OVERVIEW

Development and Application of Austempered Spheroidal Graphite Cast Iron

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The superior combination of attainable properties has caused austempered spheroidal graphite cast iron to emerge as a new class of cast iron. The metallurgy of austempered spheroidal graphite cast iron is presented with the purpose of clarifying the mechanical properties. It is emphasized that segregation of alloying elements can cause an iron to behave differently than might be expected. The role of the graphite nodule count on the strength and the toughness of this iron is discussed from the standpoint of the matrix structure, because segregation can be controlled by varying the nodule count, i.e. solidification rate during casting process. High carbon-reacted austenite increases with increasing nodule count, but instable unreacted austenite decreases. With increasing nodule count, the tensile and fatigue properties are improved, but the elasto-plastic fracture toughness decreases. This behaviour is explained without contradiction by the microstructures including the reacted and unreacted retained austenites and the notch effect of the graphite. The austempered spheroidal graphite cast iron has very desirable characteristics for a wide range of applications as component parts of automobiles, construction machinery, railroad, etc.

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I. Introduction

Interest in austempered spheroidal graphite cast iron (ADI) has markedly increased in the past few years as increasingly successful applications are being reported⁽¹⁾⁻⁽⁵⁾. Austempering consists of heating to the austenitizing temperature, followed by quenching into a salt or oil bath at a temperature in the range of 720 to 470 K and holding for the time required for transformation to occur at this temperature. Mechanical properties most often discussed in describing these materials are the ultimate tensile strength (UTS) and the ductility such as elongation or reduction of area. Figure 1 shows a plot of UTS vs. elongation in which the solid line a) represents the best attainable combination of strength and ductility in ADI. ADI can be twice as strong as commercial spheroidal graphite cast iron (area c) in Fig. 1) at the same level of elongation and approaches the properties of a structural low alloyed steel (area b) in Fig. 1). In addition, they can exhibit excellent fatigue strength⁽¹⁰⁾⁽¹³⁾⁽¹⁴⁾ and wear resistance⁽¹⁵⁾. However, the strengths and elongations of ADI can be varied over a comparatively wide range by changing the austempering conditions and chemical compositions of the iron. Therefore, precise control of the austempering transformation is necessary to insure that the desired mechanical properties can be obtained. This

includes careful control of iron chemistry and quality and strict control of a multitude of heat treatment variables including austenitizing temperature and austempering temperature and time. An understanding of the relationship between the austempering transformation and the resultant microstructures developed is key to effective process control.

Fig. 1 Comparison of tensile properties for unalloyed and alloyed ADI, commercial s.g.iron grades and structural low alloyed steels.

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This paper presents current thoughts on the reactions occurring in spheroidal graphite cast iron (s.g.iron) during the austempering. The evidence that alloying can hinder or enhance the response of s.g.iron to austempering heat treatments is presented. This paper will also discuss some aspects of the development work for successful applications of ADI in Japan.

II. Austempering Transformation

1. Austempering reaction

The austempering transformation in cast iron is different from that in steel. In most steels austempering results in essentially simultaneous formation of ferrite and carbide from austenite⁽¹⁶⁾. The ferrite and carbide phases nucleate and grow cooperatively over a relatively short time period. However, the desired matrix structure of ADI is quite different from steels as shown in Fig. 2. It consists of a two-phase mixture of bainitic ferrite and retained austenite which has been named “Ausferrite” by Kovacs⁽¹⁷⁾.

The transformation stages of ADI are illustrated in Fig. 3 which shows a schematic phase diagram and free energy curves for ferrite (α), austenite (γ) and cementite (Fe_3C) to illustrate qualitatively the thermodynamic parameters. The high silicon content in cast iron suppresses the formation of cementite phase normally associated with bainitic transformations of steels. The austempering transformation initially obeys the metastable equilibrium relationship between ferrite and austenite defined by the tangent line between the α and γ free energy curves. When the austenite is quenched from T_{γ_1} to the austempering temperature T_a , it assumes a free energy given by point f_a . The driving force for this stage of the transformation is the free energy difference ($f_a - f_b$). At the completion of the stage the free energy has fallen to f_b and the matrix structure consists of bainitic ferrite and high carbon austenite. This reaction is often referred to as the stage I reaction. The high carbon content depresses the M_s temperature below 193 K⁽¹⁸⁾, so

Fig. 2 Typical microstructure of ADI. [γ : Austenite, α : Bainitic ferrite]

that the austenite is stable after cooling to room temperature. However, the austenite is not stable indefinitely. Stage II begins with an accelerated precipitation of carbides followed by sideways growth of ferrite plates and disappearance of the austenite phase. Eventually the free energy falls to point f_c , where metastable equilibrium is reached between ferrite and cementite.

