DEFORMATION AND FRACTURE OF CAST IRON WITH FERRITE MATRIX

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Micro–specimens of cast iron with ferrite matrix have been pulled with an incresmental strain intermittentely and observed correspondingly in situ by using an SEM with a tensile holder. The morphology change on surface as well as crack initiation and propagation have been observed. It is found that conventional theories on treating graphites as voids and their relative stress concentration are difficult to interpret the experimental results and still remain some problems in practise and theory on the mechanical behaviour of cast iorn. On which a new explanation has been made on the basis of experiments.

INTRODUCTION

The conventional cast iron mechanical–behaviour theories established on the basis of treating graphites as voids have been accepted for a long time. However, the graphites are crystals and distributed randomly, of which the interatomic force on basal plane is of strong covalent bond although that along C axis van der weals', and under tension the graphite actually goes through fracture process, such as, interfacial fracture or internal fracture, which dissipates quite a lot energy, as reported by the authors (1). Therefore, the initial graphites in cast iron couldn't be regarded as voids and the conventional theories should be explained anew. In this study, by choosing cast iron with ferrite matrix, through the investigation on micro–processes from deformation to fracture of cast iron, the mechanism of crack initiation and propagation and the roles played by graphites and matrix in cast iron have been discussed.

EXPERIMENTAL PROCEEDURES

Both grey and nodular iron with ferrite matrix have been cut to prepare a single notch micro–tensile specimen with surface microstructure revealed by using the same method presented elsewhere (1). The prepared specimens have been pulled with an incremental strain intermittentely, observed in situ and the corresponding surface morphology changes have been recorded on a series of photographs by using an SEM 35–CF with a tensile holder. The amount of increase in length ($\triangle l_i$) of a specimen after various pulling is measured by using a micro scale of tensile holder and then converted into elongations (δ_i). The corresponding micro strains $\varepsilon_{i\cdot,r}$ of a micro zone in various distances r ahead of

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notch tip have been measured on a serics of photographics observed in situ by measuring the length $l_{i \cdot r}$ between marked graphites distributed in a vertical line of the corresponding r and then converted into $\varepsilon_{i \cdot r}$. A plot of $\varepsilon_{i \cdot r}$ aganist r of various incremental strain δ_i and that of ε_i (mean value of $\varepsilon_{i \cdot r}$) / δ_i aganist δ_i have been established.

EXPERIMENTAL RESULTS

The interfacial crack initiation at graphite–matrix (G–M) interface in grey and nodular iron is shown in Fig.1. It can be seen that in both G–M interfacial cracks initiate at or before a very low critical strain $\delta_c(\leqslant 0.3\%)$ and that there are no slip bands or cracks on the matrix around most cracked graphites or uncracked ones, although some evidences on plastic deformation or crack on the matrix around a few cracked flake graphites, as pointed by narrows. With increase in $\triangle l_i$, in grey iron, cracks not further extend yet, but open and blunt as well as some new ones initiate in front of it, as shown in Fig.2(a), while in nodular iron, cracks propagate along G–M interfaces with gradually changing direction at first, followed by some new G–M interfacial cracks initiate and propagate in a same way in front of the primary one, as shown in Fig.2(b). Then, the matrix (ferrite) bridges between cracked nodular graphites slip locally and go further to fracture along the severe slipped region or which between some dense cracked graphites fracture along grain boundaries directly, leading to form a crack consisting of G–M interfacial cracks and matrix crack (Fig.2(c) and Fig.3(a) and (b)). We call the crack as cast iron crack (CIC). After CIC formation, with ircrease in $\triangle l_i$, the crack tip opens and blunts locally and obviously. This accompanys an intensive slip within ferrite matrix and the interfacial craking of some other G–M interfaces even secondary CIC nucleation in front of crack tip, as shown in Fig.3(c). Finally, both CICs open and blunt again respectively and link up by fracturing of matrix bridge with prior severe slip, and as a result, CIC gets a jumpy propagation. These processes carry on until a specimen finally fracture, as shown in Figs.3(a)–(c).

Fig.4 shows the strain distribution within a micro–zone in front of notch or crack tip after various tensile strain. It can be seen that the strain distribution in nodular iron is much more homogeneous than that in grey iron. But after CIC formation, strain also becomes localized arround crack tip in nodular iron.

