THE INFLUENCE OF MICROSTRUCTURE ON THE MECHANICAL PROPERTIES AND FRACTURE TOUGHNESS OF ADI

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The bainitic microstructure produced by austempering nodular iron has been examined by light and transmission electron microscopy to explain the influence of austempering temperature and time on the mechanical properties and fracture toughness. The fracture modes were also examined and correlated with the microstructure and mechanical behaviour. In austempered ductile iron high elongation values are found to correspond with an optimum austenite content. However, this optimum content can lead to reduced fracture toughness as a result of carbide precipitation at the ferrite-austenite inteface.

INTRODUCTION

Austempered ductile iron (ADI) exibits a remarkable combination of strenght and ductility or stenght and wear resistance and has been used for variaty of components in the automotive railroad and heavy engineering industries (1-4). The mechanical properties can be varied over a wide range by differences in the type and amount of microconstituents arising from variations in composition and heat treatment variables.

The literature reveals few references to combined application of light (LM), transmission electron (TEM) and scanning electron (SEM) studies on austempered nodular cast iron and correlated the influence of the microstructure on the mechanical properties and fracture toughness, (5-7).

In conventional austempered ductile iron austenite is retained in the final structure because Si stongly retards the precipitation of carbides. It has been also reported that Al is a strong graphitizer and delays the

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formation of carbides during the bainitic reaction. Al stabilizes the retained austenite at longer austempering time in comparison to Si.

The present paper reports the results of the microstructure of ADI using LM, TEM and SEM and the observation correlated with mechanical properties and fracture toughness.

EXPERIMENTAL PROCEDURE

Tensile, impact and TPB specimens were produced from the keel block of an unalloyed Si SG cast iron containing 3.8% C, 2.9% Si, 0.008% S, 0.032% Mg and an unalloyed Al SG containing 3.27% C, 2.3% Al, less than 0.01% S, 0.052% Mg. Samples were austenitised at temperatures of 900 and 950^{o}C for 1.5 h and austempered at temperatures of 250, 300, 350 and 400^{o}C for 1, 2, 3 and 5 h in the salt bath. Samples were taken from the tensile and impact test specimens far from fractured area and prepared by standard metallographic techniques and examined by a Leitz-light microscope. In order to identify the unreacted and reacted austenite the heat-tinting technique was used (8).

Thin-foil preparation for TEM observation was carried out by i.) mechanical polishing of 3 mm discs to 0.15 mm thick, ii.) dimpling of the central zone of the specimen to a thickness 25 μm and ion milling with an argon beam with potencial of 5-6 kV at an initial angle of 30^{o}(8-10h) and an additional 0.5h at an angle of 16^{o}.

The thin foils were examined in a JEOL 4.000 FX transmission electron microscope operated at 300 kV, while for the fracture modes an JEOL JSM 35 scanning electron microscope operation at 25 kV was used.

To determinate volume fraction of retained austenite an x-ray diffraction analysis was carried out with a Mo-Kα target radiation at 40 kV and 30 mA.

Mechanical tests were performed on standard specimens using an Instron 20 kN and Charpy machine.

RESULTS AND DISCUSSION

The graphite nodules of all tested specimens Si SG and Al SG were found to be uniform in size and distribution. Spheroidisation was evident (90%) with the average size of nodules 35 μm for Si SG and 17 μm for Al SG and the average density of nodules 150/mm² and 300/mm² respectively - (Fig. 1 a,b). The microstructure of the metal matrix after isothermal transformation is strongly dependent upon the austempering temperature and time. The specimens austenitized at 900 and 950^{o}C and austempered at 250, 300 and 350^{o}C in LM show typical lower bainitic structure with an acicular appearance of bainitic ferrite and retained austenite. As the austempering temperature increases to 400^{0}C the feathery bainitic ferrite appears and also the ferrite/austenite spacing increases - (Fig. 2 a,b).There is no essential difference between the morphology of bainitic ferrite and retained austenite in austempered Si SG and Al SG. The structure of specimens austempered for 1h and 5h respectively can not also show any difference.

A typical austempered ductile iron TEM microstructure consisting of a stable highly enriched retained austenite with carbide-free bainitic ferrite was produced in the Si SG specimens austempered at 350^{0}C for 1h and in Al SG at 400^{0}C for 1h - (Fig. 3 a-d). For other heat treatment conditions three different types of transition carbide were identified- η-carbide, ε-carbide and χ-carbide. The η-carbide was formed at 350^{0}C after 2h only in Si SG specimens by precipitation from bainitic ferrite supersaturated with carbon - (Fig. 4a). By contrast, ε-carbide was associated with austempering at 300^{0}C for 2h in Si SG - (Fig. 4b), and at 300^{0}C for 5h in Al SG. There is close similarity between the structure of ε and η-carbide. The main difference is in the arrangement of the carbon atoms. In the η-carbides the carbon atoms produce a sublattice by regularly filling one half of the octahedral sites between the iron atoms and therefore produce superlattice reflections not present in the diffraction pattern from ε-carbide. The χ-carbide produced by decomposition of the enriched austenite, was observed at the ferrite-austenite boundary in that had been austempered at 400^{0}C for 5h - (Fig. 4c,d). Same phenomena was observed in Si SG specimens austempered at 350^{0}C for 3h.