The kinetics of stage I and stage II reactions have been investigated by using the volume fraction of austenite and ferrite deduced from X-ray measurement and the magnetic induction method⁽¹⁹⁾, respectively. The variations of these parameters with austempering time at temperatures greater than approximately 620 K is shown in Fig. 4⁽¹⁹⁾. During the first stage with a high average transformation rate, a relatively fast ledge growth of ferrite plate occurs into the austenite. At the same time

Fig. 3 Fe-C-Si phase diagram with schematic free energy diagram to represent the austempering process.

Fig. 4 Schematic representation of austempering reaction at higher austempering temperature. [M : Martensite, γ : Austenite, α : Bainitic ferrite, γ_{HC} : High-carbon austenite]

carbon is rejected from the growing ferrite plates into the surrounding austenite. On cooling down from the transformation temperature, a considerable amount of martensite is formed. The presence of martensite will significantly increase hardness, but at the same time, reduce ductility and toughness, especially at the beginning of this stage. With increasing holding time of isothermal transformation, the amounts of both ferrite and austenite increase and the amount of martensite is reduced in the final structure. In the later period of the stage the reaction rate is low, the transformation continues and the austenite continues to absorb carbon as high as two percent⁽¹⁸⁾. The high carbon content depresses the M_s temperature so that austenite is stable after cooling to room temperature. The amount of retained austenite reaches a plateau. Combination of the tensile strength and elongation reaches their maximum values at the present stage.

The second stage is characterized by the increased average transformation rate and reduced portions of austenite. The mixture of ferrite and carbide is found in area where bainitic ferrite plates have grown into each other. The disappearance of high carbon austenite as a consequence of carbide formation reduces ductility of ADI.

Consequently, the austempering transformation on cast iron can be described as a two-stage reaction^{(19)–(22)} with (1) austenite γ decomposing to bainitic ferrite α and high carbon austenite γ_{HC} , i.e. $\gamma \rightarrow \alpha + \gamma_{HC}$, followed by (2) further austenite decomposition to ferrite and carbide, i.e. $\gamma_{HC} \rightarrow \alpha + \text{carbide}$.

Recent investigation by Moore *et al.*⁽²³⁾ indicates that the first stage of the austempering reaction at lower transformation temperatures is accompanied by carbide formation either in the ferrite or at the moving ferrite/austenite interface. A simple diffusion calculation for carbon in fcc and bcc iron supports this evidence. At 673 K the ratio of the diffusion coefficient of carbon in ferrite divided by the diffusion coefficient of carbon in austenite is $(D_\alpha/D_\gamma)_{673} = 4700$; while at 573 K this ratio increases to $(D_\alpha/D_\gamma)_{573} = 36600$. These results are obtained by evaluating the following expressions⁽²¹⁾:

$$D_\alpha = 0.0079 \exp(-18100/RT) \text{ cm}^2/\text{s}$$

$$D_\gamma = 0.21 \exp(-33800/RT) \text{ cm}^2/\text{s}.$$

At lower transformation temperatures the austenite cannot readily accept the carbon rejected from the growing ferrite which results in carbide formation even during the first stage⁽²¹⁾. Stage II reaction also occurs at the low transformation temperatures but is delayed to much longer transformation times. Some investigators⁽²⁴⁾⁽²⁵⁾ have found precipitation of epsilon carbides within the bainitic ferrite laths of the lower temperature bainites. Carbides are not always evident in bainitic ferrite produced at higher temperatures, and it is likely that carbon has time to diffuse out of the growing bainitic ferrite plates into the surrounding austenite⁽²⁶⁾.

Fig. 5 Influence of austempering time at 623 K on the mechanical properties of low alloyed ADI and idealized characterization of bainitic reaction on γ_{HC} . [Nominal composition: 3.6% C, 2.7% Si, 0.11% Mn, 0.63% Cu, 0.82% Ni]

2. Process window

The nucleation and growth processes involved in stage I and II are controlled at rates governed by the variables of the process. However, it is thought that each stage contributes to the amount of high carbon austenite. Therefore, progress of the two austempering reactions can be followed in a graph of mechanical properties vs. austempering time with the schematic representation of γ_{HC} as shown in Fig. 5. Point A signifies completion of the first reaction ($\gamma \rightarrow \alpha + \gamma_{HC}$) and point B signifies the onset of the second reaction ($\gamma_{HC} \rightarrow \alpha + \text{carbide}$). The time interval between points A and B, the so-called “process window”⁽²²⁾, represents the allowable austempering time for processing of this iron to obtain optimum mechanical properties. Shorter austempering times would result in an iron having some martensite, while times beyond point B would result in a bainitic iron in which the embrittling reaction had already progressed to a detrimental degree. In any event, an iron not austempered within the process window would have less than the optimum properties, especially in ductility. For commercial austempering, a wide process window is desirable.

3. Factors affecting the process window

The transformation response of s.g.iron during austempering depends on the complex interaction between iron chemistry, segregation characteristics, austempering temperature and austenitizing temperature. These factors affect both the nucleation kinetics of the stage I reaction and of the austenite decomposition during the stage II, also the span and the level of the ductility plateau of the process window.

