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Thermo-mechanically processed dual matrix ductile iron produced by continuous cooling transformation

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ABSTRACT

Ductile irons with dual matrix structure were attained by controlled cooling in the austenite + ferrite region, after austenitization. Afterwards, the ductile irons were quenched to either 375 °C, so as to transform the austenite into ausferrite or to room temperature to transform austenite to martensite. Fully ausferritic and fully martensitic matrices were also produced by direct quenching from the austenite region. Furthermore, three different deformations with true-strain values of 0, 0.3 and 0.5 have been applied in the austenite region. The structures were produced in three ductile irons with aluminum contents of 0.31 wt.%, 0.96 wt.% and 1.7 wt.%. The different microstructures were produced in a thermo-mechanical simulator equipped with a dilatometry system. Dilatometry was used to monitor the structure development throughout the thermo-mechanical processes. The microstructural changes and hardness property exhibited by the produced structures were investigated. Increasing the aluminum widened the intercritical region making the control of the pro-eutectoid ferrite volume fraction easier. The quenching temperature in the intercritical region was shifted to higher value by both increasing the aluminum content and increasing the deformation. This shift resulted in a higher carbon level saturating the intercritical austenite which resulted in fundamental effects on the subsequent transformation kinetics.

1. Introduction

Ductile iron (DI) with dual matrix structure (DMS) is a new class of materials in which the matrix structure consists of a soft phase, which is the ferrite and a hard phase which is either martensite or ausferrite (acicular ferrite and high carbon austenite). The ductile iron with DMS gained interest due to the improved combination between strength and ductility.

According to Kobayashi and Yamada (1996), the combined strength-elongation of DMS with ausferrite as hard phase is better than the conventional austempered ductile iron (ADI). Wade et al. (1985) indicated that the mechanical properties of DMS with ausferrite are improved, more than those of the same ductile iron with martensite in DMS. Ferrite in the DMS is obtained by either an incomplete austenitization step (by isothermal holding at temperatures within the intercritical interval) (Kobayashi el al., 1996; Wade el al., 1985) or by controlled cooling of the austenite in the intercritical region (Soliman el al., 2011).

On the other hand, hot deformation as a method of strength improvement of ductile iron recently attracts research and industrial interest (Qi et al., 2009; Zhao et al., 2004). Hot deformation refines the as-cast structures, closes up the internal shrinkage cavities and gas porosity, and reduces the segregations of alloying elements. Additionally, it increases the dimensional accuracy and improves surface finish of the products which would finally reduce the manufacturing costs. Forged DI products have been promoted as replacements of some types of forging steels. An attempt was made by Shi et al. (1994) to describe the anisotropy of tensile strength in terms of the deformation of graphite nodules of deformed DI. Furthermore, Bonora and Ruggiero (2005) established deformation model of nodular graphite in DI to observe the relationship between the local matrix deformation and the inclusion deformation.

The current study aims at combining both thermo-mechanical processing and introducing the pro-eutectoid ferrite to the DI by controlled cooling in the intercritical region to produce thermo-mechanically processed DMS-DI in a single process chain. In the current proposed process, the dissolution of carbides and refining of the matrix structure take place during the preheating and the deformation process, respectively. Whereas the DMS is formed during the subsequent controlled thermal cycle. For the proposed

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Table 1
Composition of the ductile irons (wt.%).

Alloy	C	Si	Mn	Al	Mg	S	P
A1	3.68	2.48	0.29	0.31	0.040	0.011	0.029
A2	3.57	2.59	0.33	0.96	0.046	0.011	0.025
A3	3.70	2.66	0.31	1.74	0.040	0.017	0.026

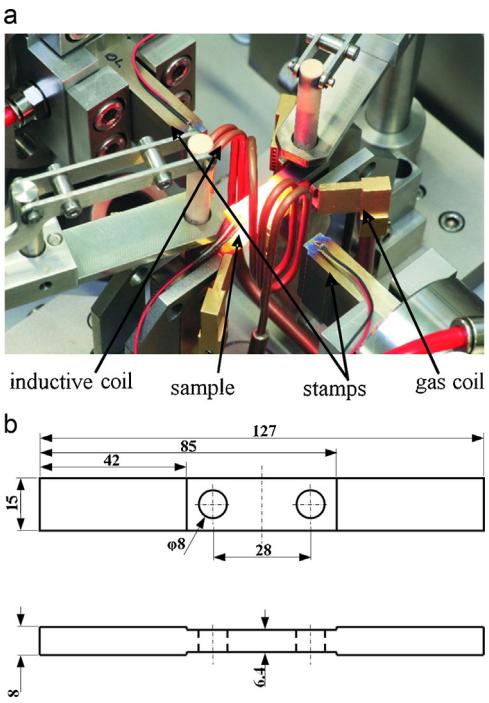

Fig. 1. (a) Experimental set-up for flat compression at Bähr TTS820 thermo-mechanical simulator and (b) with the sample geometry for flat compression test.

process a wide range of intercritical region would ease controlling the pro-eutectoid volume fraction. For this purpose aluminum is added because of its effect of widening the intercritical region as reported by Soliman and Palkowski (2008). Thus the investigation aims at:

1- Studying the effect of aluminum content and the applied deformation on the kinetics of pro-eutectoid ferrite, martensite and ausferrite formation in the thermo-mechanically processed ductile iron.

