JSCS@tmf.bg.ac.yu • www.shd.org.yu/JSCS

J. Serb. Chem. Soc. 73 (1) 113–119 (2008) UDC 539.42:669.141.24:669.112.227:539.3/.5:669
JSCS–3692 *Original scientific paper*

Influence of cooling conditions and amount of retained austenite on the fracture of austempered ductile iron

VYACHESLAV GORYANY*, ECKART HOFMANN and PAUL JOSEF MAUK

Institute of Applied Materials Technology (IAM), University of Duisburg-Essen, D-47048 Duisburg, Germany

(Received 27 December 2005)

Abstract: SEM Analysis of fracture surfaces from tensile test specimens of thick-walled, austempered ductile irons (diameter 160 mm) shows different fracture behavior depending on the austenite retained in the matrix. The results show ductile fractures only in areas containing retained austenite sections. In section areas without or with a very low content of retained austenite, only brittle fracture without any plastic deformation occurs. The content of retained austenite determines the amount of ductile fracture in the microstructure.

Keywords: austempered ductile iron; retained austenite; fracture mechanism.

INTRODUCTION

The microstructure of as-cast ductile iron has a considerable impact on the transformation process during subsequent heat treatment. The formation of the initial microstructure can be regulated *via* the chemical composition (amounts of ferrite/perlite) and inoculation (size and distribution of the graphite nodules).

Investigations[1–4] have shown that the number of graphite nodules (per mm^2) and other graphite-nodule parameters[2] have an important influence on the transformation kinetics and the mechanical properties of austempered ductile iron (ADI). It has become apparent that increasing the number of nodules has a mainly positive influence on the properties of the material because the undesired segregation at the grain boundaries is diminished. Another advantage of a larger number of graphite nodules is that the retention time during austenitizing is reduced. The diffusion paths for carbon are shorter, which means that higher carbon contents in the austenite can be attained in a shorter time and the process window is widened.

The austenitizing time can be shortened without any negative impact on the transformation and the carbon content in the austenite. Another positive effect of a larger number of nodules is that the austenitic–ferritic microstructure (ausfer-

* Corresponding author. E-mail: goryany@ihg.uni-duisburg.de or drgoryany@aol.com
doi: 10.2298/JSC0801113G

rite) becomes finer and more homogeneous. This in turn results in a greater strength and larger elongation after thermal treatment.[3–5]

It was found that austenitic–ferritic cast iron with nodular graphite contains 15 to 40 vol. % of stabilized austenite after thermal treatment[6,7] (sometimes more than 50 %[8]). When the transformation times are too short, the carbon content in the austenite is too low so that the austenite is unstable and may transform into martensite at room temperature and/or when exposed to stress. When subjected to higher temperatures, part of the austenite which had not been evolved in the transformation reaction may transform into martensite and/or into a tempered microstructure. This results in poorer material properties and changes of dimensions. The stability of austenite has been investigated in different ways by many authors[9–27] using different grades of alloyed iron (copper–molybdenum, copper–molybdenum–nickel). In these studies, a partial decomposition of austenite and dimensional changes were observed.[28] In these processes, the decisive influencing factor was not the content of alloying metal, but the thermal treatment.

The results obtained indicate that the content of stabilized austenite in the microstructure has a decisive impact on the mechanical properties.[11,12,16,28–31] Responding to a growing amount of austenite, the ductility, fracture strain and the notch impact energy were found to increase, with the maximum being attained at ≈ 25 % austenite.

The goal and objective of these present investigations was to study the influence of retained austenite on the fracture mechanism of ADI.

EXPERIMENTAL

The test specimen to measure and record the time–temperature curves was a 170 mm high truncated cone with a center diameter of 160 mm into which thermocouples were integrated.

The chemical compositions of the investigated specimens as well as the associated inoculating agents are summarized in Table I.

