Effect of Al replacing Si on mechanical properties of high carbon Q–P–T martensitic steels

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ABSTRACT

Fe–0.67C–1.55Mn–1.99Al–0.60Cr–0.038Nb (Fe–2Al) and Fe–0.63C–1.52Mn–1.49Si–0.62Cr–0.036Nb (wt.%) (Fe–1.5Si) were designed and both subjected to novel quenching–partitioning–tempering (Q–P–T) process, and the effect of 2 wt.% Al replacing 1.5 wt.% Si on mechanical properties was studied. The Fe–2Al Q–P–T martensitic steel exhibited a tensile strength of 1640 MPa and elongation of 21.1% accompanied with product of strength and elongation (PSE) of 43.65 GPa %, which is superior to the third generation advanced high strength of 30 GPa %. While the tensile strength of Fe–1.5Si Q–P–T martensitic steel is 1.950 GPa, and the total elongation only is 12.4%, whose PSE is less than 30 GPa %. High ductility of Fe–2Al Q–P–T martensitic steel origins from that the dislocation absorption by retained austenite effect and high mechanical stability of retained austenite due to the addition of Al, which avoids the formation of more brittle strain–induced twin-type martensite from retained austenite transformation. The lower strength of Fe–2Al Q–P–T steel than Fe–1.5Si Q–P–T steel is attributed to lower elastic modulus of former steel after Q–P–T process.

KEYWORDS

Quenching–partitioning–tempering martensitic steel; Al replacing Si; mechanical property; retained austenite; strain-induced twin-type martensite

Introduction

Reduction in vehicle weight has been driven by the need to increase fuel efficiency and reduce CO₂ emissions [1,2]. In last decade novel advanced automotive steels have intensively been investigated as their superior strength and ductility allow less steel to be used hence reducing overall weight. A considerable amount of Al has been added to these ‘lightweight steels’ to achieve their excellent strength and ductility [3–5]. In addition to lightweight, automotive steels require a combination of specific strength and ductility for forming complex shapes as well as improving crashworthiness qualities. Structural reinforcement components such as A- and B-pillars, side sills, and front cross members require ultra-high strength above 1 GPa, together with good ductility (total elongation of 15–20%) [6]. In general, Fe–Mn–Si base-advanced high strength steels by Al element replacing or partially replacing Si element will evolve to lightweight steels, in which the weight percent of Al is usually higher than 3% accompanying with the existence of high temperature BCC-δ ferrite at room temperature, such as Fe–0.39C–0.77Si–1.50Mn–3.35Al [7]. If the weight percent of Al is usually less than 3%, high temperature BCC-δ ferrite cannot be kept at room temperature, such as Fe–0.4C–0.26Si–2.02Mn–2.50Al [7].

Many studies focus on the strength enhancement of lightweight steels due to the negative effect of Al on strength. However, the effect of Al replacing Si on mechanical properties and its mechanism have been not yet investigated. Therefore, in this work Fe–0.67C–1.55Mn–1.99Al–0.60Cr–0.038Nb (Fe–2Al) and high carbon Fe–0.63C–1.52Mn–1.49Si–0.62Cr–0.036Nb (wt.%) (Fe–1.5Si) for comparison were designed, in which except for 2Al replaces 1.5Si other elements are almost the same in weight percent. In order to obtain high strength over 1 GPa, high carbon content was considered, and Nb element was added to realize refining-grain strengthening and precipitation strengthening [8]. The addition of Mn and Cr elements is to stabilize the austenite [9]. Besides, quenching–partitioning–tempering (Q–P–T) process proposed by Hsu [10] in 2007 was used to treat the above high carbon steels. Since Q–P–T process absorbs the core idea of quenching and partitioning (Q&P) process proposed by Speer et al. in 2003 [11]: the determination of Tq based on ‘constrained carbon paraequilibrium’ (CCE) theory of Q&P process [11], Q–P–T steels can also obtain considerable carbon-enriched retained austenite by the partitioning of carbon. Q&P process involves: a fast quenching from austenitizing temperature to a quenching temperature (Tq) between start temperature (Ms) of martensitic transformation and
finish temperature \((M_f)\); then partitioning at (one-step) or above \(T_q\) (two-step) to obtain more carbon-enriched retained austenite at room temperature.

