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Revealing the superplastic deformation behaviors of hot rolled 0.10C5Mn2Al steel with an initial martensitic microstructure

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ABSTRACT

A hot rolled 0.10C5Mn2Al steel with an initial martensitic microstructure was tensile deformed in a temperature range of 600–1100 °C. It was found that a maximum elongation to failure of 340% and a relative high strain rate sensitivity of ~0.24–0.36 were obtained at 800 °C under an initial strain rate of 10^{-3}/s. The superplasticity of the hot rolled steel was attributed to the formation of lath-typed ultrafine $\alpha + \gamma$ microduplex structure and the microstructure stabilization enhanced by substitutional element partitioning during superplastic deformation. The present results provide a promising way to develop superplastic steels with non-ideal initial superplastic microstructures.

Since the finding of super-plasticity of metallic materials for decades by Bengough in 1921 [1], Rosenhain in 1920 [2] and Pearson in 1934 [3], many metals and alloys, such as titanium, aluminium, magnesium and their alloys, have been proved to be superplastic materials [4–8]. At present time, thousands of tons of metallic sheet materials have been superplastically formed each year to meet the ever increasing demands of commercial applications. However, apart from cold rolled high alloyed austenitic steels and microduplex stainless steels, nearly all of low carbon steels do not assume superplastic behavior without complex and repeated pre-processing to obtain ultrafine microstructures required by superplastic deformation [9,10].

In our previous study, it was revealed that an excellent superplasticity with elongation to failure larger than 1000% was realized in a cold rolled low carbon medium-Mn steel containing 1–2% aluminium under tensile strain rate 10^{-2}–10^{-3}/s [10]. This excellent superplasticity was attributed to the dynamic austenite reverted transformation and the resulted ultrafine $\alpha + \gamma$ microduplex structure in the 70% cold rolled low carbon medium-Mn steel. Similarly, the superplasticity of a cold rolled FeMnAl steel with very low carbon content was demonstrated with elongation to failure larger than 1000% when it was tensile deformed in the high-temperature range [11]. Both of these two results indicated that superplasticity could be realized in cold rolled medium-Mn steels alloyed with aluminium without complex pre-treatment, which for the first time makes it possible that cold rolled low carbon and low alloy steels could be commercialized for superplastic forming just through conventional fabrication techniques in steel industry. It is well known that austenite reverted transformation and the resulting ultrafine $\alpha + \gamma$ microduplex structure could be developed both in cold rolled medium-Mn steels and in fully martensitic steels [12,13]. Thus the superplasticity could be expected in hot rolled medium-Mn steels with initial martensitic microstructures, which was not reported in literature up to now.

In the present study, the high-temperature deformation behavior of a low carbon medium-Mn steel alloyed with 2% aluminium was examined and the microstructural evolution was examined by using transmission electron microscopy (TEM) and electron backscatter diffraction (EBSD), aiming to explore the possibility of the superplasticity of the aluminium added low carbon medium-Mn steel and its underlined superplastic deformation mechanism.

The nominal chemical composition of 0.10%C, 5%Mn and 2%Al was designed and mould casted by using a laboratory vacuum induction melting furnace with 50 kg capacity. The ingot was forged into plates and finally hot rolled into strips with thickness of 6 mm in the temperature range of 900–1150 °C. Dog-bone shaped specimens with a gauge length of 10 mm and a diameter of 3 mm were machined from the hot rolled strips with tensile direction parallel to the rolling direction. The tensile experiments were carried out at temperatures from 600 °C–1100 °C with an initial strain rate of 10^{-3}/s without any atmosphere protection. Before tensile testing, the specimens were

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heated to a given temperature with a heating rate of ~2 °C/s and holding for about 5 min to get a uniform temperature distribution in the specimens. Interrupted tensile tests with varied initial strain rates of 10^{-4}/s, 10^{-3}/s, 10^{-2}/s and 10^{-1}/s were carried out at 800 °C to determine the true flow stress at various strain rates at different tensile strains of 10%, 100% and 200%, respectively.

The microstructures in the hot rolled sample, in the sample annealed at 800 °C for 6 h, in the cross-head region and the gauge region after superplastic deformation were examined by TEM and EBSD. TEM specimens were first mechanically ground down to ~40 μm thickness, followed by twin-jet polishing in a solution of 5% perchloric acid and 95% alcohol at temperature at about −20 °C. EBSD specimens were mechanical polished down to 1 μm diamond solution and electropolishing in a 10% perchloric acid alcohol solution. The volume fraction of austenite at different tensile deformation temperatures and the melting temperature of the studied steel were calculated by the commercial Thermo-Calc software with the database of TCFE7.

