DEFORMATION AND FRACTURE OF NODULAR IRON WITH FERRITE MATRIX

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ABSTRACT

Micro-specimens of nodular iron with ferrite matrix have been extended and observed in situ with an SEM. In combination with using corner technique, surface microstructure, cracks and related fractography have been observed. It is found that the conventional theories of cast iron on treating graphites as voids do not agree with reality. A new explanation of the mechanical behaviour of nodular iron has been made and the reason for improvement mechanical properties by graphite nodulizing has been clarified.

KEYWORDS

Nodular iron, ferrite matrix, deformation and fracture, interfacial fracture, interfacial mechanical behaviour.

INTRODUCTION

The invention of nodular iron was an important jump in developing processes of cast iron. Nodular iron not only has the general merits of grey iron but also strength and ductility competitive with those of steels. It is conventionally believed that this is due to considerably lowered stress concentration at graphite in the cast iron. However, graphite particles are crystals distributed randomly in cast iron , of which the interatomic forces on the basal plane are strong covalent bonds although that along C axis is van der Waals, and, in fact, in cast iron under tension graphite goes through deformation and fracture processes, which keep on till the moment just before final fracture of specimen and so use up quite a lot of the work done by applied load and deformation (He et. al., 1991a). Therefore, graphites in cast iron can not be treated as voids. Since the initial graphite particles are not voids, they should not be able to cause notch stress concentration. This can also be supported by cracks initiating first at graphite (G) – matrix (m) interfaces (Eldoky and Voigt, 1977; He et. al., 1991a,b), but not in the matrix surrounded graphite tip (He et. al., 1991b). From the above, it can be seen that there are some contradictions between the conventional cast iron theory related to graphite and the relevant practical mechanical behaviour. So the theory should be explained anew. For this reason, in this investigation, by choosing nodular iron with ferrite matrix, the micromechanism of deformation and fracture in nodular iron and the nature of mechanical property improvement by graphite nodulizing have been studied.

EXPERIMENTAL PROCEDURES

Nodular iron with ferrite matrix was cut into 30 × 2 × 0.5mm specimen blanks with a single notch having a radius 0.2mm and depth 0.5mm at the middle of a specimen by using a spark machine. The blanks were ground, polished and etched to reveal the microstructure and form micro tensile specimens. The specimens have been pulled intermittently, observed in situ and surface morphology photographed by using an SEM 35–CF with a tensile holder. The increase in length (Δl_1) corresponding to each intermittent point was measured by using a micro scale of tensile holder and converted into elongation δ_1. After fractured, specimens have been tilted about 45° under an SEM to observe both side surfaces by using the corner technique. According to the results of surface morphology change on one side surface of specimen observed in situ, the key field on fracture at the other side of specimen could be chosen for further detailed observation. By using the same technique, their matching fracture surfaces have also been observed. Thus, micro processes on deformation and fracture in nodular iron have been investigated comprehensively.

EXPERIMENTAL RESULTS

The initial morphology of a tensile specimen is shown in Fig.1.a. As pulled to increase in length 10μm (δ_1 =0.03%), the specimen was observed in situ and there were no slip bands observed in the matrix. However, it has been seen that at graphite–matrix (G–m) interface cracks nucleated already, as shown in Fig.1.b. With increase in Δl_1, the cracks propagate along the G–m interface with gradually changing direction to form an interfacial crack. At the same time some other new G–m interfacial cracks form in the same way around nodular graphites in front of the primary interfacial crack as shown in Fig.2.a. Then, local slip takes place in the matrix (ferrite) between cracked nodular graphites or adjacent to the notch of specimen and the slip mostly takes a favoured direction nearly 45° to the tensile axis, as shown in Fig.2.b. and c. Afterword, these processes keep on for quite a long while till δ_1 = 0.72% the locally slipped matrix between cracked nodular graphites or adjacent to notch fractures to link up with interfacial cracks and form a crack at least consisting of a G–m interfacial crack and a transgranular crack of matrix, as shown in Fig.3.a. For the sake of difference to interfacial cracks and convenience in discussion, the crack is called here a cast iron crack (CIC). With increase in Δl_1, the tip of CIC opens. Meanwhile, ferrite matrix in front of it slips intensively. After quite a long elongation it fractures rapidly along slip planes to connect with the G–m interfacial crack in front of it, and as a result, crack propagation takes a jump, as shown in Fig.3. b and c. Thus, the propagation of CIC is intermittent with some time spent on incubation and some time devoted to propagation.