(1) Austempering temperature

Austempering temperature influences both the transformation kinetics and the details of the transformation reaction itself. Because both the kinetics and mechanism of transformation are temperature depend-

Fig. 6 Effect of austempering temperature on the microstructure of ADI. [Austempering temperature: 1173 K]

Fig. 7 Effect of austempering temperature on tensile properties and hardness. [Austempering time: 1 h for unalloyed and 3 h for alloyed iron. Nominal composition: 3.8%C, 2.8%Si, 0.32%Mn for unalloyed iron and 3.65%C, 2.7%Si, 0.11%Mn, 0.635%Cu, 0.82%Ni for alloyed iron]

ent, the austempering time/temperature response is complex. In general, both the initiation and completion of transformation are retarded with decreasing austempering temperature owing to the decrease of the diffusion rates. This is also accompanied by a decrease in the bainitic ferrite/high carbon austenite spacing as austempering temperature decreases as shown in Fig. 6. With increasing austempering temperature the UTS, elongation and the hardness decrease as shown in Fig. 7. The low values of both elongation and strength at the low temperatures can be explained by premature fracture of the tensile test specimens, which is due to the increased amount of martensite or carbide in the matrix. The level of the ductility plateau between points A and B as in Fig. 5 would be shifted downward for lower austempering

temperatures. It must be kept in mind that at lower transformation temperatures the multiple lath structure of ferrite and austenite more resembles that of an upper bainite in steels.

The initiation of the stage II reaction is much more difficult to detect metallographically. The times for points A and B, and therefore the span of the process window can be obtained from the impact screen test results⁽²²⁾ or the tensile test by using many specimens to screen for the time-temperature regions providing high elongation.

(2) Chemical composition

Alloy additions are often necessary to increase the hardenability of s.g.iron by avoiding formation of pearlite and Widmanstatten ferrite during cooling to the austempering temperature. However, the composition must be compatible with production of quality iron, free of carbides. Segregation must be minimized to ensure uniform transformations as would be noted below. Alloy additions for promoting hardenability are limited primarily to Mn, Cu, Ni and Mo used alone or in combinations. Other hardenability promoting alloying elements can result in undesirable poor nodularity of graphite nodule or as-cast carbides in the iron upon solidification. The alloy additions delay not only the initiation of the austempering reaction but also the embrittling effect of the second reaction, thereby creating a wide process window to get high ductility. Some works⁽²²⁾⁽²⁸⁾ indicate that Ni-Mo and Cu-Mo additions delay the stage I reaction slightly, but delay the embrittling effect of the stage II reaction much more, thereby creating a wide process window for austempering for high toughness. These are shown by TTT diagrams in Fig. 8 which show the effect of Ni and Mo on hardenability. Therefore, stage I and stage II reactions are shifted to longer times and the alloyed iron has a process window A'-B' to be increased significantly over the unalloyed iron as shown schematically in Fig. 9, where the alloy elements has been assumed to be distributed homogeneously throughout

Fig. 8 Time-temperature transformation curves for unalloyed and Mo-Ni alloyed s.g. irons austenitized at 1173 K.

Fig. 9 Idealized effect of alloy addition and segregation on the level and span of the process window at high austempering temperature. [(a): unalloyed iron (—), (b): alloyed iron with no-segregation or bulk of matrix (—), (c): area of cell boundary (---), (d): iron with alloy segregation (· · · · ·)]

the matrix.

(3) Alloy segregation

During solidification most elements partition between the liquid and solid phase according to equilibrium distribution coefficient. In as-cast pearlitic s.g. irons the eutectic cell boundaries are delineated by darker etching regions, which have been shown to be associated with concentration of Mn, Mo and other elements and depletion of Si, Ni and Cu as illustrated in Fig. 10⁽²⁹⁾⁽³⁰⁾. The degree of segregation increases with cast section size because the solidification rate is slower and the freezing distance, that is, the distance from the dendrite center to the center of interdendritic volume is longer, allowing more time for partitioning. These segregation effects are not generally removed by normal heat treatments, which involve heating in the austenite range⁽³⁰⁾. These conditions favor the formation of intercellular carbides which can significantly deteriorate ductility and toughness. Therefore, it is important to recognize that the composition of the iron matrix is not homogeneous and that the cell boundary can have a very different chemistry from

Fig. 10 Schematic plots of segregation for some alloy elements in s.g. iron.

the nominal composition.

In a non-homogeneous matrix different areas will transform at different rates. This can seriously restrict the process window. The stage I reaction starts with the formation of ferrite in suitable places, such as the interface of austenite-graphite and austenite grain boundaries. In the eutectic cell boundary, the time for completion of the stage I reaction is very long, see line (c) in Fig. 9, because of the strong austenite stabilizing effect. While waiting for the reaction to go to completion in the cell boundary, the bulk of the matrix could have reached its embrittling stage, as shown by line (b) in Fig. 9. The toughness behavior of the iron would be a composite of two regions, resulting in a curve which has a low toughness maximum and short time window A*-B* as shown by line (d) in Fig. 9. The net effect of segregation will be to reduce the level and the span of the process window. Whether the window A'-B' is reduced to a shorter time will depend on the processing variables and the alloy content.