The high-temperature tensile engineering stress-strain curves of 0.10C5Mn2Al are shown in Fig. 1(a). It can be seen that the elongation to failure first increases from 600 °C to 800 °C and then decreases with further increasing of deformation temperature. The maximum elongation of 340% for 0.10C5Mn2Al was obtained at an initial strain rate of 1×10^{-3}/s and deformation temperature of 800 °C. This means that the superplasticity could be realized in the hot rolled steel when it was deformed in the temperature range from 600 °C to 1100 °C. The stress-strain curve measured by the strain rate jump test is plotted in Fig. 1(b). It can be seen that the flow stress changes sharply with the changing of strain rate, indicating a strong strain rate dependence of flow stress. The strain rate sensitivity (m) of stress on strain rate was calculated according to Eq. (1) [4–9] and the results were plotted in Fig. 1(c). It can be seen both deformation strain and deformation strain rate strongly affect the strain rate sensitivity. The strain rate sensitivity at strain rate 10^{-3}/s is about 0.24 at tensile strain of 10%, but increased to 0.32 at strain of 100% and finally increased to 0.36 at strain of 200%, which implies that microstructure is varied with strain and the pre-deformation is beneficial to the superplastic deformation. However, when the strain rate is larger than 10^{-2}/s, increasing the pre-deformation strain decreases the strain rate sensitivity.

$$m = d\ \ln(\sigma)/d\ \ln(\dot{\varepsilon}) \tag{1}$$

where m is the strain rate sensitivity, σ is the flow stress and $\dot{\varepsilon}$ is the applied strain rate.

The TEM micrographs of the hot rolling sample and the sample after static annealing at 800 °C for 6 h are shown in Fig. 2(a) and (b). It can be seen clearly that the microstructure after hot rolling is a typical lath-typed martensitic microstructure with high dislocation density as shown in Fig. 2(a). However, a well-defined martensite and ferrite microduplex structure could be developed after 6 h static annealing and cooling down to room temperature as demonstrated in Fig. 2(b). It is reasonable to assume that during the intercritical temperature holding process, the fully martensitic microstructure is transformed into a lath-typed $\alpha + \gamma$ microduplex structure by the formation of austenitic lathes between martensitic lathes. This partial phase transformation is controlled by the partitioning of Mn and C elements into the austenitic laths and the partitioning Al elements into the ferritic laths [10–14]. The EBSD orientation maps of the grip region without any deformation and of the gauge region after 340% superplastic deformation are revealed in Fig. 2(c) and Fig. 2(d). It can be seen that the lath-typed microduplex structure was still retained in the grip region with reverted austenite lath embedded in between ferrite lathes, as shown in Fig. 2(c). However, an equiaxed grain microstructure with nearly equal phase volume fractions of α and γ phases calculated from EBSD data was developed after 340% superplastic deformation, as revealed in Fig. 2(d). This implies that the initial martensitic lath-typed microstructure was transformed into the equiaxed $\alpha + \gamma$ microduplex structure during the superplastic deformation process.

Generally, an ultrafine granular microstructure with relative good microstructural stability was thought to be one of the prerequisites for the microstructure controlled superplasticity at a deformation temperature higher than 0.5Tm (Tm is the melting temperature, which is about 1465 °C calculated by the Thermo-Calc software). The ultrafine granular microstructure is thought to be benefit to the boundary sliding

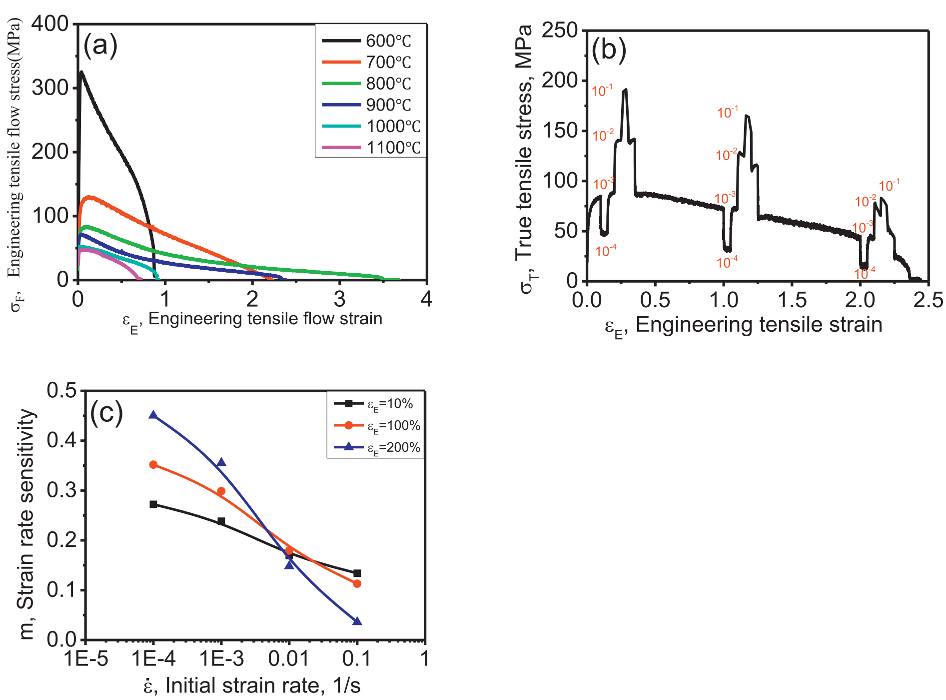

Fig. 1. High temperature tensile deformation of 0.1C5Mn2Al steel (a) engineering tensile stress-strain curves at an initial strain rate of 1×10^{-3}/s and (b) interrupted tensile test by varying initial strain rate from 10^{-4}/s, 10^{-3}/s, 10^{-2}/s to 10^{-1}/s at 800 °C, and (c) calculated results of strain rate sensitivity (m) at different strain rate.