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Tensile behaviour of a nanocrystalline bainitic steel containing 3 wt% silicon

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ABSTRACT

Much recent work has been devoted to characterize the microstructure and mechanical properties bainitic nanostructured steels. The microstructure is developed by isothermal heat treatment at temperatures as low as 125–350 °C and adapted steel grades typically contain high carbon contents to achieve sufficient depletion of the B_S–M_S temperature range, and above 1.5 Si wt.% to suppress carbide formation during isothermal holding. On the latter, most of the published literature agrees on a limit of around 1.2–1.5 wt.% to suppress cementite in high carbon steels. For this reason perhaps, additions of Si significantly above this limit have not been investigated systematically in the context of nanostructured bainitic steels. The present work is concerned with the effect of up to ~3 Si wt.% in a steel grade adapted to low temperature bainitizing. Tensile properties as compared to similar grades, though with lower Si contents, exhibited unrivalled combinations of strength and ductility, with above 21% total elongation for a UTS above 2 GPa. An attempt is made to explain the mechanical properties of this microstructure in terms of some of its most relevant and unique morphological and microstructural features.

1. Introduction

Grain refinement is a well-known and widely used method to achieve combinations of ductility and high strength in metallic materials. In heat-treatable material, this refinement can be achieved not only through complex thermomechanical processing but also through simple heat-treatments schedules [1]. An example of the latter is low temperature bainitizing, whereby one achieves a microstructure of bainitic ferrite plates a few tens of nanometres thick [2–4]. Indeed the degree of refinement achievable during low temperature bainitizing is such that the topic has received unprecedented attention in recent years [5–10].

It is established [11] that the bainitic ferrite plate thickness depends primarily on two factors, first the strength of the austenite at the transformation temperature and second, the chemical free energy change accompanying transformation. Thus, strong austenite or a large driving force results in finer plates, the former because there is a large resistance to interface motion and the latter because an increased nucleation rate leads to microstructural refinement. Both, austenite strength and driving force, increase as the transformation temperature decreases. In previous experiences [2–4] the design of low temperature bainitic steels was based on the decrease of the B_S–M_S temperature range by chemical composition control and the increase in the transformation driving force, to achieve adequate transformation times consistent with the requirements of industrial manufacturing, by Co and/or Al additions. However, due to the cost of raw material (Co) and the incompatibility of high Al additions with cleanliness requirements of ultra high strength steels, both alloys are unlikely to find industrial use.

It is known that Si is a strong austenite solid solution hardener [12,13], therefore inclusion of this element in much higher quantities, in principle, is a way to ensure that the bainite growth results in even finer plates. Silicon is originally added to the chemical composition of this type of nanostructured steels in quantities close to 1.5 wt.% in order to retard and to some extent to avoid cementite precipitation from austenite during bainite reaction. In this work, silicon additions are increased up to 3 wt.% to get an extra strengthening in the austenite prior to bainite reaction. According to Ref. [12], an increase of 1.5 wt.% of Si implies an increase of 7% in the YS of austenite. Both, Co and Al, accelerate bainite transformation, therefore their absence in the chemical composition was compensated by reducing the quantities of Mn and Cr present [14]. As in the other cases, the design was based on phase transformation theory [15] and other well-known metallurgical facts [16–18].

Results will be presented that demonstrate unprecedented levels of ductility at strength above 2 GPa.

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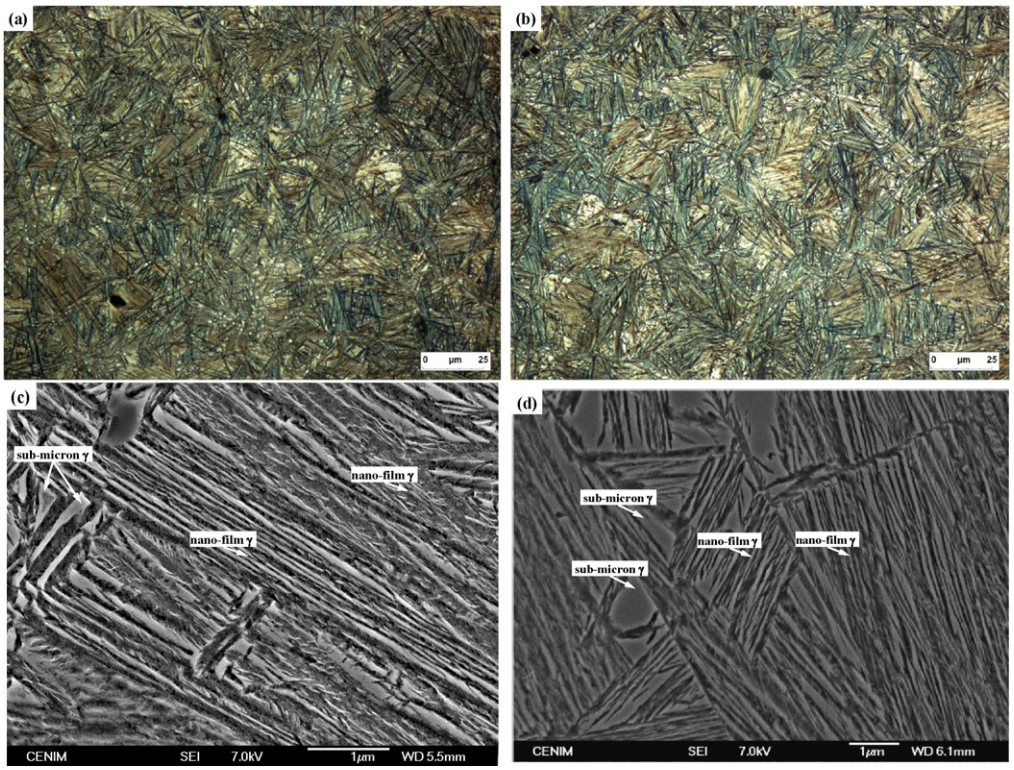