DISCUSSIONS

Graphite–Matrix's Interfacial Fracture

It is conventionally believed that graphites in cast iron can be regarded as voids with no strength, leading to stress concentration around them. If this were true, graphites in cast iron would have no fracture process, or for which the energy would spend so little that it could be neglected, and the matrix around sharp tip of graphite would slip or fructure first, having maximum stress concentration. However, experimental results indicate that graphites in cast iron go through fracture process in themselves indeed, such as, the G–M interfacial fracture and internal fracture of graphite (Fig.1 and Fig.2(c) . The G–M interfacial fracture has also been observed by a number of investigators (Adewara and Loper (2) and Eldoky and Vogit (3) . In addition, although the first interfacial crack initiation occurs at a very low $\delta_c(\leqslant 0.3\%)$, the crack propagation along spherical graphite or inclined flake graphite interfaces requires increase in strain due to the toughening effect of crack deflection proposed by Oh et al (4). Moreover, the internal fracture on non–basal plane of strong covalent bond in graphite of cast iron takes place as well (Fig.2(a) and (c) . The above interfacial fracture and external fracture of graphite carry on until final fracture of a specimen, spending quite a lot energy and constituting a noticeable part in fracture process of cast iron. Thus, the initial graphites in cast iron can't be regarded as voids with no strength, and should not cause stress concentration of micro–notch. This is proved by experimental results that the G–M interfacial cracks in both grey and nodular iron initiate at or before a same δ_c(0.3%), although in which nodular graphite is much rounder and blunter than flake one, and that no any a slip band or crack appears on matrix adjacent to sharp tips of all graphites cracked not yet and of most cracked graphites. Althougth there are a few slip bands or cracks on the matrix around cracked graphite–tip as pointed by narrow in Figs. 1 and 2, they should take place after graphite cracking. In the light of above argument it should be accepted that the initial graphite in cast iron can't be regarded as void with no strength and that the G–M interfacial fracture goes through its crack initiation and propagation, as also described by Bradly and Srinivasan (5). The crack initiation at G–M interface is not dependent on graphite shape. Whether the graphite is flake or nodular one, their δ_c are identical ($\leqslant$ 0.3%). This is in agreement with that the crack initiation depends on the stress concentration σ_i of mismatch deformation between G and M and the weak bond of interface (1) besides supports no stress concentration resulting from graphite once again, since the σ_i and weak bond are also independent of graphite shape. However, after initiating, the interfacial crack propagates along G–M interface with a different path in grey iron and nodular one respectively. In grey iron, it propagates along a nearly linear (planar) interface between flake graphite and matrix fastly, while in nodular iron, along a spherical interface between nodular graphite and matrix with changeable–direction. With increase in the deflectional angle of spherical interface crack, as indicated by the authors (6), the opening stress σ_{op}, the stress intensity factor of model 1 k_1 and the effective stress intensity factor k_{eff} all decrease due to the deflection effect of crack propagation and as a result, the interfacial crack propagate in slow down even to stops. Only if applied load increases to satisfy the conditions: $\sigma_{op} \geqslant \sigma_{f\cdot i}$ and $k_1 \geqslant K_{1c\cdot i}$, where $\sigma_{f\cdot i}$ and $K_{1c\cdot i}$ are fracture strength and fracture toughness of interface respectively, can the crack continuously propagates along G–M interface. This imply that the G–M interfacial crack propagation in nodular iron requires much more energy than that in grey iron, thus nodular iron has a much higher fracture toughness.

Matrix Deformation and Cast Iron Crack Formation

Once interfacial crack initiates, the crack tip raises stress concentration. The crack extends, leading K_1 to increase. Therefore, in grey iron the interfacial crack can propagate spontaneously without to further increase in applied load. However, in nodular iron, on one hand, the interfacial crack extends, leading K_1 to increase also but on the other hand, during interfacial crack propagating, k_1 decreases due to deflection effect (6) and usually the latter plays an important role, requiring increase in applied load to continue the crack propagation. Once applied load increases up to a limit,i.e., $\sigma_{ap}(+\sigma_i) \geqslant \sigma_{y\cdot m}$ (yield strength of matrix), the matrix bridge between cracked graphites slips locally to relieve the stress concentration caused by the interfacial crack around nodular graphites. Meanwhile, because the silicon–alloyed ferrite matrix work hardens extensively upon plastic deformation, the deformation of matrix and the G–M interfacial crack may be taking place in other regions ahead by a same way as just mentioned above. As a result, the total strain can be distributed over a lot of local slip bands of matrix bridge between cracked graphites. This explains why the $\varepsilon_{i\cdot r}$ and ε_i $/ \delta_i$ distributed over r distance are more homogeneous in nodular iron than that in grey iron (Fig.4). With further increase in strain the matrix bridge between cracked graphites slips intensely and finally fractures along a grain boundary or along slip planes within the severely plastically deformed matrix to link up with G–M interfacial crack, and as a result, a crack consisting of G–M interfacial crack and matrix crack in cast iron forms. Only if this crack has formed, can the G–M interfacial crack propagate throughout its adjacent matrix and toward the final fracture of cast iron. For ease of presentation and to distinguish this crack from interfacial crack, we have called this crack as CIC in previous paragraph.