The UTS and elongation values as function of treatment time for various temperatures can be summarised as follows. Isotermal transformation at 250^{0}C produces materials having high mechanical strenght ($\sim$1600 MPa) and low ductility 1%. In this heat treatment Fe_3C carbide was identified within the bainitic ferrite and retained austenite. This is not typical ADI microstructure and is similar to the lower bainite microstructure found in steels.

With increases temperature from 300 to 400^{0}C transformation leads to more balance in properties (1500, 1200, 1150 MPa strength and 3%, 6% and 12% elongation). The impact toughness also increases from 30 to 110 J.

It was assumied (6) that maximum ductility and impact toughness was correlated with the maximum in the amount of austenite. This is true for transformation for up to 2h in Si SG and up to 5h in Al SG because the precipitation of transition carbide within the ferrite does not lead to any confinement of the plastic zone at the tip of any possible microcrack and microslip would be continuous across ferrite/austenite interface. In that case the fractured surface shows a ductile failure mode. Extended austempering (3h) at 350^{0}C for Si SG of (5h) at 400^{0}C for Al SG lead to some decompostion of the enriched austenite to form χ-carbide at the ferrite/austenite interface. In this case amount of retained austenite is almost the same - (Fig. 5a). This form of microstructure results in a lowering of elongation, imact toughness, and fracture toughness. In Al SG impact toughness drops from 110 to 45 J, while in Si SG fracture toughness K_{IC} drops form 80 to 41 MPa $m^{-3/2}$. In addition fracture mode changes from ductile to brittle - (Fig. 5b,c).

CONCLUSION

In Si SG and new Al SG bainitic microstructure were produce by austempering at 250 to 400°C. Typical ADI microstructure consisting of carbide-free bainitic ferrite and highly enriched retained austenite was produced in the specimens austempered at 350°C for 1h and 400°C for 1h respectively. From three identified carbides, η-carbide, ε-carbide, and χ-carbide, χ-carbide is the most undesirable phase. Apart from that Al SG appears to be very promissing new material.

REFERENCES

(1) J. Dodd, R.B. Dunlach and J.A. Lincoln, Casting Engineering and Fondry World, 1984, 16(1), 15.

(2) R.A. Harding, Materials and Design, 1985, 4, 177.

(3) I.G.H. Hughes, Foundry Trade Journal, 1985, 160, 277.

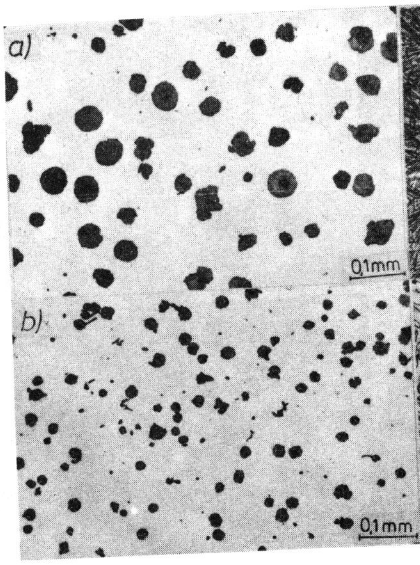
(4) T. Shiokawa, 1978, 59th Japan Ductile Iron Association Conference.

(5) V. Franetović, A.K. Sachdev and E.F. Ryntz, Metallography, 1987, 20, 15-36.

(6) J. Aranzabal, I. Gutierrez, J. M. Rodriguez-Ibabe, and J.J. Urcola, Materials Science and Technology, 1992, vol. 8, 263-273.

(7) L. Šidjanin and R.E. Smallman, Materials Science and Technology in press.

(8) B.V. Covacs, Sr., Modern Casting I, 1987, 35.



