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Influence of κ-carbide interface structure on the formability of lightweight steels

W.J. Lu [a], R.S. Qin [a,b,*]

[a] Department of Materials, Imperial College London, Exhibition Road, London SW7 2AZ, UK
[b] Department of Engineering & Innovation, The Open University, Milton Keynes MK7 6AA, UK

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ABSTRACT

κ-Carbide (κ) in high aluminium (Al) steels is grown from austenite (γ) via $\gamma \rightarrow \gamma + \kappa$ or $\gamma \rightarrow \alpha + \kappa$ (α represents ferrite), and is a lamellar structure. This work demonstrates that the formability of high Al lightweight steels is affected by the lattice misfit and interface shape between κ and matrix. The cold workability can be improved by either to change the steel chemical constitution or to implement an electro-thermo-mechanical process. For ferrite-matrix-based high Al steel, electric-current promotes the spheroidization and refinement of κ structure and reduces volume fraction of κ phase. This retards the crack nucleation and propagation, and hence improves the materials formability. The observation is caused by a direct effect of electric-current rather than side effects.

1. Introduction

Adding aluminium (Al) to steel reduces its mass density. Lightweight steel is desirable in manufacturing of fuel-economy and emission-reduction transportation components [1]. The challenge in fabrication of high Al steel includes clogging in continuous casting and cracking in cold working. The former is still outstanding due to mold flux problem. Some progress has been made in solving the latter challenge [2–6]. It is noticed that a high Al steel with austenite (γ) matrix possesses better formability than that with ferrite (α) matrix [7]. Mn stabilizes γ phase [8] and affects the plastic properties by influence on the specific stacking fault energy [9]. Fe-Mn-Al-C steel with sufficient high Mn composition is able to retain γ matrix to ambient temperature. Rapid quench helps to retain γ phase when martensitic, bainitic and ferritic transformations can be avoided. κ-Carbide (κ) has a formula $(Fe, Mn)_3AlC$ and is the face-centred-cubic crystal structure. κ in high Al steel is formed by spinodal decomposition or eutectoid decomposition of γ. Spinodal decomposition in high Mn high Al steel leads to $\gamma \rightarrow \gamma + L1_2$ reaction, where $L1_2$ phase forms from an ordering reaction of the solute-enriched low temperature γ phase and subsequently transforms into κ-carbide [10]. Eutectoid decomposition of γ in low Mn (typically with a Mn content up to 10 wt.%) high Al steel leads to $\gamma \rightarrow \alpha + \kappa$. This is ferritic-matrix-based (α) but may contain a fraction of γ. In addition to κ phase formed by the spinodal/eutectoid decomposition, κ precipitates may form at the austenite grain boundaries after long annealing periods [11]. The steels with high Mn low Al and high Mn no Al are found free from κ, with γ matrix and in excellent formability [12–13]. κ-Carbide appears when Al composition becomes significant in steel [8,14–15]. The eutectoid reaction in Fe-Mn-Al-C lightweight steels has been studied systematically [16]. The forming properties of Fe-Mn-Al-C lightweight steels have been reviewed comprehensively [17–18]. A lightweight steel fabricated with designed chemical-thermo-mechanical processing achieves a tensile strength >780 MPa and elongation >30% [7]. It consists of nanoscale κ-carbides and γ-matrix. The nanoscale κ-carbides provide strengthening to the steel [15].

As is well-known, Mn is heavier and more expensive than Al. Low Mn constitution however, leads to the formation of α matrix and poor formability [8]. Other cheap and light γ-stabilizing elements (e.g. carbon or silicon) may cause other engineering problems (e.g. weldability or surface problem) [19]. The aim of this work was to find out how the formability of high Al steel was affected by its microstructure and how the cold work property could be improved. The structure-property analysis of γ-matrix-based and α-matrix-based lightweight steels are presented in Section 2. The electric-current treatment of α-matrix-based lightweight steel is described in Section 3. The processing mechanisms are investigated in Section 4. Section 5 summarized the conclusions.

2. The microstructure-formability relationship

The first objective of present research was to identify the microstructure differences between γ-matrix-based and α-matrix-based high Al steels. The former is better than the latter in term of formability. To

* Corresponding author at: Department of Engineering & Innovation, The Open University, Milton Keynes MK7 6AA, UK.
E-mail address: qin2020@gmail.com (R.S. Qin).

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this purpose two types of ingot with respective chemical compositions of Fe-26Mn-9Al-0.75C and Fe-34Mn-9Al-0.65C (wt.%) were prepared using induction furnace. The microstructure and Vickers hardness of both steels after various annealing processes have been characterized previously [8,14]. This helps to the design of thermomechanical processing conditions for the present research. The as-cast ingots were reheated to 1200 °C for 4 h, followed by hot-rolling reduction from 30 mm to 3 mm thick plates at 1100 °C and then quenched in cold water before annealed at 600 °C and 700 °C respectively for 6 h and then quenched in cold water again.

The microstructure of steel is affected by not only its chemical constitution but also the thermomechanical processing conditions. It is possible to make γ-matrix-based steel with a chemical composition in the range of 12–35 wt.% Mn, 0–12 wt.% Al and 0.5–1.3 wt.% C using adequate processing conditions. A steel containing 26 wt.% Mn is considered high Mn steel. The Mn compositions in both steels are high enough to affect the specific stacking fault energy. Choosing of an annealing temperature of 700 °C for Fe-26Mn-9Al-0.75C steel rather than 600 °C was to maximize its κ-α interface fraction [14]. It is impossible to generate different microstructures using two steels with identical chemical compositions and also the same thermomechanical processing conditions.