The matrix structure of ADI with segregation of alloy elements consists of a mixture of bainitic ferrite and high carbon austenite, and untransformed austenite in the intercellular areas. Heat tinting technique⁽³¹⁾ causes the various phases to appear metallographically in distinct colors. The high carbon-bainitic (reacted) austenite and the untransformed (unreacted) retained austenite are represented by purple and light blue color as shown in Fig. 11. It has been thought that the most undesirable phase is the unreacted austenite for the mechanical properties of ADI. Figure 12 shows the result that each microstructure is determined quantitatively by point-counting the distinct colors for each phase⁽³²⁾. The unreacted austenite is decreased by decreasing the mean graphite nodule diameter but the reacted austenite shows the reverse behaviour to the nodule size. The increase of the amount of high carbon austenite results from the reduction in segregation in the area of eutectic cell boundary due to high graphite nodule counts.

As seen in a later chapter, the role of nodule count on

Fig. 11 Typical microstructure of ADI obtained by heat tinting technique.

Fig. 12 Effect of mean graphite nodule diameter on the amount of retained austenite for various austempering temperatures and times.

the mechanical properties of ADI will be discussed from the viewpoint of the nature and degree of segregation of the respective elements in the matrix of ADI.

(4) Austenitizing temperature

Austenitizing temperature has a direct effect on the austempering transformation kinetics through its effect on the carbon content of the austenitized matrix. The carbon content of the austenitized matrix can range from 0.6–1.1% depending on the austenitizing temperature and time and the composition of the iron^{(33)–(35)}. This will affect not only the final structure and properties of ADI but also the transformation kinetics strongly. Increasing the austenitizing temperature decreases the initial rate and increases the time for the stage I transformation. The kinetics measurements shown in Fig. 13 show that increas-

Fig. 13 Time to reach 60–80% completion of Stage I for austenitizing temperature $T_\gamma = 1144$ K (.....) and $T_\gamma = 1200$ K (—) for irons of different alloying content⁽³⁶⁾.

ing the austenitizing temperature from 1144 to 1200 K increases the time for 60–80% transformation from 3.1 to 7.1 min⁽³⁶⁾.

The influence of austenitizing temperature can be explained using the free energy composition diagram in Fig. 3. When T_γ is reduced to T_{γ_2} , the accompanying reduction in carbon content produces a greater driving force ($fa' - fb'$) for the stage I transformation. The greater driving force increases ferrite nucleation and the initiation transformation rate and provides a greater activity gradient. This drives carbon diffusion and produces a more rapid ferrite plate growth. At the present time it is thought that the influence of austenitizing temperature does not have a significant effect on the stage II reaction. Consequently, the process window narrowed by segregation may be enlarged by reducing the austenitizing temperature and allows the optimum mechanical properties to develop.

III. Improvement of Mechanical Properties of ADI

The segregation severity of alloying elements increases as the eutectic cell size increases due to the low nodule counts or low cooling rates. Therefore, the graphite nodule count remarkably influences the mechanical properties of ADI. The mechanical properties such as tensile properties, fatigue properties and elasto-plastic fracture toughness are discussed to evaluate the combination of strength and toughness of ADI.

1. Tensile properties

The UTS/elongation relationship of ADI is examined by re-plotting the data of Fig. 1, see Fig. 14⁽⁸⁾. If the heat treatment parameters, austempering temperature and time are varied, the properties will move along the solid lines in Fig. 14. It is found that high tensile strength is obtained at the low austempering temperature, where elongation shows lower values. This is caused by the formation of martensite or carbide as mentioned above. ADI treated at the low austempering temperature is used

Fig. 14 Influence of austempering temperature, time and alloy additions on the U.T.S./elongation relationship. [Austenitizing: 1173 K–3600 s, Symbol ★: varied graphite nodule count from 200 to 1400 nodules/mm²]

when the highest strength and wear resistance are required. The relatively high tensile strength and maximum elongation are obtained at the austempering temperature of about 620 K and austempering time from 3 to 5 h. The combination of strength and elongation falls off from optimum value as the temperature increases for alloyed and unalloyed irons. The properties of the alloyed iron have low values in comparison with that of the unalloyed iron due to segregation of alloy elements. However, a high graphite nodule count in iron matrix makes it possible to improve the combination, especially in elongation as shown by dotted line in Fig. 14. The remarkable combination of properties attainable has caused this material to emerge as a new class of s.g.iron, when comparing its combination with JIS requirements for s.g.iron. These irons would offer wider possibilities of engineering applications as a potential replacement material for steel.