**Experimental procedures**

The designed steel was melted in a medium frequency furnace, and a hot-rolled plate with 20 mm thickness was prepared by the Central Iron and Steel Research Institute, Beijing, China. Specimens with 2 mm thickness were cut from the hot-rolled plate. After cutting, Fe–1.5Si specimens were subjected to the Q–P–T process, which involves austenitizing at \(830°C\) for 300 s, followed by quenching in a salt bath at \(160°C\) \((T_q)\) for 10 s, then partitioning and tempering at \(400°C\) for 60 s in molten salt, and finally quenching to room temperature in water. While Fe–2Al specimens were subjected to the Q–P–T process, involving austenitizing at \(890°C\) for 300 s, followed by quenching in a salt bath at \(160°C\) \((T_q)\) for 10 s, then partitioning and tempering at \(400°C\) for 600 s in molten salt, and finally quenching to room temperature in water.

Rectangular tensile specimens were cut to the gauge length of 15 mm, thickness of 1.5 mm, and width of 5 mm. Tensile tests were executed on Zwick/Roell Z100 universal testing machine at a strain rate of \(2 \times 10^{-3} \text{s}^{-1}\) at room temperature. Specimens for transmission electron microscopy (TEM) were prepared by mechanically polishing and then electropolishing in a twin-jet polisher using 4% perchloric acid and 96% ethanol from \(-20°C\) to \(30°C\). TEM was performed in a JEOL-2100F microscope operated at 200 kV. The volume fractions of retained austenite \(V_{RA}\) in the Q–P–T samples were measured by X-ray diffraction (XRD) with CuKα radiation using a D/max-2550 X-ray diffractometer based on a direct comparison method of the integrated intensity of the austenite \((200)_{\gamma}\), \((220)_{\gamma}\), and \((311)_{\gamma}\) peaks and the martensite \((200)_{\alpha}\) and \((211)_{\alpha}\) peaks. The average dislocation densities in both martensite and retained austenite were measured by XRD linear profile analysis (XLPA), and a detail description of XLPA see Refs [12,13].

**Results**

**Mechanical properties**

The tensile tests of Fe–1.5Si and Fe–2Al Q–P–T samples were performed, respectively, and their engineering stress–strain curves in Figure 1 indicate that Fe–1.5Si Q–P–T steel exhibits high tensile strength of 1.950 GPa and a total elongation of 12.4% accompanying product of strength and elongation (PSE) of 24.180 GPa % [9]. The tensile result of Fe–2Al samples shows a tensile strength of 1.500 GPa and a total elongation of 29.1% accompanying low PSE of 43.650 GPa %, reaching the PSE property (30 GPa %) of the third generation advanced strength steels.

**Microstructural characterization**

Based on the XRD spectra of Fe–1.5Si and Fe–2Al Q–P–T samples before tensile test in Figure 2, their microstructures are determined as BCC phase and FCC phase. The volume fraction of FCC phase is determined to be 29.1% in Fe–1.5Si Q–P–T sample and 25% in Fe–2Al Q–P–T sample. TEM characterization further indicates that the BCC phase in Fe–1.5Si sample consists of dislocation-type martensite plates and some thin plates of twin-type martensite embedded in them, as shown by the bright-field (BF) image in Figure 3(a), in which twin-type martensite is identified by the combination of dark-field (DF) images in Figures 3(b) and 4(b) and selected area electron diffraction (SAED) patterns inserted in them [9]. While the FCC phase is chunky-like retained austenite, as shown in Figure 3(c,d) for Q–P–T specimen. Moreover, the SAED patterns in Figures 3(d) and 4(d) both show \((11\overline{1})_\alpha//(11\overline{1})_\gamma\), \([11\overline{1}]_\alpha//(01\overline{1})_\gamma\), K–S orientation relationship and \((01\overline{1})_\alpha//(1\overline{1}1)_\gamma\).
overline{100}a // [101]g

The N–W orientation relationship between martensite and retained austenite [9]. TEM characterization reveals that Fe–2Al sample also consists of dislocation-type martensitic matrix but flake-like retained austenite between martensite laths, and there are (110)a // (111), (111), K–S orientation relationship and (011)a // (111)x, [100], // [101], N–W orientation relationship, such as shown in Figure 4.

