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The origin of the brittleness of high aluminum pearlite and the method for improving ductility

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ABSTRACT

High aluminum bearing steels demonstrate poor ductility due to the presence of κ-pearlite in the as hot-rolled condition. The addition of aluminum shifts the equilibrium eutectoid carbon concentration to a higher value. Consequently, the poor ductility of high aluminum pearlite is believed to originate from the brittleness of κ-pearlite due to the higher fraction of brittle κ-carbide. The spheroidization treatment of κ-carbide is proposed and proven for the first time to improve the ductility by reducing the stress concentration at the carbide/ferrite interface and enhancing the cracking resistance of κ-carbide due to its spheroidized morphology.

Low density steels, achieved by addition of light elements such as aluminium, owing to the combined effects of atomic mass reduction and lattice parameter expansion by light lattice replacement [1–4], have the potential to strengthen the competitiveness of steels for automotive applications. The κ-carbides, usually formed in the Fe-Al-C system [5], have the formula of Fe_3AlC and the structure of "*anti*-perovskite" [6]. The κ-carbides form in two patterns: 1) in a lamellar structure of κ-pearlite during eutectoid transformation or 2) at boundaries of κ-pearlite or ferrite as films or particles [7–9]. The κ-pearlite exhibits much higher strength but lower ductility compared with cementite based pearlite (θ-pearlite) [7–9]. It is believed that the low ductility of κ-pearlite usually brings big trouble for cold-rolling the high aluminium-bearing steel sheets [1,5,7].

As other studies were mainly focused on explaining the mechanism of crack initiation and propagation on the basis of morphology of κ-pearlite and κ-carbide in the high aluminium steels [7,9], the authors in this research proposed that the presence of much larger fraction of brittle κ-carbide as one of the two constituents of κ-pearlite is responsible for the higher brittleness of the steel, compared with cementite based pearlite. The reasoning is based on an equilibrium phase diagram analysis as well as a detailed microscopic investigation. Then a spheroidization process for the lamellar κ-pearlite is proposed as a potential solution for improving the ductility of high aluminium bearing steels involving κ-pearlite. To the authors' knowledge, spheroidization annealing, as a typical process for softening lamellar cementite pearlite, has never been attempted on high aluminium steels containing κ-pearlite. It is proved at the end of this paper that κ-pearlite could be spheroidized by an isothermal holding for a reasonable duration below the A_1 temperature [10–12] with surface free energy reduction as the driving force [13], and consequently the ductility of high aluminium steels could be improved for the subsequent cold-rolling.

A high aluminium alloy with the compositions of Fe-0.55C-1Mn-5.0Al in wt.% was designed and manufactured as a 30 kg ingot of 100 mm × 160 mm × 230 mm dimensions using a vacuum furnace. The ingot was reheated to 1200 °C for forging to make 60 mm × 60 mm × 1000 mm slabs. The slabs were reheated to 1200 °C for approximately 2 h and then hot rolled at a temperature between 1100 °C and 900 °C to approximately 4 mm × 150 mm in section dimensions, followed by air cooling. The actual critical temperature A_{c1} of the experimental alloy was measured using a push–rod Formastor–FII high–speed dilatometer with radio frequency induction heating. The hot-rolled slabs were soaked at 700 °C for 32 h in an electric furnace, for subcritical spheroidization annealing. The spheroidization annealed sheets were cold-rolled to approximately 1.4 mm in thickness. "Dog-bone" shape tensile specimens machined according to ASTM 370 standard, with a gauge length of 25 mm, gauge width of 6.3 mm, were applied for the tensile tests which were carried out using a universal tensile test machine at a strain rate of $2 \times 10^{-3}\,s^{-1}$. Two measurements were repeated for each test.

The samples for microscopic observations were prepared by standard methods and then etched with 4% nital for 20 s. High-resolution observations and elements analysis were conducted using the electron probe microanalysis (JEOL JXA-8530 F) operated at 20 kV accelerating voltage. The phase volume fraction was analyzed by using Image J software. The transmission electron microscopy (TEM) work was