Figure 15 shows the relationship between tensile properties and mean graphite nodule size of the specimens which is treated at 623 and 673 K for 3 and 5 h, respectively. Tensile strength and elongation are increased with decreasing nodule diameter. The rise and fall in elongation with decreasing nodule diameter can occur in the 673 K treated iron. Maximum elongation at 673 K is obtained at the nodule diameter of about 15 μm . The drop of elongation for the iron having smallest nodule diameter may be considered as the result of the formation of carbide due to onset of the stage II reaction. Because, the homogenization of alloying elements by increasing of graphite nodule consequently accelerates the second reaction at the center of eutectic cell, allowing a structure of bainite and carbide to occur.

However, both the tensile strength and elongation are increased simultaneously at a shorter austempering time of 30 min to 1 h at 673 K. Therefore, it is possible that simultaneous improvements of strength and elongation are accomplished by decreasing of nodule diameter, that is, by increasing of graphite nodule count.

Figure 16 shows the results by EPMA analyses of ADI

Fig. 15 Effect of mean graphite nodule diameter on tensile properties.

where s.g.irons with high and low nodule count are austempered at 623 K for 5 h. The curves show that the concentration of Mo is much lower in the cell boundary region of specimen which has the smallest nodule size. As the segregation of Mo acts to delay the beginning of stage I reaction, the unreacted austenite region has limited width for the iron having high nodule count. Molybdenum silicate is easily formed at the eutectic cell boundary. The frequency of silicate formation is varied according to the degree of nodule size and it is rarely detected in the specimen with high nodule count. Fine martensite is also found in the unreacted austenite on the fracture surface after the tensile test, which may be considered a strain induced martensite formed during tensile deformation. From the analysis of the quantity of retained austenite at vicinity of the fracture surface, the reacted high-carbon austenite does not change in its quantity before and after the deformation, but the metastable unreacted austenite is decreased after the deformation as shown in Fig. 17. The tendency becomes stronger with increasing nodule size.

It is well known that unalloyed ADI indicates higher strength and elongation than alloyed ADI as shown in Fig. 14 and this is attributable to the martensite transformation in the latter. Therefore, the formation of strain induced martensite and precipitation of compounds, as a consequence of the formation of large amount of unreacted austenite, cause a premature initiation of microcracks, and both the tensile strength and elongation may decrease due to premature fracture during the tensile deformation.

It is concluded that high strength and high elongation in ADI are obtained by controlling the segregation of alloying elements at eutectic cell boundary, that is, by decreasing the area of the cell boundary or by increasing the nodule count.

2. Fatigue properties

Fatigue properties obtained in ADI containing alloy

Fig. 16 EPMA analyses of C, Ni, Si and Mo of ADI with (a) 200 nodules/mm², (b) 1400 nodules/mm² of graphite.

Fig. 17 Effect of mean graphite nodule diameter on transformation of unreacted austenite to martensite after tensile test.

elements and in various s.g. irons with ferritic, bull's eye and pearlitic matrix structure are shown in Fig. 18. The bending fatigue strength σ_{wp} are increased by increasing the tensile strength. Such a tendency has been generally recognized for various s.g. irons when iron hardness is not greater than about Hv 500 in the matrix⁽⁴⁰⁾ such as ferritic and bull's eye s.g. irons. The structural change in matrix which raises the hardness does not raise the endurance ratio (m in Fig. 18) in proportion to the increase in tensile strength and may also in some cases lower the

Fig. 18 Relationship between fatigue limit and U.T.S. for alloyed ADI and other forms of s.g. irons. [(a): Austempered at 673 K for 3 h, (b): at 623 K for 3 h]

endurance limit. Alloyed ADI also have a lower endurance ratio than pearlitic s.g. irons as shown in Fig. 18. However, an ADI with reduced nodule diameter by controlling the rate of solidification appears to have excellent high fatigue strength in proportion to the reduction of nodule count or the increase of tensile strength as shown by lines (a) and (b) in Fig. 18.

Figure 19 shows the relationship between fatigue limit and mean diameter of graphite nodule. Fatigue limit is linearly increased with decreasing nodule size, regardless of austempering condition. Sofue⁽⁴⁰⁾, Niimi *et al.*⁽⁴¹⁾ and Shiota *et al.*⁽⁴²⁾ have reported that fatigue limit of ferritic s.g. iron is increased with decreasing nodule size. The graphite nodules are considered as a sort of internal

Fig. 19 Effect of mean graphite nodule diameter on the bending fatigue strength for ADI.