The fractography corresponding to structure and crack observed on the surface and its matching fractography are shown in Fig.4. It can be seen here that most parts of fracture are cracked along the G–m interface and appeared as some nodular graphite particles protuberant and some their dimples depressed, that a few parts are mixed trangranular and interfacial fracture of a few graphits and so their fractographies appear as a G–m interfacial crack ring around a trangranular fracture surface of nodular graphite on a match fractographs and that another a few parts are ferrite matrix fracture with dimples coalescence.

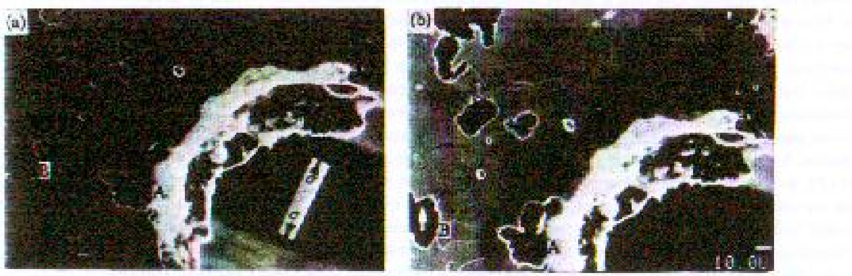

Fig.1. Initial microstructure and crack initiation at G–m interface in nodular iron (a) initial microstructure (b) in situ observation of (a) specimen after pulled (σ_1 = 0.03%) arrows point to crack initiation at G–m interface and its adjacent matrix with no slip band.

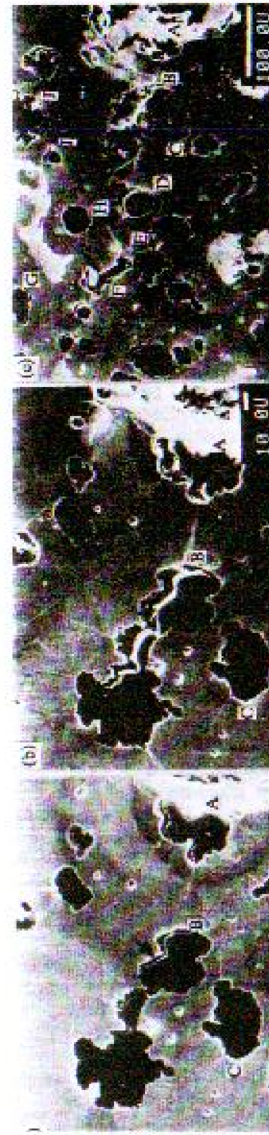

Fig. 2 In situ observation of Fig.1 specimen after various strains arrows point to crack or slip band and A, B, ···, J mark graphites, the same in Fig.3 and 4 (a) δ_1 = 0.05%, cracks propagate along G–m interfaces with changeable direction (b) δ_1 = 0.28%, cracks propagate further and slip on matrix between the cracks (c) δ_1 = 0.68%, the matrix slips intensely and new interfacial crack initiates ahead

Fig. 3 In situ observation of Fig.2 specimen after further strain (a) δ_1 = 0.72%, CIC formation by fracture of intensely slipped matrix to link an interfacial crack with another one or micro–notch (b) δ_1 = 0.85%, primary CIC crack tip opening and secondary CIC formation in front of it (c) δ_1 = 1.03%, primary and secondary CIC link up by fracture of intensely slipped matrix between them.

