be achieved by microalloying additions. However, microalloying significantly decreased the impact toughness and led to the dominance of cleavage facets in the fracture surfaces. Cast steel products are produced by casting methods. Casting is a solidification process. Therefore, the microstructure can be finely tuned, such as grain structure, phase transformations and precipitation. However, defects such as shrinkage porosity, cracks and segregation are also intimately linked to solidification [7]. These defects can lead to lower mechanical properties. For eliminating such defects and achieving the desired mechanical properties, a subsequent hot working is often required before the steels can be applied. Steels used for structures, machine and vehicle parts are traditionally produced by hot forging, and the required properties are obtained by quenching and tempering treatments. This multi-stage process extends production time and leads to higher production costs. Instead, a simple process of hot forging followed by air cooling has been successfully developed on microalloyed cast steels to realise cost savings [8]. The

air-cooled microstructures of microalloyed forging steels, in contrast to the martensite of quenched and tempered steels, are generally composed of ferrite and pearlite [9]. Since the strength improvement of the microalloyed ferritic-pearlitic steels mainly depends on the increment of pearlite volume fraction, ferrite grain refinement, solid solution and precipitation strengthening, it is limited to obtain a higher level of strength, leading to an obstacle to the development of its potential uses [10]. Krauss [11] has shown that microalloyed forging steels with the same hardness have much lower fracture toughness compared to highly quenched and tempered steels. Consequently, the application of microalloyed forging steels has often been limited based on fracture toughness consideration at the required strength level [12].

In contrast to those studies performed on direct-cooled microalloyed forging steels [13], investigation with purpose of improving both impact fracture toughness and tensile properties of microalloyed cast steels by focusing on hot forging and as well as post-forging heat treatment conditions has been rare. For broadening the application of microalloyed forging steels, it is essential to study the mechanical properties of microalloyed steels in both hot forging and post-forging heat treatment conditions in order to find the optimal process correlated with the best combination of excellent fracture toughness and mechanical properties. In the present study, the effects of hot forging and post-forging heat treatment on the mechanical properties of a microalloyed cast steel were investigated. This study aims to acquire the best combination of excellent impact fracture toughness and tensile properties of the steel through controlled hot forging and post-forging heat treatment processes. The research has immediate practical applications, and the outcomes can be successfully applied in steel manufacturing industries.

2. Experimental procedure

The material used in this study was a microalloyed cast steel. The chemical compositions are shown in Table

1. The cast steel was heated to 1230°C for 1 h austenitisation and then hot forged to be a cube with section size of 140 mm × 140 mm and height of 400 mm followed by air cooling. The final forging temperature was controlled to be higher than 950°C (Fig. 1). Prior to heat treatment, dilatometer tests were carried out to determine the austenitisation starting (A_{c1}) and finishing (A_{c3}) temperatures in heating process. For annealing heat treatment, two groups of rectangular specimens with sizes of 11 mm × 11 mm × 60 mm and 11 mm × 11 mm × 110 mm respectively were cut from the forged piece. Specimens were heat treated at 840, 860, 880, 900, 920 and 950°C for 1 h using an electric furnace in laboratory followed by air and furnace cooling, as illustrated in Fig. 1.

Specimens were metallographically polished and etched with 2% nital solution for microstructural observation by an OLYMPUS BX51M optical microscope (OM). Slightly overetched specimens were examined under JEOL JSM-6300F field-emission scanning electron microscope (FE-SEM) to present the detailed microstructure at relatively high resolution. For observing prior austenite grain boundaries, specimens were heated to 1230°C and various annealing temperatures for 1 h followed by water quenching. The water quenched specimens were etched with a “picric acid + FeCl₃ + dodecyl benzene sulfonic acid sodium salt” solution to reveal prior austenite grain boundaries, and then observed by OM. To examine the characteristic and composition of precipitates, extraction replicas as well as thin foils mounted on Cu grids specimens were prepared by the following steps: (i) polish samples with SiC sand papers; (ii) release surface residual stress of the samples with a “methanol + acetylaceton + tetramethylammonium chloride” solution; (iii) do the carbon coating using Vacuum Evaporator JEE-420, JEOL; (iv) separate the carbon foils with a “HCl + methylalcohol” solution; and (v) mount the thin foils on Cu grids. Replica specimens were analysed using energy dispersive X-ray spectroscopy (EDS) method on JEOL JEM-2100F transmission electron microscope (TEM) operated at 200 kV.