**Discussions**

*Origin of Al lowering strength*

Carbon is the strongest element affecting strength among all element in present Fe–1.5Si and Fe–2.0Al steels. The carbon content (0.67 C wt.%) in Fe–2Al Q–P–T steel is slightly higher than that (0.63 wt.%) of Fe–1.5Si Q–P–T steel, but the tensile strength of the former is 1.500 GPa, much lower than 1.950 GPa of the latter. Since the Nb content of two steels almost is the same, the difference of strengthening effect from the Nb between Fe–1.5Si and Fe–2Al Q–P–T steels can be ignored. Therefore, it can imagined that Al replacing Si results in the decrease of tensile strength. Al replacing Si lowers the density of Fe–2Al Q–P–T steel comparing with Fe–1.5Si, but the slight decrease of density by Al replacing Si cannot markedly affect the strength of steel, and thus there must be other factor in affecting strength. Theoretical calculation [14] indicates that the single crystal elastic parameters $C_{11}$, $C_{12}$ and $C_{44}$ of BCC Fe–2.5wt% Al alloy are 270.71, 128.03 and 108.77 GPa, respectively, and those of BCC Fe–2.5wt% Si alloy are 277.10, 130.25 and 110.61 GPa, respectively. Since in Fe–2Al and Fe–1.5Si steels the amount of other elements is very close, and thus it can be considered that the elastic

**Figure 3.** TEM micrographs of undeformed Fe–1.5Si Q–P–T sample: (a) twinning martensite embedded in dislocation-type martensite matrix, BF image, (b) twinning martensite, DF image and inserted SAED pattern, (c) BF image of Q–P–T sample, (d) DF image of retained austenite and inserted SAED pattern of the retained austenite and martensite, (e) BF images, and (f) DF images and inserted SAED pattern of NbC.
parameter of Fe–2Al Q–P–T steel is less than those of Fe–1.5Si Q–P–T steel. It is evident that the lower elastic parameters of Fe–Al alloy than that of Fe–Si alloy implies the cohesive energy of Fe atom and Al atom is less than that of Fe atom and Si atom, and thus it is reasonably believed that the lower strength of Fe–2Al Q–P–T steel than Fe–1.5Si is mainly attributed to the lower cohesive energy of Fe atom and Al atom than Fe atom and Si atom.

**Origin of Al enhancing ductility**

In Fe–2Al and Fe–1.5Si Q–P–T steels there are plenty of retained austenite. As well-known, the ductility enhanced by retained austenite stems from not only transformation-induced plasticity (TRIP) effect [15] and blocking crack propagation (BCP) effect [16] proposed nearly 50 years ago, but also a new effect: dislocation absorption by retained austenite (DARA) effect [17]. TRIP effect enhances the ductility by strain-induced martensitic transformation, which relieves the stress within untransformed retained austenite and its adjacent martensite during deformation [18], and thus TRIP effect delays the necking. It is worthy to point that strain-induced martensite is brittle twin-type martensite, which harms greatly the ductility and toughness of steels, especially in Fe–1.5Si, the twin-type martensite induced by chunky retained austenite is large, and thus it more detrimental to ductility. The retained austenite blocks crack propagation after necking of tensile sample, i.e. the cracks either grow continuously in the austenite or trace a zig-zag path through interlocking martensite laths, which will require an increased energy for fracture [16]. Obviously, the above BCP effect does not contribute the uniform elongation, and only increases the elongation after necking of sample. We proposed DARA effect in medium-carbon Q–P–T steel based on the measurement of average dislocation densities in both martensite and retained austenite during deformation by XLP5. The DARA effect makes the martensite exist in a ‘softening’ state since retained austenite flakes absorb dislocations moving from nearby martensite laths, and the deformation ability of hard phase martensite is intensified [19]. After that, we proposed that two empirical conditions of DARA effect are both the sufficient amount (over 10% volume fraction) of retained austenite and the coherent or semi-coherent interface between martensite and retained austenite [20]. DARA effect was successively found in low-carbon Q–P–T steel [21] and low car carbon bainitic steel [20]. In this work these two high carbon steels also meet the conditions of DARA effect, and thus DARA effect should exist in these two high carbon Q–P–T steels. For example, the average dislocation densities in both martensite and retained austenite in Fe–1.5Si Q–P–T samples at different strain stages were measured by XLP5 [9], as shown in Figure 5. During deformation, both \( \rho_M \) and \( \rho_A \) are expected to increase with increasing strain owing to dislocation multiplication, but the XLP5 results were out of our expectation. The \( \rho_A \) in the retained austenite rapidly increase with raising the strain; however, \( \rho_M \) in martensite did not change in the same way. For instance, before tensile test, \( \rho_M \) were 11.44 × 10^{14} \text{m}^{-2}, and 10.80 × 10^{14} \text{m}^{-2} at 4% strain, which were less than those before tensile test.