Cast Iron Crack Propagation

After forming , with strain increase, CIC in grey iron opens up and becomes blunt with an acompaniment of a small–scale plastic zone at crack tip rather than further ex–

tends. Nevertheless, a new CIC maybe forms in front of the primary one without matrix plastic deformation in evidence except the small–scale plastic zone. Then, these processes carry on further in the same way, followey by that the primary and new CIC link up by fast fracture of matrix bridge between them, and specimen final fractures at once due to the crack length a to approch to a critical value $a_{c \cdot g}$, i.e., $a > a_{c \cdot g}$. On account of this, it is explained why in grey iron strain severely concentrates at crack tip and why matrix has no slip bands basically, as shown in Fig.4.a and Fig.1 respectively. In nodular iron, however, after forming , with strain increase CIC opens and becomes blunt much more than that in grey iron. Besides, this accompanys with an intensive plastic deformation of matrix in front of cast iron crack tip and reformation of some new out–of–plane G–M interfacial cracks even CICs, as shown in Figs.3(c)–(e). Then, the primary and new CICs propagate respectively. Meanwhile, two kinds of matrix bridges, internodular bridge and bridge between CICs, slip intensely and finally fracture to link up with G–M interfacial cracks or CICs to take a leap in the CIC propagation. Afterwards, CIC propagates in the same way that mentioned above until it finally fractures. These processes mentioned above carry on locally around crack tip, leading to the $\varepsilon_l \;/ \delta_l$ to increase intensely after ICC formation, as shown in Fig.4.b. However, the large amount of strain, especially crack bridging and deflection effect (4) dissipate a major energy, leading to considerable increase in fracture toughness of nodular iron.

CONCLUSIONS

Under tension cast iron goes through initiation and propagation of interfacial crack and that of cast iron crack, and their relevant deformation and fracture in matrix.

Crack initiates first at the interface between graphite and matrix, which depends on the weak bond in the interface or in its subsurface graphite and on the internal stress concentration arising from mismatch deformation around interface between graphite and matrix rather than on the graphite shape. Nevertheless, on which the crack propagation depends. Interfacial crack of nodular graphite and matrix propagates along a changeable direction and so in a growth rate slow down.

During or after interfacial crack propagation, matrix bridge between cracked graphites in nodular iron goes througth slip, severe slip and fracture to connect the matrix crack with interfacial crack, and to form a cast iron crack.

After forming, the cast iron crack propagtes through its opening and blunting, new interfacial crack formation as well as local slip even fracture of matrix to form a secondary CIC ahead, and the fracture of matrix between CICs, leading to a final fracture.

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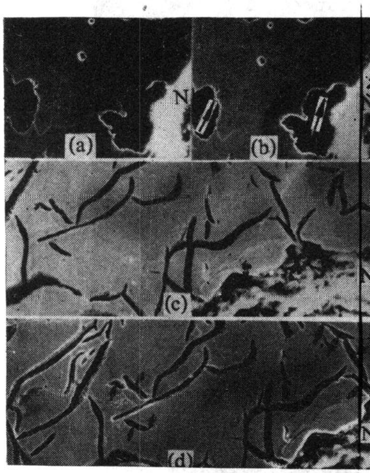
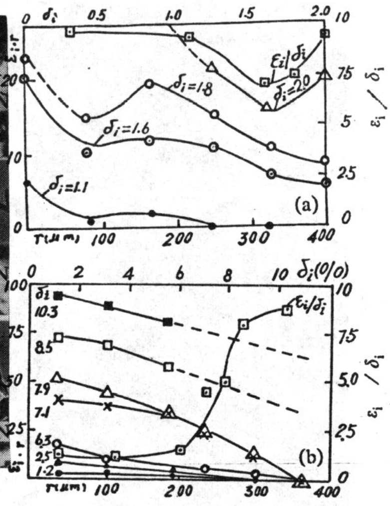