Fig. 20 Relationship between the fatigue limit and the amount of unreacted austenite.

defect, because their strength and hardness are negligibly low as compared with the matrix iron⁽⁴⁰⁾⁽⁴¹⁾. It is well-known that fatigue strength of metals is lowered by a larger defect than by a smaller one, especially when the strength of metal is high. If the concept of fatigue notch factor is applied to graphite nodule, the notch factor of graphite nodules β can be defined as follows:

$$\beta = \sigma_s / \sigma_{wp}$$

where σ_s is fatigue limit of the matrix of iron, σ_{wp} is fatigue limit of the iron specimen. Apparent notch factor β is smaller when the iron contains smaller graphite nodules⁽⁴⁰⁾⁻⁽⁴²⁾. However, as the tensile strength of these irons does not improve remarkably, the significant improvement of the obtainable fatigue limit is probably attributed to the change of matrix structure.

The high values in fatigue limit are obtained with decreasing the amount of unstable unreacted austenite as shown in Fig. 20. It is suggested that the fatigue limit is affected by the amount of reacted austenite. Fatigue cracks in unnotched specimens for s.g.iron are initiated at surrounding of the graphite nodules which acts as the portions of high stress concentration and then fracture is occurred by the linking of isolated microcracks generated from graphite nodules ahead of the main crack⁽⁴²⁾⁻⁽⁴⁴⁾. However, in ADI crack initiation at the graphite nodules is depressed by reacted-high carbon austenite. On the other hand, the unreacted austenite containing non-metallic inclusions is increased with increasing nodule size, that is, with decreasing nodule counts as shown in Fig. 12. The unreacted austenite is easily transformed to the martensite under plastic deformation as mentioned above. The notch effects of non-metallic inclusions in the unreacted retained austenite are increased by this martensitic transformation. Figure 21 shows microcracks initiated at the cell boundary area in the ADI having low nodule count. However, the initiation of microcracks is depressed by increase of nodule count⁽³¹⁾.

The improvement of fatigue strength is accomplished when the nodule size is reduced in order to decrease the notch effects due to graphite, the most suitable austempering is applied to the formation of large amount of reacted austenite and crack initiations at the neighborhood of graphite are depressed by the formation of a fine mixed structure of bainite and stable reacted austenite which increases the tensile strength. In addition, the reduction of unreacted austenite at the eutectic

(a) ADI having low nodule count.

(b) Enlargement of region A.

(c) ADI having high nodule count.

(a), (b): $\sigma = 216 \text{ MPa}$, $N = 2 \times 10^6$ cycles

(c): $\sigma = 490 \text{ MPa}$, $N = 2 \times 10^6$ cycles

Fig. 21 Scanning electron micrographs of the surface of fatigued unnotched specimens. (a) and (b) show the stress-induced martensite with many micro-cracks.

cell boundary greatly improves the fatigue strength.

The useful procedures for improvement in fatigue strength of ADI are the reduction of nodule size which leads to the refinement of mixed structures of bainitic ferrite and austenite and the increment in stable austenite, and the induction of compressive stresses at the surface by shot peening⁽³⁹⁾⁽⁴⁵⁾⁽⁴⁶⁾ or surface rolling⁽⁴⁶⁾.

3. Fracture toughness

Fracture toughness K_{IC} of ADI achieves high strength levels comparable with that of pearlitic and bull's eye s.g. irons as shown in Fig. 22. It can be found that the toughness decreases with increasing the tensile strength, where the data vary considerably from investigator to investigator. The higher toughness is obtained by treating

Fig. 22 Relationship between fracture toughness K_{IC} and U.T.S. for ADI and commercial s.g. irons.

in the process window. However, an increase of graphite nodule count does not improve the fracture toughness. The iron has the same level as the toughness of pearlitic irons as shown by lines (a) and (b) in Fig. 22.

Elasto-plastic fracture toughness J_{IC} for ADI with an identical chemical composition is arranged according to the graphite nodule size in Fig. 23. J_{IC} decreases with decreasing mean diameter of graphite nodule or increasing nodule counts for each austempering temperature. This relation between J_{IC} and the nodule size in ADI is consistent with the data on ferritic s.g. irons which have been reported by Komatsu *et al.*⁽⁵⁰⁾, Maezono *et al.*⁽⁵¹⁾ and Bhandhubayong *et al.*⁽⁵²⁾ The fracture surface of specimens after J_{IC} test consists of the dimple pattern as shown in Fig. 24. Therefore, the development of cracks

Fig. 23 Effect of nodule diameter on elasto-plastic fracture toughness J_{IC} for ADI.

(a) ADI having high nodule count.

(b) ADI having low nodule count.

Fig. 24 Scanning electron micrographs of fracture surface after J_{IC} test.

for ADI occurs by ductile tearing fracture in the same way as that of ferritic s.g.iron. The formation of dimples may be initiated due to nodule-cavity separation which can occur ahead of the main crack front. It is clear from Fig. 24 that graphite nodules are never seen to be cut by propagation of a moving crack. ADI with a larger mean free distance of graphite nodules is required to yield in a larger scale, and then the cavities are combined with each other at the front of the crack tip. The energy to propagate the crack increases with increasing the mean free distance of graphite nodules, because large plastic deformation is required to tear the matrix. Therefore, J_{IC} increases with increasing nodule size.

