Crack Nucleation and Propagation Mechanisms in Low-Temperature Impact Fracture of Ferritic Ductile Iron (Postprint)

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Abstract

Charpy notched impact tests were conducted on QT400-18L ferritic ductile iron castings at low temperatures to investigate the effect of temperature on crack nucleation and crack propagation capability during impact. OM was utilized to observe and analyze the crack initiation and propagation paths as well as the microstructural evolution near the fracture surface at different temperatures. The results show that above the ductile-to-brittle transition temperature, numerous graphite-matrix interfaces near the fracture surface cracked after impact, and the voids generated by graphite interface cracking played a role in blunting the crack and reducing the crack propagation rate; in the ductile-to-brittle transition temperature range, the impact specimens exhibited mixed dimple and cleavage fracture morphology, and both the fracture mode and crack nucleation were related to graphite nodules; below the ductile-to-brittle transition temperature, vertical intersecting twins nucleated and subsequently led to microcrack propagation, cleavage fracture was primarily initiated by twins, and this crack nucleation and propagation mode induced by deformation twins caused a dramatic decrease in both crack initiation work and crack propagation work.

Full Text

Preamble

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Mechanism of Crack Nucleation and Propagation in Ferritic Ductile Iron During Low-Temperature Impact Fracture

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Abstract

Due to its excellent ductility and moderate strength, QT400-18L ferritic ductile iron has been widely used for producing core components of wind power equipment such as wind turbine hubs. Most previous research has focused on exploring mechanical properties at low temperatures, but none have explained the microscopic mechanism of ductile iron during low-temperature impact, and the mechanism of crack nucleation and propagation during impact fracture has not been analyzed. In this work, the impact toughness of QT400-18L ferritic ductile iron was measured through V-notch Charpy impact tests at different temperatures, and the influence of temperature on low-temperature impact toughness and fracture behavior was discussed. The results show that the cleavage fracture resistance of QT400-18L ferritic ductile iron decreases with decreasing impact temperature. Above the ductile-brittle transition temperature (DBTT), most of the total fracture energy is expended during the crack propagation process. Below the DBTT, both crack initiation energy and crack propagation energy decrease significantly. Using in-situ fracture metallographic observation methods, crack initiation and propagation in QT400-18L ferritic ductile iron at different temperatures were analyzed. Above the DBTT, graphite nodules play a role in blunting cracks and reducing crack propagation rate. In the DBTT range, the fracture morphology shows mixed fracture features with cleavage and dimples, both related to graphite nodules. Below the DBTT, deformation twins lead to microcrack nucleation and result in cleavage fracture. Deformation twinning could play a significant role in the ductile-to-brittle transition of QT400-18L ferritic ductile iron.

Keywords: graphite nodule, ductile-brittle transition, cleavage fracture, deformation twin

Introduction

In recent years, the demand for large wind turbine components such as hubs and bases made of ductile iron has increased dramatically. These castings operate in low-temperature environments and have extremely strict requirements for low-temperature impact performance. QT400-18L ductile iron is commonly used for wind power accessories. It possesses good toughness and moderate strength, making brittle fracture unlikely under normal temperature conditions. However, at low temperatures, its mechanical properties change significantly: strength increases while ductility and toughness decrease, making brittle fracture more likely [1-4]. Currently, numerous studies have investigated the fracture failure behavior of ductile iron. Wei et al. [5] and Dai et al. [6] studied crack nucleation and propagation in wear-resistant bainitic ductile iron, finding that cracks primarily nucleate at graphite-matrix interfaces, particularly at sharp corners of non-spherical graphite, and also at inclusion-matrix interfaces. Dai et al. [7,8] used scanning electron microscopy (SEM) and transmission electron microscopy (TEM) with dynamic tensile testing to observe in-situ the micro-fracture process of austempered ductile iron, studying its microstructure and fracture mechanisms. They analyzed graphite behavior during fracture and the influence of matrix microstructure on crack initiation and propagation, discovering that voids formed by graphite-matrix interface decohesion could blunt the main crack and reduce crack propagation rate. However, systematic studies on the low-temperature impact fracture behavior of ferritic ductile iron are scarce, and the evolution of microscopic fracture mechanisms with temperature has not been investigated in relation to the ductile-brittle transition characteristics of ductile iron. Therefore, this work conducted impact tests on V-notched QT400-18L ductile iron specimens at various temperatures, analyzed impact fracture morphologies using SEM, and studied crack initiation, propagation, and cleavage crack formation from both macroscopic and microscopic perspectives, with particular emphasis on clarifying the cleavage fracture behavior during impact below the ductile-brittle transition temperature to elucidate the low-temperature impact fracture mechanism of ductile iron.