TABLE I. Chemical composition of the employed test material and inoculating agents

Experiment No.	Inoculating agent	Chemical composition, wt. %									
		C	Si	Mn	P	S	Cr	Ni	Mo	Cu	Ti
1, 2	Reseed®	3.62	2.38	0.28	0.024	0.006	0.036	2.01	0.39	0.99	0.01
3, 4	Ultraseed®	3.68	2.26	0.27	0.025	0.006	0.031	1.97	0.42	0.96	0.01

Reseed® is a strong inoculant based on a ferrosilicon alloy with 75 wt. % Si (FeSi 75) also containing calcium and rare earths. The use of this inoculating agent allowed a large number of small and uniformly distributed graphite nodules to be achieved.

With regards to its basic composition, the inoculant Ultraseed® is identical with Reseed®. However, small amounts of sulfur and oxygen are additionally added to this alloy. In this way, a good inoculation effect with medium-sized nodules and lower wall-thickness sensitivity are supposed to be obtained, even in melts with a low content of oxygen and sulfur. The tendency of shrinkage is reduced.

All specimens were austenitized for a period of 90 min at 900 °C and then quenched using different cooling media.

The specimens for test 1 and 3 were cooled with water (5 l min^{-1} at a pressure of 3.6 bar) to 350 °C at the reference thermocouple level 2. The cooling rate was 1.85 K s^{-1}. In the next step, they were subjected to age-hardening in a furnace for 6 h at 350 °C.

The specimens for tests 2 and 4 were cooled by means of a mixture of water and air (water < 1 l min^{-1} at a pressure of 1 bar; air 150 l min^{-1} at a pressure of 2 bar) to a temperature of 350 °C at thermocouple level 2. The cooling rate during this test series was 1.52 K s^{-1}. Subsequently, the specimens underwent age-hardening treatment in a furnace for 6 h at 350 °C.

The samples for the determination of the mechanical properties and the microstructural characteristics were machined out from different levels of these specimens. The measurement of the retained austenite was made by X-ray examination. SEM Analysis of the fracture surfaces was made from the same samples.

RESULTS AND DISCUSSION

In the samples inoculated with Reseed®, a microstructure with predominantly austenite and acicular ferrite and few parts of perlite existed on level 2 after heat treatment. In the center of sample level 4, with the lower cooling rate, a microstructure with less austenite and acicular ferrite and more perlite existed. There were segregated zones in spite of the inoculation.

In the samples inoculated with Ultraseed®, a microstructure with austenite and acicular ferrite was found level 2 after the heat treatment. In the center of the sample level 4, a microstructure with austenite and acicular ferrite and a small amount of perlite was found. There were no segregated zones. More details concerning the microstructure follow.

Depending on the employed inoculant, and hence the size and distribution of the nodular graphite, clearly different contents of retained austenite were found after comparable thermal treatment (Fig. 1).

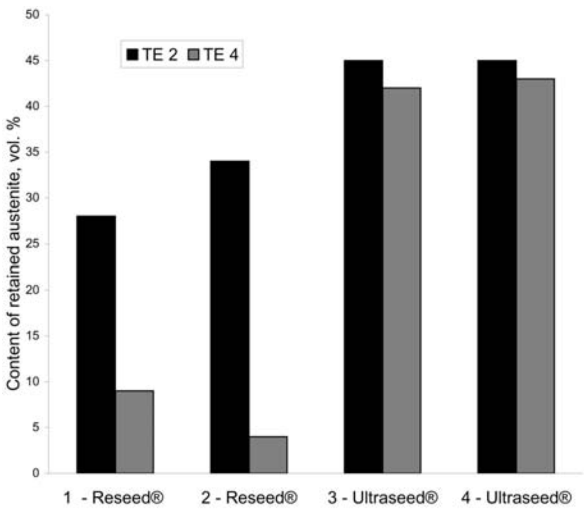


Fig. 1. Content of retained austenite at thermocouple levels 2 (TE 2) and 4 (TE 4) of the specimens subjected to different tests.

In the present case, smaller and more uniform nodules coincide with a smalller amount of retained austenite. Independent of whether water or a water–air mixture was adopted for cooling, the use of the inoculant Reseed® leads to a smaller amount of retained austenite and a higher content of perlite. The impact of the wall thickness was very pronounced. At the level of the thermocouple TE 2 (which means more rapid cooling than on the level TE 4) the amount of retained austenite amounted to ≈ 28 vol. %, respectively 34 vol. %. Inside of the specimen center at the level of thermocouple TE 4 (a lower cooling rate compared with TE 2) the amounts of retained austenite were only 9 vol. %, respecttively 4 vol. %. The castings inoculated with Ultraseed® feature a comparable amount of retained austenite of ≈ 42 to 45 vol. %, irrespective of the cooling conditions. This result confirms the experience that the use of the sulfur- and oxygen-containing inoculant leads to a reduction of the dependence of the wall-thickness during graphite formation.