Standard Charpy impact specimens were machined according to ASTM E 23. The Charpy V-notched specimens with cross section of 10 mm × 10 mm, length of 55 mm, notch angle of 45° and notch depth of 2 mm were employed to study the room temperature impact fracture toughness on a Zwick/Roell impact tester. Tensile specimens were prepared according to ASTM E 8M (Gauge length 30 mm, gauge diameter 6 mm). Tensile tests were performed on INSTRON 8801 at room temperature with strain rate of $5 \times 10^{-3} \text{ s}^{-1}$. The resulting stress-strain curves were analysed to determine the yield strength (YS, 0.2% proof stress) and ultimate tensile strength (UTS). Fracture surfaces of the tested specimens were examined on SEM using an accelerating voltage of 15 kV to determine the failure mode.

3. Experimental results

3.1. Microstructures

Fig. 2 shows the OM structure of the as-cast specimen. As demonstrated, dendrite was caused during casting process. The microstructures of the forged specimens are shown in Fig. 3. It can be seen that the forged specimen consists mainly of coarse polygonal ferrite (PF) and pearlite, companied with small quantity of acicular ferrite (AF).

Fig. 4 shows the microstructures of the heat-treated specimens. It can be seen from Fig. 4a that ferrite non-uniformly distributes in the microstructure of the 840°C-treated specimen. When the temperature is increased to 920°C, the microstructure consists of equiaxed ferrite and pearlite, and the ferrite grain size distribution is very uniform (Fig. 4b). With further increasing temperature from 920 to 950°C, abnormal growth of some grains occurs, as shown in Fig. 4c. Moreover, furnace cooling induces coarser grains in contrast to that followed by air cooling at the same temperature, as presented in Fig. 4b and d. According to

Figs. 3a and 4b, it is apparent that 920°C-treatment followed by air cooling has significantly decreased the pearlite colony size, and as well as refined and homogenised the microstructure. After determined by using circular intercept method according to ASTM E112, the average ferrite grain size of the 920°C-treated specimen suffered from air cooling is obtained to be 10.0 μm .

3.2. Impact fracture toughness and tensile properties

The impact energies of the as-cast and forged specimens are obtained to be 5.4 and 19.3 J, respectively, from Charpy impact tests. Obviously, the impact fracture toughness of the as-cast steel is very poor. Although the impact energy of the as-cast steel has been increased after hot forging, the fracture toughness level is still very low. Fig. 5 shows the dependence of impact energy on temperature of the specimens after 1 h heat treatment followed by both air and furnace cooling. It can be seen that the impact energy of the specimen subjected to air cooling is higher than that cooled in furnace. The impact energy of the air-cooled specimen increases gradually at first, then to a maximum value at the temperature of 920 °C, and it finally drops when the temperature is higher than 920 °C. The highest impact energy under air cooling condition is obtained to be 208.3 J, which is about 38 times higher than that of the as-cast one.

The heat-treated specimen with the highest fracture toughness level was employed to do the tensile tests. Table 2 lists the tensile properties of the as-cast, forged and heat-treated specimens evaluated from room temperature tensile tests. It can be seen that both the tensile strength and strain have been significantly increased after hot forging. Post-forging heat treatment followed by air cooling induces further increase in YS and strain. Differently, heat treatment shows little effect on the increment of UTS.

3.3. Fractography

Selected fracture surfaces in the shear lip areas of some impact test specimens are demonstrated in Fig. 6. It can be seen in Fig. 6a that the fracture surface of the forged specimen is composed of cleavage facets and shear cracks, which indicates the brittle nature of fracture. The fracture mode has been remarkably changed after post-forging heat treatment. As shown in Fig. 6b, the fracture surface of the 920°C-treated specimen followed by air cooling is composed of small, deep and uniformly distributed dimples, which shows a characteristic of ductile fracture. Larger dimples exist in the fracture surface when the specimen is heat treated at 950°C in contrast to that at 920°C, as shown in Fig. 6c.