The dependence of K_{IC} on the graphite nodule count has been proposed by Wolffensberger *et al.*⁽⁵⁴⁾ as follows:

$$K_{IC} = C(\varepsilon_f \cdot \lambda \cdot \sigma_{0.2} \cdot E)^{1/2}$$

where C is a constant, ε_f is the fracture strain, λ is the mean free distance of graphite nodules, $\sigma_{0.2}$ is the 0.2% offset strength and E is the Young's modulus. As the linear relationship between K_{IC} and $(\varepsilon_f \cdot \lambda \cdot \sigma_{0.2} \cdot E)^{1/2}$ holds for s.g.irons⁽⁵³⁾, ADI treated in the conditions in the process window strongly depends on λ , because the other factors remain almost unchanged. Therefore, it is also induced that the decreasing λ or increasing nodule count reduce K_{IC} .

On the other hand, it is clear from Fig. 23 that J_{IC} is dependent upon the microstructures as well as upon the nodule size. Figure 25 shows that J_{IC} increases linearly with increasing quantity of the unreacted austenite, but it is not dependent upon the austempering temperature. It can be expected that J_{IC} is improved by the unreacted austenite. The unreacted austenite is predominantly formed in the eutectic cell boundary region and this region indicates low Si and Ni and high Mo concentrations. As the transformation of austenite to bainite is delayed at this area, unreacted austenite is held as the retained austenite. Therefore, this metastable unreacted austenite is transformed to martensite under the applied stress with large strain. During the J_{IC} test, as the martensitic transformation can take place at the head of crack tip, the stress concentration at the crack tip is

Fig. 25 Effect of unreacted austenite on elasto-plastic fracture toughness J_{IC} for ADI.

Fig. 26 Effect of nodule count on hardness, tensile properties, elasto-plastic fracture toughness J_{IC} and fatigue limit for alloyed ADI.

relaxed and the crack propagation is depressed.

It is concluded that high J_{IC} is obtained in the specimens which have a large nodule size and a large amount of untransformed retained austenite.

The mechanical properties of ADI are rearranged in relation to graphite nodule count in Fig. 26⁽³²⁾. The ultimate tensile strength, elongation, hardness and fatigue strength are improved with increasing nodule counts, but the elasto-plastic fracture toughness shows a reverse phenomenon. This is a normal relation between the increased strength and the decreased toughness. High strength in material means the difficulty in deformation, therefore, the stress concentration at stress raiser cannot relax during the deformation.

Though it is hard to improve both the strength and toughness at the same time, the toughness in ADI is 2–3 times higher than that obtained in pearlitic cast iron or in quenched and tempered cast iron with the same tensile strength. This good toughness of high strength ADI is due to the microplasticity associated with both the fine ferrite and the retained austenite phases. The production of high quality s.g.iron casting with high nodule counts may be the initial step to obtain high strength and toughness irons.

IV. Application of ADI

In the past twenty years ADI has changed from experimental interest to a competitive engineering material. In the early seventies limited production of ADI was initiated in the United States and Europe and in the late seventies many practical applications were opened in the

Table 1 Examples of applications for ADI.

Classification	Parts
Automotive industry	crankshafts, camshafts, timing gears, tie rods, suspension arms, differential gears, CV joints, brake components, ball joint sprockets, pump components, engine mounting brackets, gear cases, spindle supports, connecting rods, torque arms, planer carrier, drive yokes
Construction and Industry	truck-rear spring pads, truck-contact seats, digger teeth, slides, yokes, collars, structural members, sprockets, connectors, track-plates, track-shoes, end connectors, gears, crankshaft, camshafts, snow shoes, bulldozer-cases, chain sprockets, housings, concrete mixer car-shaft outputs, pump components
Agriculture	plow points, chisels, tillage points, sickle, guards, tie rods, slip clutches, hitches, pins, fertilizer, lift arms, lift crank, power take off components
Railroad	wheels, suspensions parts, covers, shoes, tie bars, hardware
Military	ordnance, armors, truck components, other vehicle components

industrially advanced nations. The production of ADI in Japan is estimated to be about 8000 t in 1990 which is seven times that of five years ago. It is estimated that ADI in the world may be produced in excess of 50000 t in 1991⁽⁵⁵⁾. Now is the period of trial applications to industrial components and ADI will be used in large quantities in the near future.

ADI now have wide applications as parts in automotive, construction, agricultural, mining, military and railroad components worldwide as shown in Table 1⁽¹⁾⁻⁽⁵⁾⁽⁵⁵⁾⁻⁽⁶⁵⁾. The driving forces to promote the use of ADI for engineering parts are as follows⁽⁴⁾:

(1) To reduce manufacturing cost by replacing conventional steel parts.

(2) To reduce weight and down-size by replacing conventional s.g.iron parts.