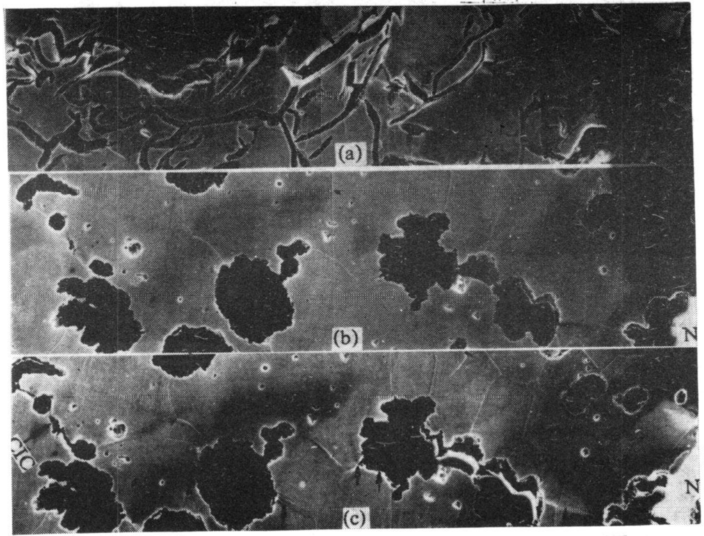
(a),(c) $\delta_i = 0$ (b),(d) $\delta_i = 0.3\%$ 10.0U
↑ = Crack or slip pointer.
N = notch.

Fig.1 Interfacial crack initiation observed in situ.



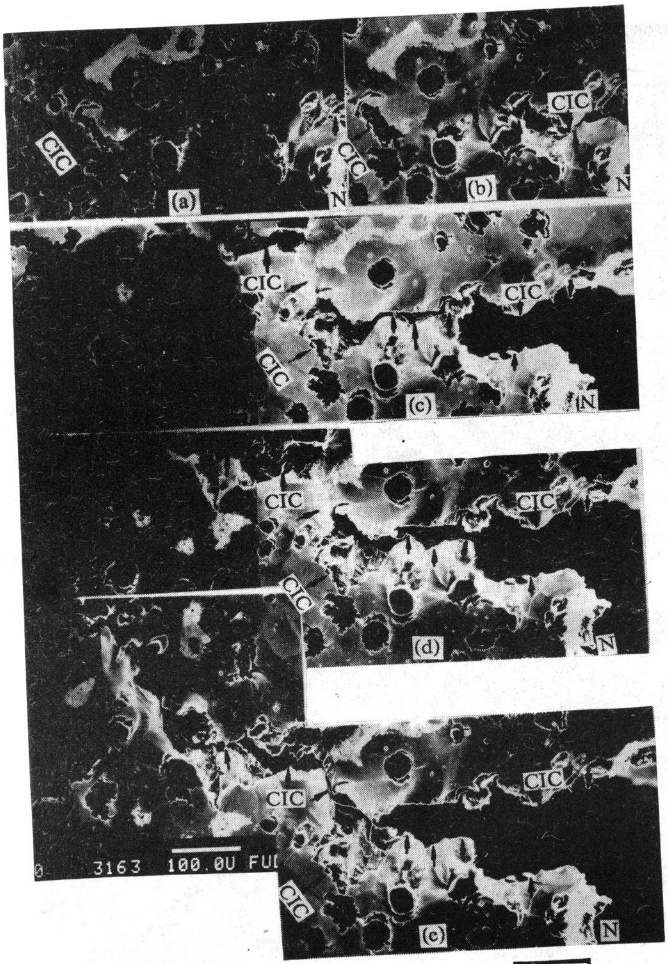
(a) In grey iron. (b) In nodular iron.

Fig.4 A plot of $\varepsilon_{i \cdot r}$ and ε_i / δ_i against r and δ_i respectively.



(a) $\delta_i = 1.0\%$ (b) $\delta_i = 0.5\%$ (c) $\delta_i = 2.8\%$ 10.0U
↑ = Slip band or crack pointer. N = notch.

Fig.2 Interfacial crack propagation observed in situ.



(a) $\delta_i = 6.8\%$ (b) $\delta_i = 7.2\%$ (c) $\delta_i = 8.5\%$
(d) $\delta_i = 9.2\%$ (e) $\delta_i = 10.3\%$
↑ = Pointer of matrix bridge between cracks or the remains of bridge fracturing.

$\underset{CIC}{\uparrow}$ = Pointer of CIC.

N = notch.

Fig.3 Cast iron crack (CIC) formation and propagation observed in situ.