As the content of retained austenite increases, the hardness decreases (Fig. 2). All hardness values measured are within the range typical of ADI materials.

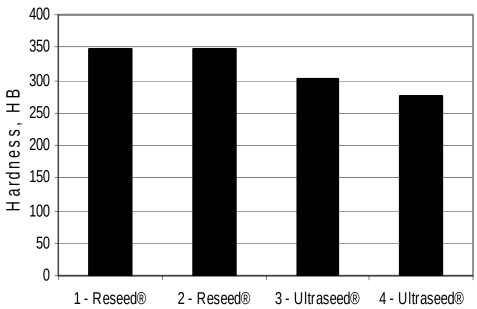


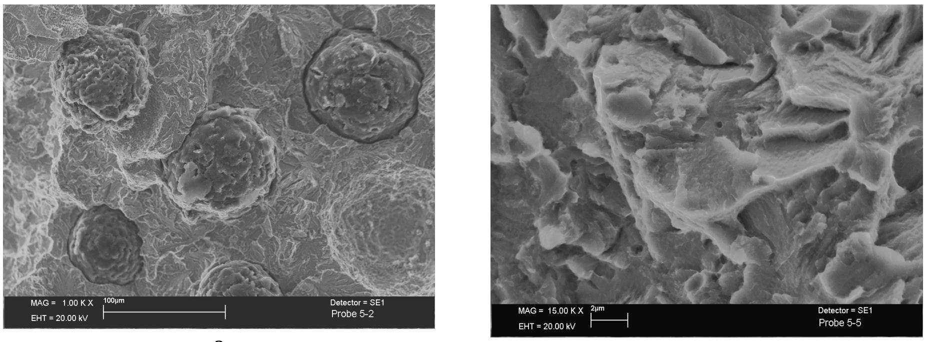
Fig. 2. Hardness values in the center zone of the specimens (TE 4).

The SEM (scanning electron microscope) analysis of the fracture surface of the tensile specimens from the center part of the castings (level 2) revealed a different crack extension as a function of the retained austenite in the basic microstructure.

Specimen 1 (thermocouple level 2) with ≈ 28 % retained austenite exhibited exclusively cleavage fracture without any signs or traces of plastic deformation (Fig. 3). This is demonstrated by the smooth plate-like to rosette-like configuration of the surface (Fig. 3b). The gap that exists between the graphite nodules and the metal matrix is probably caused by the relatively high amount of perlite in the specimen (Fig. 3a).

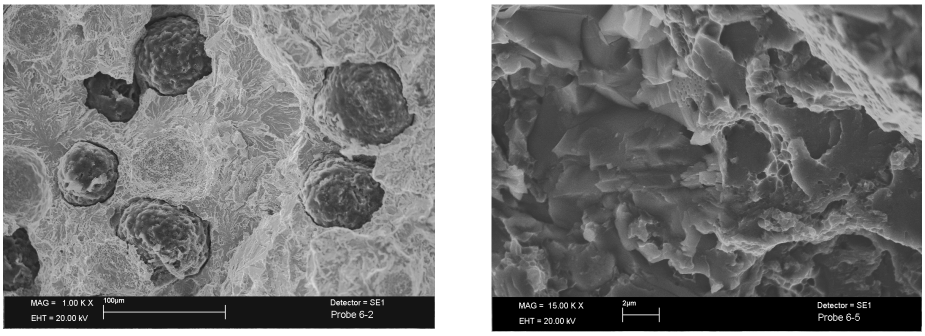
A further increase of the amount of retained austenite to ≈ 34 % (specimen 2, thermocouple level 2, Fig. 1) along with a corresponding reduction of the perlite content leads to mixed fracture, *i.e.*, mainly cleavage fracture with a few areas of ductile fracture (Fig. 4).