Figure 27 shows ADI-sprockets for chain-block⁽¹⁾ which have replaced steel components. The sprockets were formerly produced by forging of steel and then carburizing and heat-treating after a complicated process of machining. This complex method of manufacturing requires a high investment in equipment and labor. The trial sprockets showed satisfactory performance without breakage, spalling and other troubles under an excessively high-testing load and showed good resistance to wear against chains. The production processes for ADI-sprocket are as follows:

(1) Castings are machined with ease in the state of an as-cast ferritic matrix structure.

(2) Austenitizing at the temperature of 1173 K for 1 h.

(3) Austempering at the temperature of 593 K for 30 min in the salt bath and then air-cooling to the room temperature. For the castings with heavy section size, a small amount of Cu or Sb is added and it is heat-treated by the stepped austempering in order to obtain amore uniform

Fig. 27 Chain sprockets.

structure⁽¹⁾.

It may be noted that the grinding process after austempering are omitted by previously correcting the dimensional increase during austempering. Thus, the remarkable saving in manufacturing costs and energy consumption can be achieved by using ADI. Since the start of commercial production in 1978, the ADI-sprockets manufactured have amounted to about 2000000 pieces.

Figure 28⁽⁵⁷⁾ shows a ball joint socket which is one of the heavy-duty components for joining front wheels to the suspension arm. Before replacement, it has been machined after forging and quench-tempering the carbon steel for machine structural uses. Therefore, the adoption of ADI makes it possible to reduce the machining cost drastically, because the machining can be conducted in an easily machinable as-cast state.

Other examples for automotive application are shown in Fig. 29⁽⁴⁾. The engine mounting bracket in Fig. 29(a) is used to connect an engine to the body and imposed a load of high stress levels. It must not be broken to prevent the engine from becoming dislodged on collision. Therefore, it requires a high fatigue strength and impact properties. The replacement from traditional welded steel to ADI made it possible to reduce their weight and size.

Camshafts for the V-6 DOHC/SOHC gasoline engine in Fig. 29(b) are used in the roller-type valve operating mechanism in order to improve fuel economy by reducing internal friction. During the operation they suffer an exceedingly high stress at the contact surface to cause pitting trouble. Conventional forged steel with induction-hardening or chilled cast iron has been used. The ADI camshaft was designed to improve reliability and to provide cost reduction.

ADI-gear case incorporated into the transaxle for the 4-wheel-drive vehicle in Fig. 29(c) is used for the center differential gear. It is replacing conventional s.g.iron (JIS:FCD 700 grade) because of its higher load-carrying capacity and durability. The complicated shape of the gear case is difficult to make by any other processes except by casting. The ADI-gear case which has nearly equal mechanical properties as forged steel, makes it possible to design a high strength parts without increasing the size. Utilization of ADI is possible to omit the induction hardening process, because of the high wear

Fig. 28 Ball joint socket.

Fig. 29 Automotive parts replaced to ADI. [(a) Engine mounting brackets, (b) Camshafts for 4WD-center differential, (c) Gear case]

Fig. 30 Contact seat for truck.

resistance of ADI itself.

The contact seats⁽⁶⁵⁾ for a truck with two rear drives shaft support a truck load through the leaf springs and are subjected to the sliding wear induced by deflection of the leaf spring as shown in Fig. 30. To obtain high wear resistance and high strength, it is austempered at 643 K for 2 h from the austenite of 1158 K.

ADI is a material having a good combination of

strength, toughness and wear resistance as mentioned above. However, one of the problems with this type of material is that the machinability has been found to be low in many cases. It is well-known that the reduction in the amount of unreacted austenite is important to improve machinability of ADI⁽⁶⁶⁾⁻⁽⁶⁸⁾, because it is unstable and transforms to martensite at the front of the cutting edge through machining. Employment of in-mold in-

oculation or permanent mold is very efficient in greatly increasing the graphite nodule count. As a result the amount of unreacted austenite is reduced and the machinability of ADI is improved⁽⁶⁹⁾.

In the future ADI will be used more widely as there is an uncompromising requirement for reliability. Therefore, the efforts to produce high-grade ADI must be continued in the foundry shops.

V. Conclusion

Much attention of foundrymen have been focussed on producing lightweight, thin walled and stronger castings in recent years. ADI has great potentialities to meet with the industrial requirements because of its excellent combinations of mechanical properties. Regardless of the excellent potential offered by ADI, if the heat treater cannot process the casting to be free of embrittling, or if the foundryman cannot make the casting without chill or excessive intercellular carbides and phosphides, then the development of ADI will never be realized.

The reasons which have prevented a more generalized use of ADI lie mainly in the complexity of the interactions between the austempering treatment variables and the difficulties associated with the controlled production of parts of consistent quality. The first step of overcoming these difficulties will be the understanding of the kinetics and mechanism of austempering transformation. Only by combining an understanding of metallurgical principles with good control of processing, ADI can be used to provide reduction of overall costs and improvements in performance.

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