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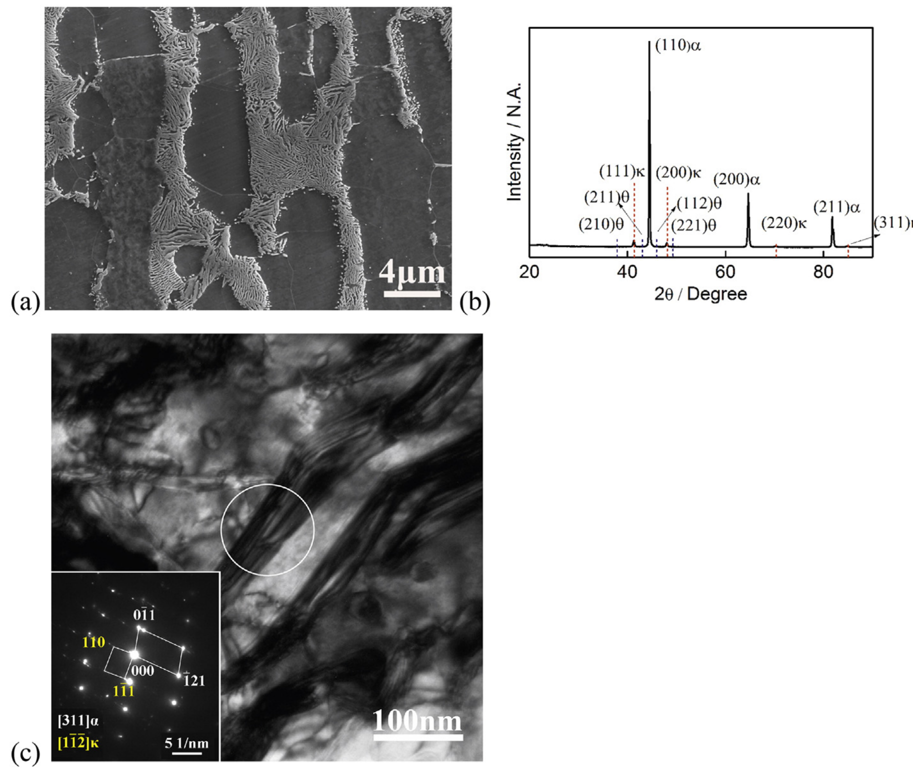

Fig. 1. Microstructure of the as hot-rolled steel. (a) scanning electron micrograph indicating banded structure of ferrite and pearlite; (b) X-ray diffraction pattern showing just κ-carbide and ferrite in the alloy, without cementite; (c) transmission electron micrograph and diffraction pattern along the $[1\bar{1}\bar{2}]$ zone axis showing lamellar κ-carbide in κ-pearlite. (α, κ and θ represent ferrite, κ-carbide and cementite, respectively. The red short dash lines show the reference position of κ-carbide, and the blue short dash lines show the reference position of cementite). (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

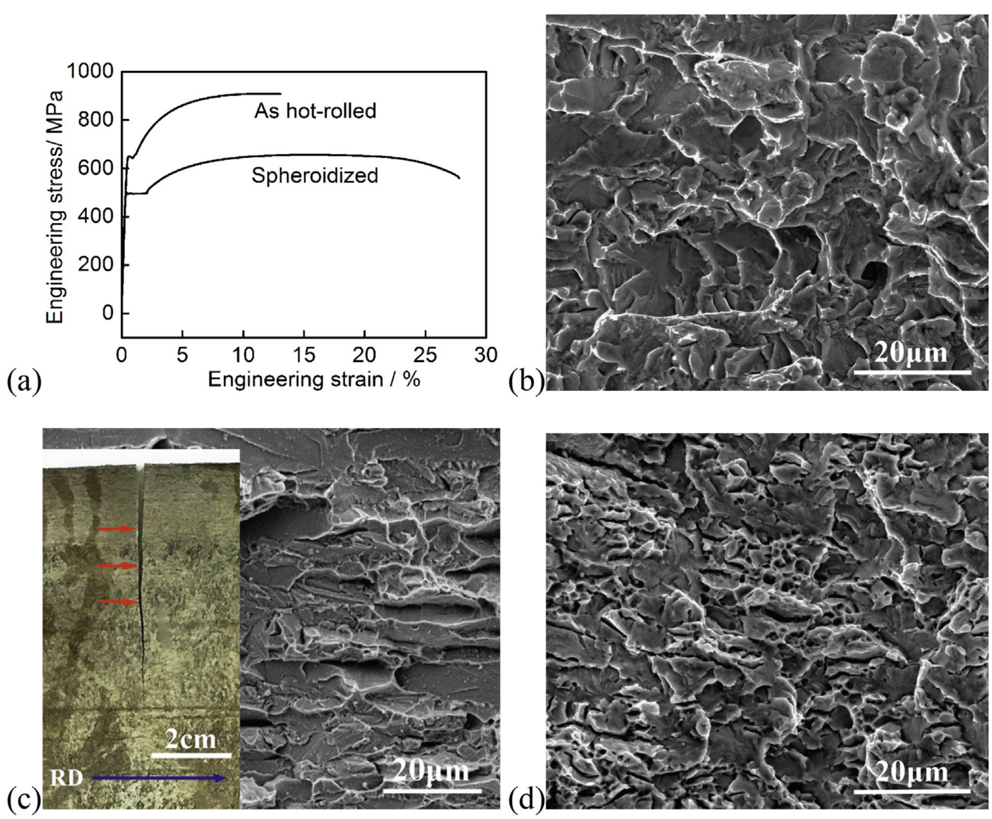

Fig. 2. Mechanical properties by tensile tests and fracture morphology by scanning electronic microscopy of the as hot-rolled and spheroidized steels. (a) engineering stress-strain curves indicating effective ductility improvement by spheroidization; (b) tensile fracture of the as hot-rolled steel showing quasi-cleavage fracture; (c) cold-rolling fracture (RD: rolling direction) of the as hot-rolled steel implying cleavage fracture with obvious transgranular pattern; (d) tensile fracture of spheroidized alloy exhibiting ductile fracture with some spherical dimples.