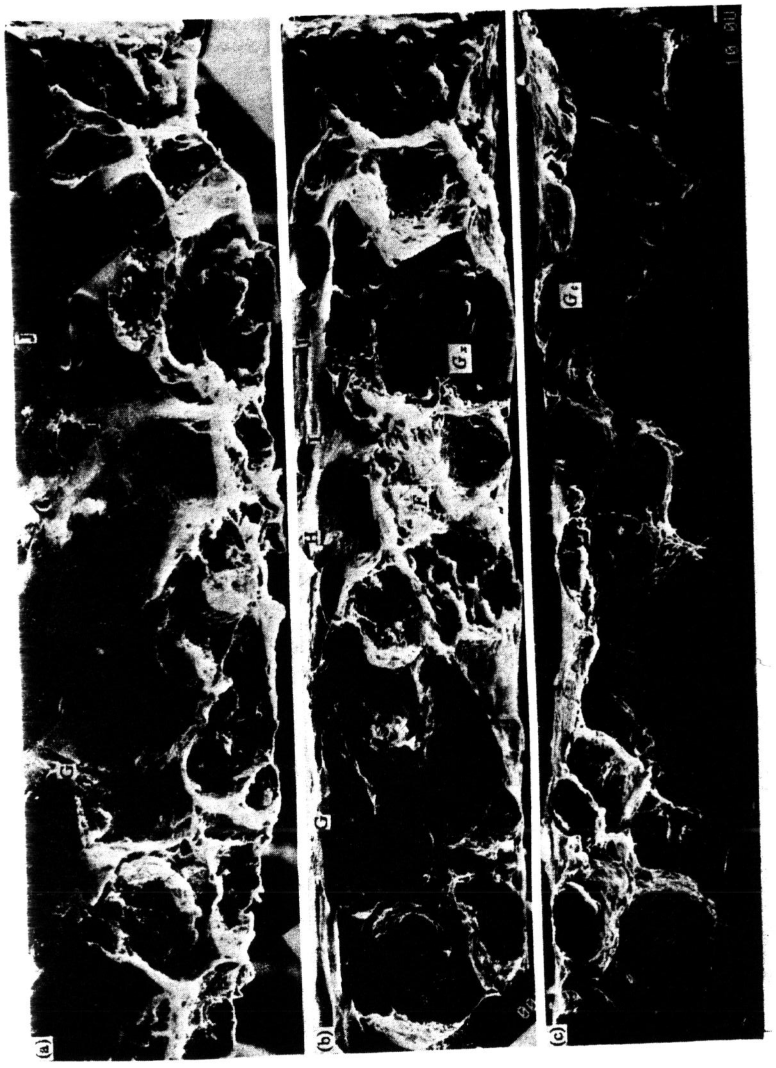

Fig.4 Matched fractographs and their relevant surface morphologies of nodular iron (a) fracture surface and its relevant morphology observed by corner technique (b) and (c) corresponding matching fracture surfaces G_c — graphite dimples depressed after interfacial fracture G_x — graphite particles protuberant after interfacial fracture F_d — dimple coalescence on ferrite fracture

DISCUSSIONS

Relation Between Nodular Graphite and Crack Nucleation in Nodular Iron. For a long time, it was considered that graphites in cast iron can be simply regarded as voids with no strength. Thus, it was always believed that nodular iron has good strength and ductility because graphite nodulizing decreases notch stress concentrations caused by graphite in cast iron. If the graphite were void and caused stress concentration, the maximium stress concentration $\sigma_{c,max}$ would take place in the metrix just adjacent to a sharp graphite tip, and so would cause first deformation or fracture. However, in fact, cracks always nucleate first at the G–m interface perpendicular to the tensile axis in cast iron (He et. al., 1991a.b). What is more, the critical strains for interfacial crack nucleation are the same for both grey and nodular irons ($\delta_i = 0.03\%$, under the same testing conditions), as shown in Fig.1, although in which nodular graphites are much rounder and blunter than flake ones. These evidences and illustrations prove that there is no relation between crack nucleation and the shape of initial graphite in cast iron. The only explanation is that the initial graphite particles are not voids and so do not cause notch stress concentration.