The samples cut from each ingot were cold rolled to 70% reduction using an automatic rolling machine in order to examine their cold formability. Microstructure analyses were performed at room temperature using optical microscopy (OM), LEO scanning electron microscopy (SEM) and JEOL 2000FX transmission electron microscopy (TEM). Samples for OM and SEM analyses were grounded, polished and etched in 2 wt.% Nital for 10 s. Samples for TEM analysis were mechanically polished to 30 μm thick, punched to a disk of 3 mm diameter by a copper disk cutter and then jet polished to thin foil specimens in perchloric acid (10%) and acetic acid (90%) mixture under 20 V at 15 °C. The electrical conductivity was measured using Microhmmeter (DO5000 series).

Fe-26Mn-9Al-0.75C samples (hereinafter to be referred as lower Mn samples) are cracked severely after cold rolling. A representative OM image for these samples is shown in Fig. 1(a). Fe-34Mn-9Al-0.65C samples (hereinafter to be referred as higher Mn samples) however, have no crack observed in their cold-rolled samples surface, as is illustrated in Fig. 1(b). Both steels have the same Al but slightly different Mn compositions. The higher Mn steel has better formability than that of the lower Mn steel. SEM analysis reveals the microstructure of the samples, as are presented in Fig. 1(c) for a lower Mn sample and Fig. 1(d) for a higher Mn samples. Both samples are with lamellar structure despite higher Mn sample demonstrates finer and more regular structure than that of lower Mn samples. TEM selected area diffraction pattern (SADP) results are presented in Fig. 1(e) for a lower Mn sample and in Fig. 1(f) for a higher Mn sample. Based on the lattice parameter calculation and crystal structure analysis, the lamellar phase in both steels is found to be κ-carbide with lattice parameter 3.75 Å. The matrix in

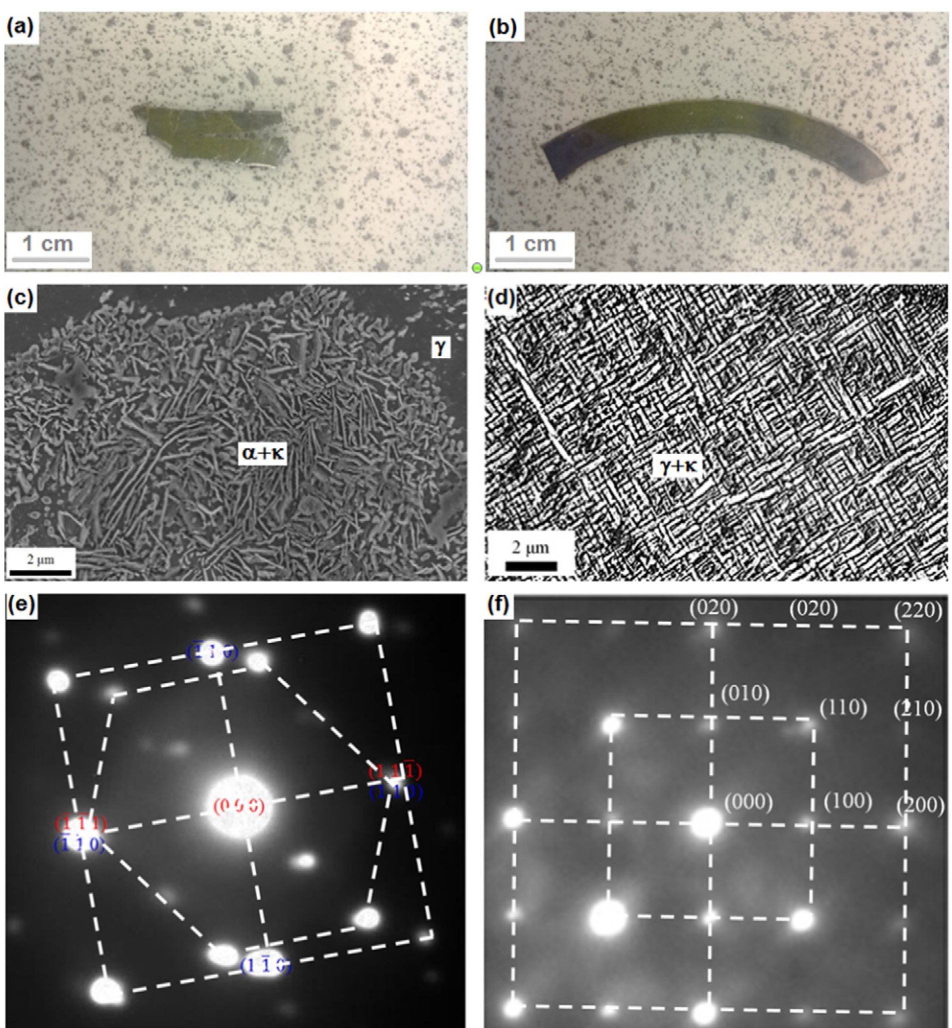

Fig. 1. Optical images for (a) Fe-26Mn-9Al-0.75C (wt.%, lower Mn) and (b) Fe-34Mn-9Al-0.65C (wt.%, higher Mn) steels after cold rolling at room temperature. SEM micrographs for (c) a lower Mn and (d) a higher Mn steels. TEM selected area diffraction patterns (SADP) for (e) a lower Mn sample and (f) a higher Mn sample.