Fig. 7a, b and c show the tensile fracture morphologies of the as-cast, forged and 920°C-treated specimens, respectively. As demonstrated in Fig. 7a, the as-cast specimen exhibits a brittle cleavage fracture. Although hot forging has changed the fracture mode, big and non-uniformly distributed dimples exist in the fracture surface (Fig. 7b). After 920°C-treatment followed by air cooling, the fracture surface is still composed of dimples (Fig. 7c). Differently, the dimples are smaller and deeper, and the distribution of dimples is more uniform in contrast to that in the forged specimen, as shown in Fig. 7b and c. According to Fig. 7a, b and c, it is clear that post-forging heat treatment has a favourable effect on the generation of small, deep and uniformly distributed dimples after tensile tests.

4. Discussion

4.1. Microstructural evolution

Ferrite grain size is closely related to austenite grain size and cooling rate by the following relation [14]:

$$d_{\alpha} = 11.7 + 0.14d_{\gamma} + 37.7R^{-1/2} \quad (1)$$

where d_{α} is the ferrite grain size (mm), d_{γ} is the austenite grain size (mm), and R is the linear rate of cooling in degree centrifuged per minute. In this study, the cast steel was hot forged and then followed by air cooling,

and the heat-treated specimens were cooled in air and furnace. Under the same air cooling condition, the ferrite grain sizes of both the hot forged and heat treated specimens should increase in linear manners with the austenite grain size based on Eq. (1).

According to the results of dilatometer tests, the A_{c1} and A_{c3} temperatures of the studied steel were determined as 766 and 903°C, respectively. Holding specimens at temperatures higher than 903°C induces complete austenitisation, and prior austenite grain boundaries of the specimens water-quenched from these temperatures can be revealed. Fig. 8a and b present the prior austenite grains of the water-quenched specimens after holding at 920 and 1230°C for 1 h, respectively. As can be seen in Fig. 8a, the austenite grain size of 920°C-treated specimen is small. Increment of austenitising temperature induces coarser austenite grains, as shown in Fig. 8b. The average austenite grain sizes of the 920°C- and 1230°C-treated specimens were determined as 14.6 and 158.6 μm , respectively, by using circular intercept method according to ASTM E112. Based on Eq. (1), finer ferrite grains are expected to form in the 920°C-treated specimen than that in the 1230°C-treated specimen. According to Figs. 3a, 4b and 8, it is apparent that refinement of austenite grain size in microalloyed steels is critical in producing fine-grained ferritic-pearlitic microstructures.

The addition of microalloying elements can form carbides, nitrides or carbonitrides. These precipitates very effectively restrict austenite grain growth [15]. From a processing standpoint, the forged steels are reheated to a high austenitising temperature and then deformed at elevated temperature prior to transformation during air cooling. Heat-treated steels are held at lower temperatures before cooling to produce the final microstructure. The reheating temperature, in combination with the steel composition, controls the extent of microalloying elements dissolution, and consequently, the amount of microalloying elements addition available for subsequent precipitation. The dissolution and precipitation behaviour of the microalloying

elements strongly influence the austenite grain size of microalloyed steels.

Fig. 9 shows the characteristic and composition of precipitates in the specimens held at 920 and 1230°C for 1 h. When holding at 920°C, coarse precipitates with irregular shape are found, as shown in Fig. 9a. Additional EDS analysis shows that the irregular particles contain Ti, Nb and V elements (Fig. 9c). Based on the compositions given in Table 1 and previous work [16,17], the precipitates are probably Ti, Nb and V carbides/nitrides (or carbonitrides). Moreover, the spectrum shows high Cu, Si and O peaks. The Cu peaks are from the Cu grid when preparing the replica specimens. The Si and O peaks may correspond to silica or silicic acid formed by oxidation and hydration of the Si in the steel during the electrochemical extraction [18]. With increasing austenitising temperature from 920 to 1230°C, the size of precipitates decreases, and just square precipitates exist in the specimen, as shown in Fig. 9b. The result of EDS analysis shows that these square precipitates only contain Ti and Nb, and the content of Nb in the precipitates is very low in contrast to that in the precipitates of the 920°C-treated specimen, as shown in Fig. 9c and d.