Since crack initiation at the G–m interface is not due to the graphite notch stress concentration, it is mostly due to the weak van der Waals' force and the short–range stress concentration caused by the incompatible deformation between different phases on both sides of the interface. To the former, according to graphite structure it is understandable that graphite crystal growth must rely on the strong covalent bond of atoms in a series of sheets parallel to basal plane to catch carbon atoms from solution in front of graphite and extend tangently along the sheets. Therefore the external layer in graphite must consist of the sheets parallel to basal plane and this has been proved by electron diffraction (Zhang Buo et. al., 1988). Certainly, the normal of graphite suface layer must be parallel to the C axis of graphite along which bonding force between the sheets is van der Waals', and the structure must accordingly be regarded as a molecular one. From this it can be deduced that the structure of G–m interface should be a transition–structure with weak bonds from external surface graphite structure to matrix structure. This is one reason why the crack initiates preferentially at the G–m interface or its subsurface graphite perpendicular to tensile axis in cast iron in tension. To the latter, because both sides of the G–m interface are respectively graphite and matrix with different phases, they certainly manifest themselves different mechanical behaviour under loading and cause a short–range internal stress distributed near the interface. According to Fig.1.b, G–m interfacial crack initiates before slip of matrix adjacent to it so that the mechnical behaviour near interface can be treated as elastic one. For convenience of analysis, first let's remove graphite from matrix and then they are able to freely deform elastically with different amount strain ε_g and ε_m respectively under loading due to each different elastie modulus E_g and E_m. Because $E_g < E_m$, therefore $\varepsilon_g > \varepsilon_m$ under equal applied load. Then let's replace the graphite into its original location in matrix, they will be deformed to match each other and form a continuous body. At this moment, the elastic deformation of the matrix is very small and can be neglected due to its strong strength and large size, compared to that of graphite. Thus, as shown in Fig.5, the graphite surface carries an applied tensile stress and short–range internal stress σ_i arising from incompatable deformation between graphite and matrix

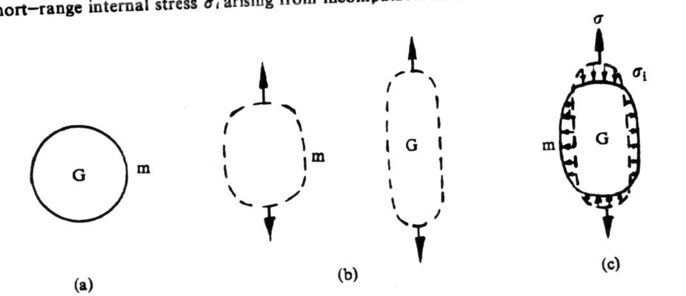

Fig.5. Schematic drawing of incompatible deformation and internal stress between G and m. (a) initial state (b) free deformation of G and m (c) match deformation and internal stress σ_i (arrows indicate deformation direction of G)

beside. The σ_g is compressed at two poles of nodular graphite and tensile at the end of its equator. Nevertheless, at matrix adjacent to where σ_s is of apposite direction, Thus, the result tensile stress acting on graphite and matrix is accordingly $\sigma_{l\cdot g}$ and $\sigma_{l\cdot m}$ and have been deduced as (He et. al., 1991a)

$$\sigma_{l\cdot g} = \sigma_{op} \cdot E_g / E_m \tag{1}$$

$$\sigma_{l\cdot m} = \sigma_{op} \cdot (2 - E_g / E_m) \tag{2}$$

when $\sigma_{l\cdot g} > \sigma_{f\cdot g}$ (fracture strength of graphite), i.e.,