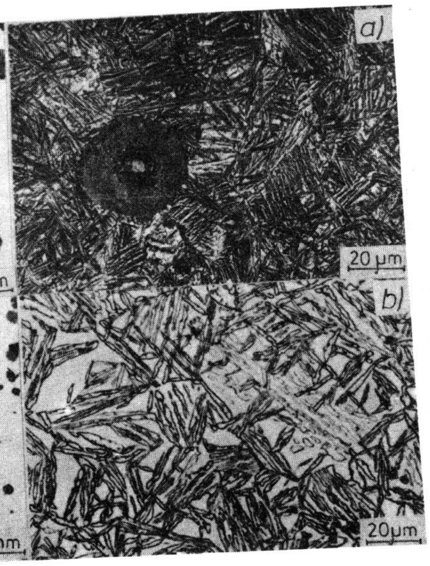


Fig. 1 Grafite nodules
a) Si SG; b) Al SG

Fig. 2 Metal matrix at
a) $350^{o}C/1h$; b) $400^{o}C/5h$

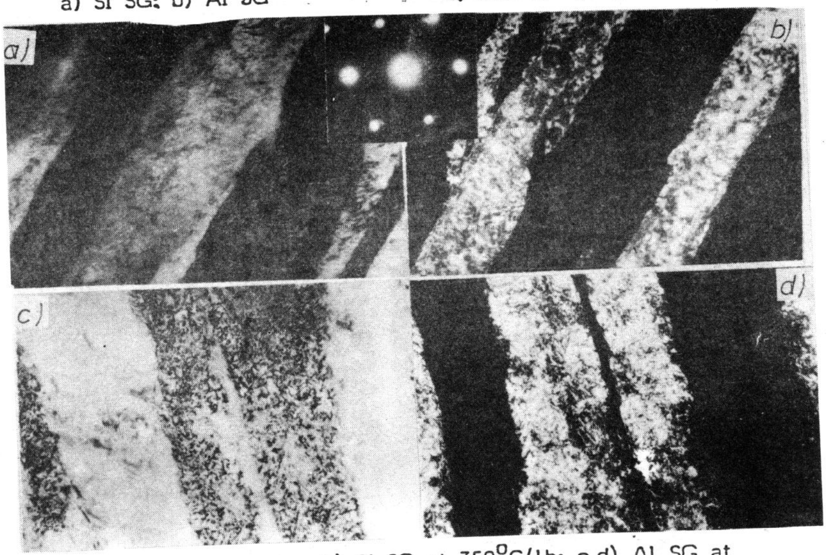


Fig. 3 TEM microstructure: a,b) Si SG at $350^{o}C/1h$; c,d) Al SG at $400^{o}C/1h$

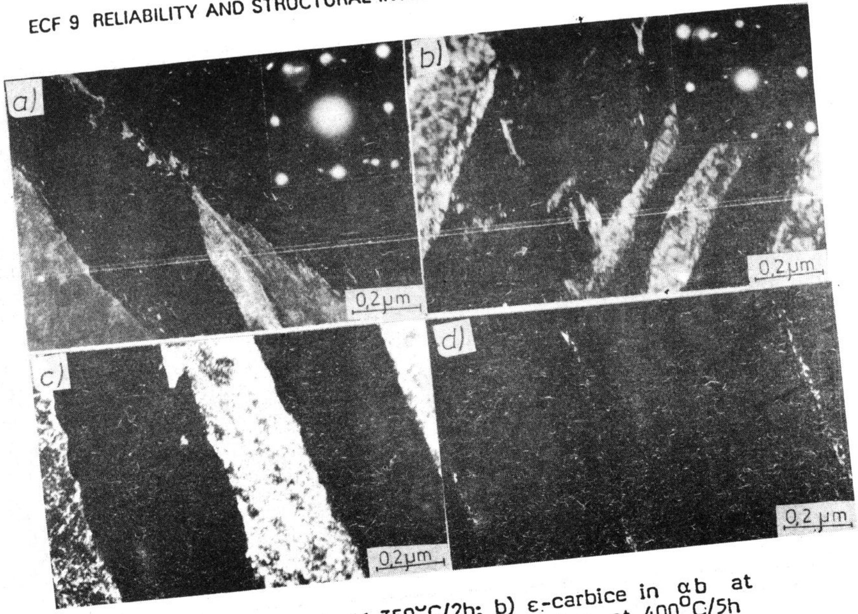


Fig. 4 a) η-carbide in αb at 350°C/2h; b) ε-carbice in αb at 300°C/2h; c,d) χ-carbide in αb/γ interface at 400°C/5h

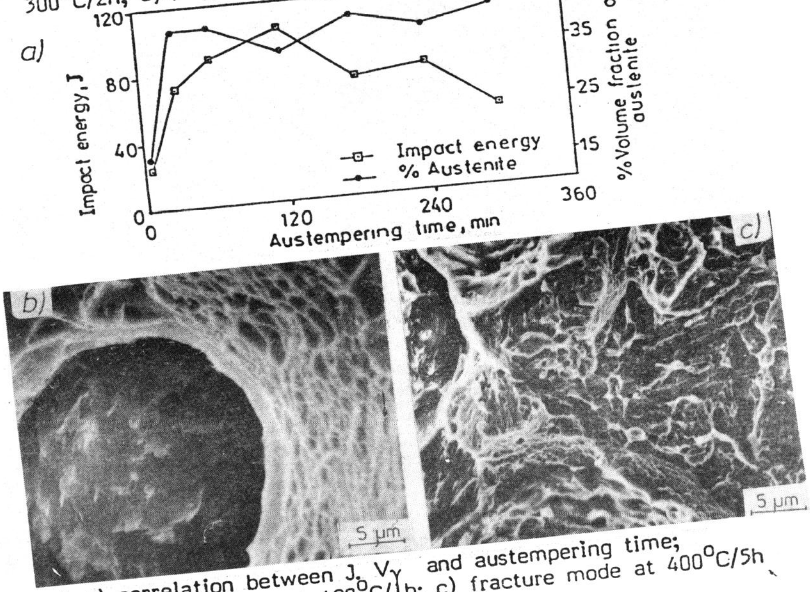


Fig. 5 a) correlation between J, V_{γ} and austempering time; b) fracture mode at 400°C/1h; c) fracture mode at 400°C/5h