1. Experimental Methods

The raw materials consisted of low-silicon pig iron and scrap steel, melted in a 50 kg SL1400 medium-frequency induction furnace. Spheroidization was performed using the sandwich method, with inoculant covering the spheroidizing agent and 75SiFe used for floating silicon secondary inoculation to obtain qualified QT400-18L ductile iron with the following chemical composition (mass fraction, %): C 3.82, Si 1.92, Mn 0.08, P 0.03, S 0.02, and Fe balance. V-notch specimens were taken from the castings and subjected to Charpy impact tests at different temperatures according to ISO 148-1-2009 standards. Impact tests were conducted on an MTS instrumented impact testing machine, recording the absorbed energy curves during impact. Fracture morphologies were observed using an S-3400N SEM. To directly observe and analyze the deformation of graphite nodules and surrounding matrix throughout the entire impact fracture process at different temperatures, including crack initiation and propagation, the following method was employed: First, the two side surfaces of the V-notch impact specimens were coarse-ground, fine-ground, polished, and etched. Typical regions near the V-notch were selected under a GX51 optical microscope (OM). Charpy V-notch impact tests were then performed at temperatures ranging from 20°C to -80°C. Finally, the microstructure of the same region before and after impact was obtained by combining the characteristic features of surrounding graphite nodules, enabling systematic and direct analysis of crack initiation, propagation, and instantaneous fracture processes during room- and low-temperature impact of ductile iron.

2.1 Influence of Temperature on Impact Properties of Ferritic Ductile Iron

[Figure 1: see original paper] shows the load and absorbed energy curves of ferritic ductile iron during impact as a function of displacement at different temperatures. At -20°C, the total absorbed energy for ductile fracture was 13.23 J, exhibiting a ductile fracture mode. At -45°C, the total absorbed energy decreased to 9.79 J, placing the material in the ductile-brittle transition stage. When the temperature was further reduced to -80°C, the absorbed energy was only 4.18 J, showing cleavage fracture behavior. In Figure 1b, points Fm and Fy represent the maximum dynamic load and yield load, respectively. The area under the curve to the left of Fm represents the crack initiation energy (Ei), while the remaining area represents the crack propagation energy (Ep) [9,10]. Both Ei and Ep decrease with decreasing temperature, indicating that crack nucleation and propagation become easier at lower temperatures [11]. At -20°C, the crack initiation energy accounts for approximately 28% of the total energy, with a maximum impact load of 11.38 kN. At -45°C, the crack initiation energy increases to about 33% of the total energy. However, at -80°C, this proportion rises to approximately 40%, meaning that although the crack initiation process consumes less energy than propagation, the proportion of crack initiation work increases as temperature decreases. The a-b segment in Figure 1b is nearly vertical to the horizontal axis, indicating that after brittle fracture occurs, cracks propagate rapidly with almost no resistance from the specimen. Clearly, temperature significantly affects both crack nucleation and propagation capabilities: above the ductile-brittle transition temperature, crack initiation energy changes little while crack propagation energy decreases; below the transition temperature, both crack initiation and propagation energies drop dramatically [12,13].

2.2 Crack Initiation, Propagation, and Fracture Morphology at Different Temperatures

[Figure 2: see original paper] shows the fracture paths of ferritic ductile iron after impact at different temperatures, revealing significant differences. At -

20°C, cracks propagate in a zigzag pattern; at -45°C, the zigzag undulations decrease; and as the temperature drops to -80°C, the crack propagation path becomes relatively flat.

2.2.1 Crack Initiation and Propagation at -20°C

[Figure 3: see original paper] shows the impact fracture morphology of ferritic ductile iron at -20°C. The V-notch is located at the top of the image. Micro-voids are observed around graphite nodules, formed by decohesion at the graphite-ferrite interface or by localized matrix deformation. After micro-voids form at the graphite-matrix interface, cracks extend into ferrite grains along the strain direction, suggesting that crack sources are micro-voids formed around graphite nodules near the V-notch.