2- Selecting thermal parameters depending on this study.

3- Investigating the effect of both the aluminium content, the applied deformation and introducing ferrite to the matrix on the hardness properties.

2. Experimental procedure

Three ductile irons with different aluminium contents were investigated. The chemical compositions of the alloys are given in Table 1. Melting was performed in an induction furnace. Ductile iron treatment was performed in an open ladle using the sandwich method. The base irons were treated with a 9.5 wt.% MgFeSi alloy for spheroidisation followed by post-inoculation with 75 wt.% FeSi. The base iron was tapped at about 1520 °C onto the treatment alloys. The melt was then cast into Y shape sand moulds. The dimensions of the Y-block were 15 × 180 × 250 mm with the 250 mm to the direction of the Y form.

For thermo-mechanical processing and dilatometric study, a Baehr TTS820 thermo-mechanical simulator, see Fig. 1a, is used. Flat compression samples with dimensions shown in Fig. 1b are machined from ductile iron blocks. With the deformation simulator, the dimension variations of the specimens during the thermal-deformation cycle are transmitted via a laser system with 1 μm accuracy. The thermal cycles are performed under vacuum conditions with 0.005 Pa by inductive heating using a high frequency (HF) generator. Helium is used as cooling gas.

Table 2
Measured critical temperatures (A_{r1} and A_{r3}) and estimated intercritical temperatures (T_Q) corresponding to 10% of ferrite in the matrix.

φ_t	0			0.3			0.5		
Alloy	A1	A2	A3	A1	A2	A3	A1	A2	A3
A_{r1} (°C)	671	680	691	668	682	693	670	678	690
T_Q (°C)	698	730	750	707	750	760	710	748	761
A_{r3} (°C)	724	757	779	732	770	789	728	775	788

The specimens are subjected to two schedules, namely "Schedule I" and "Schedule II" designated in Fig. 2. In both schedules, specimens were heated up to 960 °C and subjected to the deformation steps given. Moreover, three total true strains (ϕ_t) of 0, 0.3 and 0.5 are applied. In "Schedule I" the specimens were quenched after the last deformation either to room temperature (RT) or to an austempering temperature of 375 °C to obtain the structures "M" and "Af", respectively. In "Schedule II" - after the deformation steps - the specimens were cooled to the two phase region ferrite+austenite with a rate of 1 Ks^{-1} to form a certain ferrite amount before quenching at T_Q.

To investigate microstructural constituents, the specimens were prepared by mechanical grinding followed by polishing up to 1 μm with diamond paste. The microstructures were examined with a light optical microscope (LOM) after etching with 2% nital. The samples for scanning electron microscope (SEM) investigations were deep nital etched. Macro-hardness testing was carried out at RT using Vickers hardness according to ASTM E384 (2011) applying a load of 20 kg (HV20). Average values of the hardness were obtained based on 5 measurements.

3. Results and discussion

3.1. As cast structure

Fig. 3 shows the as-cast (AC) microstructures. The nodule characteristics in the AC condition are given in Table 1. A remarkable decrease in the nodule count and increase in the nodule size is observed when increasing the aluminium content. A clear deterioration in the nodularity of A3 is also observed. Electron probe microanalysis (EPMA) results of Zeraati et al. (2010) show that the segregation and accumulation of aluminium have their maximum values around the graphite nodules. This accumulation is non-homogeneous and irregular, which leads to the non-uniform diffusion of carbon and affects the shape of graphite nodules.

3.2. The intercritical region

The dilatometric curves shown in Fig. 4 are records of the length change during cooling with 1 Ks^{-1} of the three alloys investigated. Increasing the aluminum content has a remarkable effect on the intercritical region; its effect on A_{r1} and A_{r3} is shown in Table 2. The increase in aluminum resulted in increased A_{r1} and A_{r3} values. The results indicate that the aluminum exerts a stronger influence on the transformation start temperature A_{r3} when the ferrite phase starts to form from the austenite than on the transformation finish temperature A_{r1}, when the complete austenite phase is exhausted.

Applying $\phi_t = 0.3$ on the material resulted in a clear increase of A_{r3} for all alloys. The applied deformation caused refinement of austenite grains. The latter effect resulted in creating more grain boundary areas for the nucleation of ferrite which motivated its formation (Palkowski et al., 2007). This effect seems to reach a