Fig. 1. Examples at different magnifications of the microstructures obtained by isothermal transformation at, 220 °C (a) and (c) and 250 °C and (b) and (d).

Table 1
Chemical composition, all in wt.%.

	C	Si	Mn	Cr	Cu	Ni	P	S
1CSi	0.98	2.90	0.77	0.45	0.21	0.16	0.016	0.014

2. Material and experimental procedure

The designed alloy has a chemical composition, Table 1, with enough C to ensure low transformation temperatures and 3 wt.% Si to ensure solid solution strengthening as well as to avoid cementite precipitation. As Co and Al, no longer present in the chemical composition, have an accelerating effect on bainitic transformation, Cr and Mn were kept as low as possible to maintain the transformation times with the range of the Co and Co + Al alloys, and, at the same time, to ensure sufficient hardenability to avoid transformation during cooling from the austenitisation temperature to the bainite transformation temperature. For this purposes 0.45 and 0.77 wt.% of Cr and Mn were added to the chemical composition, an important reduction if compared with 1 and 2 wt.% of Cr and Mn in the Co and Co + Al alloys [4].

The alloy was produced as a 40 kg ingots and forged to a final diameter of about 50 mm. Cuts of about $15 \times 15 \times 100\,mm^3$ (tensile tests pre-forms) were austenitised at 950 °C for 60 min, then transferred to a salt bath at the required bainitizing temperature. Bainitizing was thus carried out at 250 °C and 220 °C for 16 and 22 h respectively, to achieve a completely bainitic microstructure. Those reactions times compare well with those reported in Ref. [4], validating the design process in terms of obtaining similar transformation kinetics as those of the benchmark alloys but without the use of expensive alloys elements as Co and Al.

To reveal the microstructure, metallographic samples were cut, ground and polished following the standard procedures. A 2% Nital etching solution was used to reveal bainitic microstructure. Scanning electron microscopy observation was carried out on a JEOL J8M-6500 field emission gun scanning electron microscope (SEM-FEG) operating at 10 kV. Before etching, samples were polished using colloidal silica suspension. High magnification SEM-FEG micrographs were used to determine the distribution and size of the different retained austenite morphologies and also the bainitic ferrite plate thickness t, by measuring the mean lineal intercept $\bar{L}_T = \pi t/2$ in a direction normal to the plate length [4,19].

TEM specimens were sliced from 3-mm-diameter rods of the heat-treated material, mechanically thinned to 0.06 mm, and then twin-jet electropolished to perforation using a mixture of 5% perchloric acid, 25% glycerol and 70% ethanol at 10 °C at 45 V. The samples were examined on a TEM JEOL 2010 transmission electron microscope operated at 200 keV.

Quantitative X-ray diffraction analysis was used to determine the fraction of retained austenite (V_γ) and its carbon content (C_γ). For this purpose, samples were machined, ground and finally polished using colloidal silica suspension. They were then step-scanned in a SIEMENS D5000 X-ray diffractometer using unfiltered Co Kα radiation. The scanning speed (2θ) was less than 0.3°/min. The machine was operated at 40 kV and 30 mA. The volume fraction

Table 2
Quantitative data on microstructure. V_i and C_i stands for the fraction and C content of the phase i, where i could be, α_b = bainitic ferrite and γ = austenite. t stands for the plate thickness of bainitic ferrite, γ_{film} and γ_{block} stands for the thickness of both morphologies of retained austenite size, a_γ is the austenite lattice parameter.

Samples	a_γ (nm)	V_{α_b}	V_γ	C_{α_b} (wt.%)	C_γ (wt.%)	t (nm)	γ_{film} (nm)	γ_{block} (nm)
220 °C	0.362	0.64 ± 0.02	0.36 ± 0.01	0.08 ± 0.06	1.22 ± 0.06	28 ± 2	34 ± 1	630 ± 19
250 °C	0.363	0.66 ± 0.02	0.34 ± 0.01	0.05 ± 0.06	1.47 ± 0.06	28 ± 1	37 ± 2	870 ± 29