$$\sigma_{op} > E_g / E_m \cdot \sigma_{f\cdot g} \tag{3}$$

a crack will be able to nucleate at poles of nodular graphite, while as $\sigma_{l\cdot m} > \sigma_{f\cdot i}$ (fracture strength of interface), i.e.,

$$\sigma_{op} > \sigma_{f\cdot i} / (2 - E_g / E_m) \tag{4}$$

at the G–m interface on the poles of nodular graphite. However, to realize crack nucleation, only when the $\sigma_{l\cdot g}$ or $\sigma_{l\cdot m}$ is larger than $\sigma_{y\cdot m}$ (yield strength of matrix) to make matrix yield a concessive deformation can the crack be opened. According to this, the following equations have been deduced respectively (He et. al., 1992).

For cracking at graphite $\quad \sigma_{op} > \sigma_{y\cdot m} \cdot E_g / E_m \quad (5)$

For cracking at interface $\quad \sigma_{op} > \sigma_{y\cdot m} / (2 - E_g / E_m) \quad (6)$

Formulas (3), (5) and (4), (6) are the essential and full conditions for interfacial crack initiation in cast iron. They indicate that critical stress for crack initiation depends on fracture strength of interface and of graphite, elastic modulas of graphite and of matrix, and the yield strength of matrix, i.e., the lower the strength and the larger the difference between E_g and E_m, the esier the crack initiation. According to the order to satisfy formulas (3, 5) and (4, 6), it is decided whether crack nucleates first at graphite or at G–m interface. However, they also show that crack initiation not relates to the shape of graphite. That is the reason why critical strain δ_i for crack initiation of grey iron and that of nodular iron both are the same(= 0.03%, in this investigation).

Nodular Graphite Relates to G–m Interfacial Crack Propagation. After nucleated, the G–m interfacial crack is difficult to propagate throughout its both sides, because on one side there are graphite sheets parallel to basal plane with covalent bond and on the other side there is matrix. They both are much stronger than G–m interface. Thus, it is only a way for the crack propagation to propagate still along G–m interface. That is why a lot of interfacial cracks propagate around nodular graphites. Because crack propagation in such a way gradually changes crack direction, the opening stress σ_{op}, stress intensity factors of model I and II, i.e., K_I and K_{II} and the effective stress intensity factor K_{eff} should all be changed with the deflectional angle increase. In reference to the critical normal stress law, cleavage occurs when the component of the stress normal to the cleavage plane (σ_n) reaches a critical value (Schmid and Boas, 1950), and the σ_n can be written as

$$\sigma_{op} = \sigma_n = \sigma_{ap} sin^2(90 - \theta). \tag{7}$$

If regarded the nodular G–m interfacial crack as the envelope of a series of tangent lines being a model I crack nuckeating at poles of nodular graphite and a series of further propagated cracks l_i with deflection angles θ_i, $i = 1,2,3,\cdots,n$, decline to its prior and neglected the small difference among θ_i, i.e., let $\theta_i = \theta$, thus, the model I, II and effective stress intensity factors of i–th deflections $k_{I\cdot i}$, $k_{II\cdot i}$ and $k_{eff\cdot i}$ can be respectively obtained from following equations (He et. al., 1992):

$$k_{I\cdot i} = cos^3(\theta/2)k_{I\cdot i-1} - 3sin(\theta/2)cos^2(\theta/2)k_{II\cdot i-1} \tag{8.a}$$

$$k_{II\cdot i} = sin(\theta/2)cos^2(\theta/2)k_{I\cdot i-1} + cos(\theta/2)[1 - 3sin^2(\theta/2)]k_{II\cdot i-1} \tag{8.b}$$