![Figure 4](https://example.com/fig4.jpg)

**Figure 4.** TEM BF image of martensite laths and retained austenite flakes in Fe–2Al (a) and DF image of retained austenite as well as inserted SAED (b).

![Figure 5](https://example.com/fig5.jpg)

**Figure 5.** The variation of average dislocation density with strain in martensite or retained austenite in Fe–1.5Si Q–P–T steel.
Obviously, such a phenomenon cannot be explained by TRIP or BCP effects. During initial deformation, the $\rho_M$ in martensite reduces with increasing strain, indicating that the amount of dislocation transported to retained austenite is larger than that of the dislocation multiplication in martensite, and then after 6% strain the $\rho_M$ in martensite gradually rises with increasing strain, indicating that the amount of dislocation transported to retained austenite is less than that of the dislocation multiplication in martensite. In addition to DARA effect, these two high carbon Q–P–T steels exhibit TRIP effect. For example, the volume fraction of retained austenite ($V_{RA}$) in Fe–1.5Si gradually decreases from 29.1%, 25.9%, 20.4%, 17.9%, and 15.1% to 10.4% with increasing strain from 0%, 2%, 4%, 6%, and 8% to 12.4%, respectively, which exhibits TRIP effect. TRIP effect gradually occurs during deformation accompanying with the decrease of the $V_{RA}$. The TRIP effect was confirmed by TEM observation. Since retained austenite in Q–P–T steel has more than 0.8 wt.% carbon content, the strain-induced martensite is twin-type martensite. Figure 6 shows TEM micrographs of strain-induced twinning martensite in deformed Q–P–T specimen after fracture [9]. For BCP effect, Fe–2Al Q–P–T steel increases by about 3% elongation comparing with Fe–1.5Si Q–P–T steel, but the uniform elongation of Fe–2Al Q–P–T steel increases by about 13% comparing with Fe–1.5Si Q–P–T steel. What leads to so large difference in their uniform elongations? Unique explanation is that Al increases the stacking fault energy of retained austenite [22], and in turn raises the mechanical stability of retained austenite and avoid the formation of more brittle strain-induced twin-type martensite. Experiment verifies the correctness of the above consideration. For Fe–1.5Si Q–P–T steel, before tensile test the $V_{RA}$ is 29.1%, at fracture strain of 12.4 the $V_{RA}$ is 10.4%. While for Fe–2Al Q–P–T steel, before tensile test the $V_{RA}$ is 25.0%, at the same strain of 12.4 the $V_{RA}$ is 15.6%. At the same strain, the amount of twin-type martensite formed by retained austenite transformation in for Fe–2Al Q–P–T steel is 9.4%, much less than that 18.7% for Fe–1.5Si Q–P–T steel.

Conclusions

Fe–2Al and Fe–1.5Si high carbon Q–P–T martensitic steels were designed, and the effect of 2Al replacing 1.5Si on mechanical properties was studied. Main conclusions are described as follows.

The tensile strength of Fe–2Al Q–P–T martensitic steel is 1.5 GPa, and the total elongation is 29.1% accompanying with the PSE over the third generation advanced high strength of 30 GPa %, while the tensile strength of Fe–1.5Si Q–P–T martensitic steel is 1.950 GPa, and the total elongation only is 12.4% accompanying with PSE less than 30 GPa %.

(1) The lower strength of Fe–2Al Q–P–T steel than Fe–1.5Si Q–P–T steel is mainly attributed to lower elastic modulus of Fe–Al Q–P–T steel, corresponding to lower cohesive energy of Fe atom and Al atom than Fe atom and Si atom.

(2) The higher ductility of Fe–2Al steel than Fe–1.5Si one is attributed to the higher mechanical stability of retained austenite of Fe–2Al steel, which markedly decreases the formation of brittle strain – induced twin-type martensite in Fe–2Al Q–P–T steel.

Disclosure statement

No potential conflict of interest was reported by the authors.

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