In microalloyed steels, vanadium precipitates have the lowest stability and readily dissolve on heating to forging temperatures. Niobium precipitates have higher stability, and therefore may be harder to dissolve during heating to forging temperatures. Titanium nitride TiN is stable at very high temperatures, and it usually precipitates in liquid steel or at high temperatures in the austenite regime [19]. According to the analysis results of Fig. 9, it is probable that TiN is the initial precipitate, and Nb and V mostly form on the outside layer of TiN to be a complex precipitate in the 920°C-treated specimen (Fig. 9a and c). In fact, because Ti, Nb and V carbides, nitrides or carbonitrides have similar crystal structures, co-precipitation is very common in multi-microalloyed steels [16,17]. When the reheating temperature is elevated to 1230°C, all V- and most Nb-containing particles dissolve from the outside layer of the complex precipitates and return to

the austenite solid solution. Subsequently, the size of precipitates decreases, as shown in Fig. 9b. According to previous work [20], decrement of the precipitates size is associated with significant reduction of the amount of precipitates when the austenitising temperature is increased. Therefore, based on Figs. 8 and 9, it can be deduced that complex precipitates have a favourable effect on the formation of refined austenite grains. Coarsening of austenite grains at 1230°C should be attributed to the dissolution of V- and Nb-containing particles and the reduction of the amount of precipitates.

For the heat-treated specimens, heat treatment at temperatures lower than 903°C induces incomplete austenitisation, leading to non-uniform distribution of ferrite grains. In contrast to the 920°C-treated specimen suffered from air cooling, growth of ferrite grains in the specimens heat-treated at high austenitising temperature or cooled in furnace is considered to be caused by the coarsening of austenite grains.

4.2. Enhancement of impact fracture toughness and tensile properties

There are three major property requirements for microalloyed steels: high yield strength, low impact transition temperature and good weldability [21]. Both the high yield strength and low impact transition temperature can be achieved through ferrite grain refinement. The quantitative relationship between the yield strength and grain size is known as the classic Hall-Petch equation [22,23]:

$$YS = \sigma_0 + k_y d^{1/2} \quad (2)$$

where YS is the yield strength, σ_0 is the friction stress needed to move individual dislocations, k_y is a constant (often referred to as the Hall-Petch slope and is material dependent), and d is the average grain size (d refers to the ferrite grain size in ferritic-pearlitic steels). According to Eq. (2), fine-grained structure has a pronounced effect on the increment of yield strength. In another supporting work, similar conclusion was obtained by Pickering [24] with an empirical equation for ferritic-pearlitic steels under 0.25% C.

Grain refinement in microalloyed steels simultaneously produces increased strength, fracture toughness and ductility, without affecting weldability, as shown in Fig. 10 [21]. The fracture toughness, YS and strain of the heat-treated specimens suffered from air cooling are significantly enhanced in contrast to the forged specimens. According to Figs. 3 and 4, and as well as the above analysis, these enhanced properties are closely related to the remarkable ferrite grain refinement.

During hot forging process, plastic deformation produces an increase in the number of dislocations, resulting in a higher state of internal stress. In V-microalloyed forging steels, ferrite nucleates and grows not only on austenite grain boundaries but also intragranularly on MnS particles within austenite grains. The resulting intragranularly formed AF (Fig. 3a) breaks up the massive pearlitic structures formed in a coarse-grained austenite and improve strength and fracture toughness [25,26]. After heat treatment, the internal stress was relieved from the forged steel, and the AF-containing microstructure was replaced by homogenised ferritic-pearlitic structure. Therefore, no significant increment in UTS after heat treatment in contrast to the forged specimens even though refine-grained microstructure, which can improve the strength, is obtained.