$$k_{eff\cdot i} = k^2_{I\cdot i} + k^2_{II\cdot i})^{1/2} \tag{8.c}$$

Only as $i = 1$, equation (8) is exceptioned and should be

$$k_{I\cdot i} = cos^3(\theta/2)K_I \tag{9.a}$$

$$k_{II\cdot i} = sin(\theta/2)cos^2(\theta/2)K_I \tag{9.b}$$

From the above equations (8) and (9), It can be seen that when G–m interfacial crack propagating, whether the opening stress of crack tip σ_{op}, intensity factor of opening stress $k_{I\cdot i}$ or effective stress intensity factor $k_{eff\cdot i}$ all decrease with increase in deflection angle θ, and so crack growth rate should be certaicly slowed down. Once they go down to $\sigma_{op} < \sigma_{f\cdot i}$ or $k_{eff\cdot i} < K_{Ic\cdot i}$ (fracture toughness of interface), crack

stops propagating. Only if k_{eff} and σ_{op} increased by further increasing in load to satisfy

$$k_{eff\cdot i} \geqslant K_{Ic\cdot i}; \quad \sigma_{n\cdot i} \geqslant \sigma_{f\cdot i} \tag{10}$$

can the interfacial crack be further propagated around nodular graphite. This implies that the cast iron is both strengthened and toughened.

Plastic Deformation of Ferite in Nodalar Iron. Although ferrite has very good ductility, the ferrite matrix in grey iron becomes brittle. In nodular iron, however, it is still ductile. This has been explained by stress concentrations arising from graphite in iron. However, as pointed out above, initial graphites in cast iron can not be treated as voids and do not cause notch stress concentrations in the initial stages of deformation. Thus, the conventional arguments on ductiliy of ferrite in cast iron should be discussed again. First, as mentioned above, although it is not related to crack nucleation, graphite shape has a strong effect on propagation of interfacial cracks. In the case of flake graphite in grey iron, after nucleation, the G–m interfacial crack propagates along the G–m interfacial to form a linear interfacial crack a_g in a length of flake graphite quickly and its adjacent matrix still has no plastic deformation in evidence. While in nodular graphite the G–m crack propagates with changeable direction along G–m interface of nodular graphite. Consequently, the effect of crack deflection leads to a decrease in the opening stress and effective stress intensity factor. Thus, the crack growth rate goes down and may even stop gradually. Only if the applied load is further increased enough to compensate for the deflection effect, can the interfacial crack further propagate. Once the applied load approches $\sigma_{ap} > \sigma_{y\cdot m}$, the ferrite can slip as shown in Fig.2 and 3, and the above mentioned processes still keep on to a moment just before final fracture. Next, if the volume fraction of graphite is kept constant, the G–m interfacial area of nodular graphite is a minimum in cast iron, i.e., the areas of G–m interfacial crack or that to cut matrix is minimum, and the equivalent model I linear crack length a_s of interfacial crack in terms of fracture mechanics is equal to or less than nodular graphite diameter d_s, i.e., $a_s \leqslant d_s$. While that of flake graphite a_g is equal to its own length. Obviously, $a_g \gg a_s$. In addtion, after interfacial fracture, graphite in cast iron can be regarded as voids, and so, as pointed in conventional theory, nodular graphite causes stress concentrations much less than flake graphite, It will also lead to the fracture toughness of nodular iron $K_{Ic\cdot s}$ being much larger than that of grey iron $K_{Ic\cdot g}$, i.e. $K_{Ic\cdot s} \gg K_{Ic\cdot g}$. Supposing they are stressed under a same applied load, say $\sigma_{ap} = \sigma_{y\cdot m}$, the critical crack lengths of nodular iron and grey iron $a_{s\cdot c}$ and $a_{g\cdot c}$ can be caculated by substituting $K_{Ic\cdot s}$ and $K_{Ic\cdot g}$ into K_{Ic} respectively in the following equation