In specimens 3 and 4 (thermocouple, level 2) with an amount of retained austenite of ≈ 42 vol. %, respectively 44 vol. %, and only traces of perlite in the basic microstructure, a distinct increase of the amount of ductile fracture is recognizable in the fracture surfaces (Fig. 5). The share of cleavage fracture is markedly smaller. These results clearly show that ductile fracture occurs only in the areas of retained austenite. The share of ductile fracture corresponds to the amount of retained austenite.



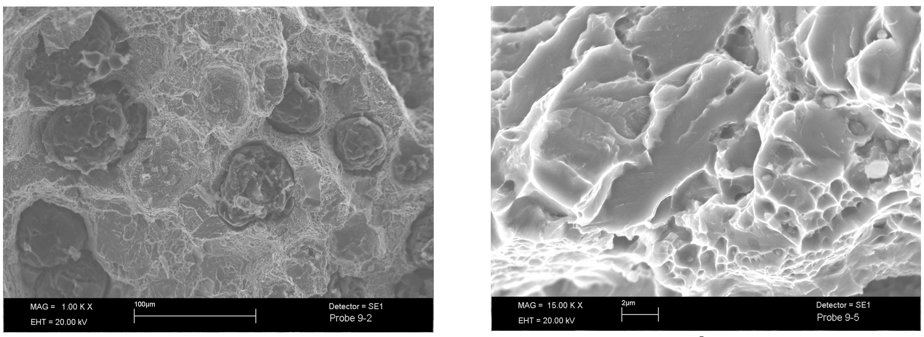
a b

Fig. 3. SEM Micrographs of the fracture surfaces of the specimen 1; in this present case there is only cleavage fracture.



a b

Fig. 4. SEM Micrographs of the fracture surfaces of the specimen 2; the area of the retained austenite shows ductile fracture.



a b

Fig. 5. SEM Micrographs of the fracture surfaces of the specimen.

The existence of perlite in the microstructure seems to be the primary cause of brittle fracture behavior. It is remarkable that, responding to a growing amount of brittle fracture, the separation between the graphite nodules and the metal matrix increases.

CONCLUSIONS

In the samples inoculated with Reseed® there was a nearly uniform nodular graphite formation over the full wall-thickness range. Under the given experimental conditions, the influence of different cooling rates on the microstructure is small. The microstructure was unfavorable for ADI.

In the samples inoculated with Ultraseed®, there were a lower nodule count and a less homogeneous nodule size over the entire cross section without segregations. This means a better microstructure for the ADI. Depending on the cooling rate, there are only traces of perlite. The influence of the cooling rate on the microstucture was slightly greater than when Reseed was employed.

The SEM analysis of the fracture surfaces from tensile test specimens of thick-walled austempered ductile irons (diameter 160 mm) showed different fracture behavior depending on the amount of austenite retained in the matrix. The results showed ductile fractures only in sections areas with retained austenite. In section areas without or with very low amount of retained austenite, only brittle fracture without any plastic deformation occurred. The content of retained austenite determines the amount of ductile fracture in the microstructure.

И З В О Д

УТИЦАЈ УСЛОВА ХЛАЂЕЊА И КОЛИЧИНЕ ЗАОСТАЛОГ АУСТЕНИТА НА ЛОМ АУСТЕМПЕРОВАНОГ ДУКТИЛНОГ ГВОЖЂА

VYACHESLAV GORYANY, ECKART HOFMANN и PAUL JOSEF MAUK

Institute of Applied Materials Technology (IAM), University of Duisburg-Essen, D-47048 Duisburg, Germany

SEM анализа површина лома узорака танкозидног, аустемперног дуктилног гвожђа тестираних на изтезање указује на различите карактеристике лома у зависности од количине аустенита заосталог у матрици. Дуктилни ломови су присутни само у областима које сардже заостале аустенитне секције. У областима са веома малим садржајем заосталог аустенита односно онима у којима није заостао аустенит јавља се крт лом без пластичне деформације. Садржај заосталог аустенита одређује удео дуктилног лома у микроструктури.