Among the heat-treated specimens, heat treatment at temperatures lower than 903°C induces incomplete austenitisation, leading to non-uniform distribution of ferrite after air cooling. Coarser ferrite grains will be caused if the austenitising temperature is higher than 920°C due to the increase of austenite grain size with temperature. These will consequently reduce the fracture toughness. Under furnace cooling condition, the coarsening of ferrite grains during the very slow cooling process is responsible for the reduced impact energy.

5. Conclusions

In the present study, the enhancement of impact fracture toughness and tensile properties of a microalloyed cast steel by hot forging and post-forging heat treatment processes was investigated. Following conclusions are drawn from the present work:

1. Under hot forging condition, the specimen exhibits a mixed microstructure of coarse polygonal ferrite, pearlite and small quantity of acicular ferrite. 920°C-treatment followed by air cooling significantly decreased the pearlite colony size, and as well as refined and homogenised the microstructure.
2. The morphology and composition of complex precipitates vary with austenitising temperatures. Presence of complex precipitates has a favourable effect on the formation of refined austenite grains at 920°C. Coarsening of austenite grains at 1230°C is attributed to the dissolution of V- and Nb-containing particles and the reduction of the amount of precipitates.
3. Hot forging is beneficial to increase the impact fracture toughness and tensile properties of the microalloyed cast steel. After 920°C-treatment followed by air cooling, the impact energy of the forged specimen is significantly increased from 19.3 to 208.3 J, and further enhancement in YS and strain is obtained.
4. In contrast to the forged specimen, the increased impact energy, YS and strain after post-forging heat treatment are closely related to the refined ferritic-pearlitic microstructure. Differently, grain refinement shows little effect on the increment of UTS. The impact fracture toughness and tensile properties of the microalloyed cast steel are significantly enhanced by obtaining the refined and homogenised microstructure after hot forging and post-forging heat treatment processes.

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Figure Captions:

Fig. 1. Schematic illustration of the hot forging and post-forging heat treatment processes.

Fig. 2. OM structure of the as-cast specimen.

Fig. 3. Microstructures of the forged specimens. (a) OM structure, and (b) SEM of pearlite.

Fig. 4. Microstructures of the heat-treated specimens subjected to (a,b,c) air and (d) furnace cooling. (a) 840°C, (b,d) 920°C, and (c) 950°C.

Fig. 5. Charpy impact energies of the heat-treated specimens.

Fig. 6. Fracture morphologies of the specimens after Charpy impact tests. (a) Forged, (b) 920°C + Air cooling, and (c) 950°C + Air cooling.

Fig. 7. Fracture morphologies of the specimens after tensile tests. (a) As-cast, (b) Forged, and (c) 920°C + Air cooling.

Fig. 8. Prior austenite grains of the water-quenched specimens after holding at (a) 920°C and (b) 1230°C for 1 h.

Fig. 9. Extraction replicas exhibiting complex precipitates observed in the specimens held at (a,c) 920°C and (b,d) 1230°C for 1 h. (a,b) Dark field images, and (c,d) EDS spectra.

Fig. 10. Effect of ferrite grain size on properties.

Table Captions:

Table 1 Chemical compositions of the studied steel (wt.%).

Table 2 Tensile properties of the specimens under different processing conditions.

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Fig. 8. Prior austenite grains of the water-quenched specimens after holding at (a) 920°C and (b) 1230°C for 1 h.

Fig. 9. Extraction replicas exhibiting complex precipitates observed in the specimens held at (a,c) 920°C and (b,d) 1230°C for 1 h. (a,b) Dark field images, and (c,d) EDS spectra.

Fig. 10. Effect of ferrite grain size on properties.

Table 1 Chemical compositions of the studied steel (wt.%).

C	Mn	Cr	Ni	Cu	Si	Al	V	Nb	Ti	P	S	N
0.17	1.2	0.1	0.02	0.02	0.47	0.02	0.01	0.02	0.02	0.01	0.01	0.006

Table 2 Tensile properties of the specimens under different processing conditions.

Specimens	YS (MPa)	UTS (MPa)	Strain (%)
As-cast	351.6	515.5	4.1
Forged	363.2	592.7	29.4
920°C + Air cooling	390.1	593.2	35.0