$$a_c = (K_{Ic} / Y\sigma)^2 \tag{11}$$

where Y is a coefficient. Obviously, $a_{s\cdot c} \gg a_{g\cdot c}$. In terms of the above description, $a_g \gg a_s$. and $a_{s\cdot c} \gg a_{g\cdot c}$, thus, under loading, in grey iron the G–m interfacial crack after nucleated can immediately grow to the length a_g due to no crack deflection effects. If $\triangle l_i$ only increases in a little, could the crack length equal to $a_g + \triangle l_i$ be larger than $a_{g\cdot c}$. Consequently ferrites fracture in britlleness at once under a low load. However, nodular iron is quite different. If the nucleated G–m interfacial crack propagate continuously along G–m nodular interface, its direction will change gradually, and so the effective opening stress as well as effective stess intensity factor decrease, and it slows down even prevents crack propagation. If only applied load increases to high enough to compensate the decrease in that, can the interfacial crack propagate continuously. Once the applied load increases up to or over $\sigma_{y\cdot f}$, ferrite will slip, as shown in Fig. 2. and 3. These processes keep on till the interfacial crack propagates around a whole nodular graphite. What is more, even the processes on deformation and fracture are going to a moment just before final fracture, the crack length is still less then $a_{s\cdot c}$. Therefore, the ferrite matrix in nodular iron is able to yield and continue flow till final fracture. This brings the ability of plastic deformation and work hardening of ferrite into full play to the ductility and to increase in strength in cast iron, and that is another true reason for nodular iron strengthening and toughening.

Nucleation and Propagation of Cast Iron Crack. As described above, the crack always initiates at G–m interfaces and propagates to form an interfacial crack in cast iron. However, the interfacial crack, only if after passing through its adjacent matrix to form a cast iron crack, can further propagate to lead to a final fracture, as shown in Fig.3. The formation of CIC in nodular iron is a process, in which, after nucleated, interfacial cracks propagate around nodular graphite with increasing applied load, as shown in Fig. 2 and 3. Meamwhile, some other interfacial cracks may nucleate and propagate at some graphite particles adjacent to the cracked graphite particles as well if equations (4, 6) (7) are satisfied with increase in stress. When the applied stress is larger than $\sigma_{y\cdot f}$, between cracked graphite particles the ferrite slips locally along a favourable direction, and finally, if σ_{ep} is larger than $\sigma_{f.F}$ of the slipped ferrite, the adjacent interfacial cracks link up and convert into a cast iron crack by fast shearing or coalescence of voids along the slipped planes, as shown in Fig. 2.c and 3.a. Afterwards, with increase in load crack tip opening and blunting accompany ferrite deforming around crack tip. Meanwhile, in front of it, many interfacial cracks and even secondary cast iron cracks may be nuclcate. The cracks connect with the main crack by fast shearing or void coalesecence along slip planes of ligaments between cracks leading to a cast iron crack propagates forward in a big jump, as shown in Fig.3.b and c. Crack propagation in this way keeps on till final fracture.

CONCLUSIONS

Different mechanical behaviour in both sides near interface between G and m leads to interfacial stress concentration $\sigma_{I.c}$ arising from their incompatible deformation. The $\sigma_{I.c}$ and weak bond of G–m interface cause crack to initiate first at the interface rather than the notch stress concentration arising from graphite regardedd as void causes that. Therefore, graphite nodulizing does not delay or affect crack initiation.

Graphite nodulizing affects interfacial crack propagation to be changeable in direction and slowed down due to the effect of G–m interfacial crack deflection. After graphite cracked, ferrite matrix in cast iron can be fully plastically deformed and fully brought into play in strain strengthening due to interfacial crack deflection effect and inerease in fracture toughness to avoid britlle fracture owing to decrease in area to cut matrix and stress concentration and increase in bonding energy between fracture surfaces by graphite nodulizing.

Nodular iron fracture consists of G–m interfaial cracking, cast iron crack formation by connecting matrix crack with G–m interfacial crack and crack propagation in such a way that primary cast iron crack to connect with secondary one in front of it and cast iron crack to propagate once by leap and bound. It carrys on till final fracture in nodular iron.

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