[Figure 4: see original paper] shows the morphology near the fracture of ferritic ductile iron before and after impact at -20°C. Figure 4b reveals the fracture path after impact: after complete decohesion at the graphite-matrix interface, a void shaped like the graphite nodule forms in the matrix, and the graphite nodule separates from the matrix after fracture. Consequently, some dimples without graphite nodules are visible in the fracture surface. The fracture path shows that cracks often bypass graphite nodules by propagating along the graphite-matrix interface.

[Figure 5: see original paper] shows the deformation morphology of the matrix near the fracture surface after impact at -20°C. The matrix surface near the fracture appears wrinkled and uneven, with wavy slip traces visible. These wavy slip lines result from cross-slip, where slip transfers from one slip plane to another via screw dislocation cross-slip. When extensive cross-slip occurs, wavy slip lines form on the matrix surface [14,15], indicating that the matrix near the notch root of the ferritic ductile iron specimen undergoes significant plastic deformation at -20°C. When the plastic deformation exceeds a certain limit, crack initiation and propagation occur [16].

[Figure 6: see original paper] shows typical micro-void morphology around graphite nodules after impact at -20°C. Micro-voids form around graphite nodules in the direction of the tensile axis, with fine slip bands visible in the surrounding matrix, indicating localized plastic deformation. Microcracks are also observed within graphite nodules themselves, suggesting that graphite nodules undergo slight deformation along with the surrounding matrix.

[Figure 7: see original paper] shows an SEM image of the longitudinal section at the fracture after impact at -20°C. The ferrite matrix between graphite nodules undergoes severe plastic deformation, and the final tearing region contains numerous parabolic-shaped fine dimples formed by micro-voids nucleated at inclusions under tensile loading. This demonstrates that at -20°C, the ferrite matrix undergoes ductile tearing, and the plastic deformation in graphite-free regions of the ferrite matrix is substantial.

Based on the above analysis, the basic process of impact fracture in ferritic ductile iron at -20°C is as follows: First, decohesion occurs at graphite-matrix interfaces, followed by non-uniform plastic deformation of the ferrite matrix, then microcrack initiation, and finally connection of multiple microcracks into a main crack. Due to stress concentration, microcracks near the notch root easily develop into the main crack. Ahead of the main crack, additional microcracks initiate and extend, connecting with the main crack front through rapid, jump-like propagation along paths of lower resistance and energy consumption until complete fracture occurs. These processes are not strictly sequential but rather overlapping and intersecting [17,18].

2.2.2 Crack Initiation and Propagation at -45°C

At -45°C, ferritic ductile iron is in the ductile-brittle transition temperature range, representing the transition from dimple fracture to cleavage fracture. [Figure 8: see original paper] shows the fracture morphology near the notch after impact at -45°C. The V-notch is at the top, and the cleavage plane below the notch is the crack initiation region. At the origin of the river pattern, graphite nodules are found at the crack source, indicating that crack initiation is related to graphite nodules [19].

[Figure 9: see original paper] shows an SEM image near the fracture surface after impact at -45°C. Cracking is observed within the matrix near the impact fracture, with diverse manifestations. Most cracks initiate at graphite nodules and terminate within the matrix, demonstrating that matrix cracking is closely related to the presence of graphite nodules. This extensive matrix cracking phenomenon only exists near the fracture surface of impact specimens within the ductile-brittle transition temperature range. In this temperature interval, the matrix cannot undergo sufficient plastic deformation, so cracking occurs to release stress.

[Figure 10: see original paper] shows graphite morphologies near the fracture before and after impact at -45°C. Comparing Figures 10a and 10b reveals significant changes in graphite nodules near the fracture surface after impact. Based on the original graphite shape, voids of various shapes form around graphite nodules at the graphite-matrix interface. After impact fracture, radial cracking occurs at the graphite-matrix interface, with internal cracking also observed within graphite nodules. Some irregularly shaped graphite nodules generate microcracks in the surrounding matrix, suggesting that these graphite nodules act as crack sources during low-temperature impact.