(Примљено 27. децембра 2005)

REFERENCES

1. A. Bedolla, B. Hernandez, *Trans. Am. Foundrymens Soc.* **105** (1997) 715
2. J. Hemanth, *Mater. Sci. Technol.* **15** (1999) 878
3. G. Coopert, A. Roebuck, H. Bayati, *Int. J. Cast Metals Res.* **12** (1999) 227
4. S. Hasse, K. Röhrig, *Gießerei-Praxis* **4** (1999) 154

5. J. Liu, R. Elliott, *Int. J. Cast Metals Res.* **12** (1999) 189
6. M. P. Shebatinov, A. A. Zhukov, W. M. Kowalenko, *Vestnik Mashinostroeniya* **3** (1986) 61
7. A. A. Zhukov, *Liteinoe Proizvodstvo* **11** (2000) 28
8. A. A. Zhukov, *Liteinoe Proizvodstvo* **11** (1999) 39
9. M. N. Ahmadabadi, *Metall. Mater. Trans.* **28A** (1997) 2159
10. Th. Hirsch, P. Mayr, *Konstruieren Gießen* **18** (1993) 25
11. R. Böschen, H. Bomas, H. Vetters, P. Mayr, *Konstruieren Gießen* **15** (1990) 34
12. T. J. Marrow, H. Cetinel, C. S. Wang, H. Bayati, *20th ASM Heat Treating Conference Proceedings*, 2000, St. Louis, USA, ASM International, 2000, p. 528
13. H. Vetters, J. M. Schissler, *HTM Härterei-Technische Mitteilungen* **55** (2000) 166
14. R. Kazerooni, A. Nazarboland, R. Elliott, *Mat. Sci. Techn.* **13** (1997) 1107
15. K. Herfurth, *Konstruieren Gießen* **16** (1991) 4
16. K. Röhrig, **11** (1986) 43
17. J. L. Garin, R. L. Mannheim, *Zeitschrift für Metallkunde* **91** (2000) 841
18. P. Mayr, H. Vetters, J. Walla, *Gießereiforschung* **38** (1986) 86
19. J. M. Schissler, J. P. Chobaut, *HTM Härterei-Technische Mitteilungen* **43** (1987) 205
20. M. Johansson, *2nd International ADI Seminar*, 1994, Helsinki University of Technology, Otaniemi, SF, p. 1
21. M. Takita, Y. Ueda, K. Shibayama, H. Hiramitsu, *Physical Metallurgy of Cast Iron IV, Proceedings of the 4. International Symposium*, Tokyo, Japan, 1990, Pittsburgh, PA, USA, p. 211
22. M. Takita, Y. Ueda, K. Shibayama, H. Hiramitsu, *Physical Metallurgy of Cast Iron IV, Proceedings of the 4. International Symposium*, Tokyo, Japan, 1990, Pittsburgh, PA, USA, p. 219
23. M. Takita, *Physical Metallurgy of Cast Iron IV, Proceedings of the 4. International Symposium*, Tokyo, Japan, 1990, Pittsburgh, PA, USA, p. 227
24. K. F. Laneri, J. Desimoni, R. C. Mercader, R. W. Gregorutti, J. L. Sarutti, *Metall. Mat. Trans. A* **32A** (2001) 51
25. I. Schmidt, A. Schuchert. *Konstruieren Gießen* **13** (1988) 18
26. K. Röhrig, *Konstruieren Gießen* **13** (1988) 4
27. C. Chengija, J. J. Vuorinen, M. Johansson, *Casting 1997, International ADI and Casting Simulation Conference*, Otaniemi, Helsinki, SF, 1998, p. 1
28. J. Aranzabal, I. Gutierrez, J.M. Rodriguez–Ibabe, J. J. Urcola, *Metal. Mater. Trans.* **28A** (1997) 1143
29. T. Toshio, A. Toshihiko, T. Shuji, *Metal. Mater. Trans. A* **27A** (1996) 1589
30. S. Yamamoto, H. Yokoyama, K. Yamada, M. Niikura, *ISIJ Int.* **35** (1995) 1020
31. H. Bayati, R. Elliott, *Mater. Sci. Technol.* **13** (1997) 319.