Compared with graphite nodules and surrounding matrix features after impact fracture at -20°C, decohesion at graphite-matrix interfaces occurs more readily at -45°C, while plastic deformation around graphite nodules is less severe than at -20°C, with fewer wavy slip lines in the matrix. [Figure 11: see original paper] shows the deformation morphology of the matrix around graphite nodules after impact at -45°C. Figure 11a shows limited slip lines around graphite nodules at the fracture surface. Some regions around graphite nodules undergo plastic deformation, while others exhibit cleavage fracture (Figure 11b). Thus, after impact at -45°C, the matrix around graphite nodules shows mixed ductile-brittle fracture characteristics.

2.2.3 Fracture Morphology and Crack Initiation Mechanism at -80°C

[Figure 12: see original paper] shows the SEM image of the fracture surface after impact at -80°C. Cleavage plane crack propagation primarily occurs across grain boundaries. In ferritic ductile iron, cleavage fracture regions typically reflect grain size. Grain boundaries with different orientations can arrest cleavage cracking, resulting in variously arranged steps. When a cleavage crack crosses a twist boundary, the crack cannot simply pass through due to different orientations of cleavage planes on either side. Instead, new nucleation occurs at stress concentration points where the cleavage crack intersects the grain boundary. The new nucleus expands in the new grain and merges with other cracks during propagation to form cleavage steps [20].

[Figure 13: see original paper] presents a schematic of the cleavage fracture process in ferritic ductile iron. The crack propagation direction is indicated by arrows, showing that the low-temperature cleavage fracture process consists of three continuous stages: microcrack initiation at graphite-matrix interfaces, microcrack propagation across graphite-matrix interfaces, and crack propagation controlled by the ferrite matrix.

Compared with matrix cracking near the fracture at -45°C, only limited cracking and some small microcracks are observed near the fracture surface at -80°C. However, the fracture edge characteristics change significantly. Compared with graphite nodules and surrounding matrix features after impact fracture at -45°C, the matrix at -80°C shows smooth fracture surfaces, with some regions exhibiting step-like fractures. Graphite nodules themselves show almost no deformation (Figure 14a), and matrix deformation around graphite nodules is minimal. For individual graphite nodules, the surrounding matrix appears nearly planar, indicating large-section cleavage fracture in the surrounding matrix, as shown by the cleavage planes and river patterns in Figure 14b.

Notably, when the impact temperature decreases to -80°C, numerous tongue-like patterns appear on cleavage planes. Research [21] indicates that tongue-like patterns result from a cleavage crack turning into a nascent twin, an already-formed twin, or a crack created at the twin-matrix interface during twin formation. The formation of cleavage tongues is related to cleavage crack propagation along the interface between deformation twins and the matrix. Such deformation twins form ahead of the crack tip when the cleavage crack propagates at high speed [22]. In ductile iron matrix, tongue-like patterns often appear in groups on relatively large cleavage planes accompanied by graphite nodules, with two sets of tongue patterns perpendicular to each other, as shown in [Figure 15: see original paper]. Connecting the model of perpendicular tongue patterns with perpendicular twin systems, the straight [110] lines and some [110] lines along the bottom of tongue patterns are traces of the intersection between the main crack cleavage plane {100} and twin plane {112}. The edges of a step match well with the bottom of tongue patterns (Figure 16). Therefore, the microscopic mechanism of cleavage fracture caused by twins during impact below the ductile-brittle transition temperature is "microcrack initiation and propagation induced by perpendicular intersecting deformation twins." Based on this, it can be inferred that cleavage fracture below the ductile-brittle transition temperature in ductile iron is primarily caused by twin-initiated cracking.

During impact fracture below the ductile-brittle transition temperature, the condition for deformation twin formation is that the applied stress reaches the yield strength level. Stress concentration caused by deformation twins leads to cleavage fracture, so forming critical-sized deformation twins is the critical condition for cleavage fracture. Assuming stress concentration occurs at twins, the stress concentration factor q is proportional to twin size p, meaning larger twins produce greater stress concentration [23,24]. According to Griffith's energy criterion for cleavage fracture, the cleavage fracture stress σ is given by:

σ (where σ is yield stress, σ is lattice friction resistance, g is effective surface energy, E is elastic modulus, b is Burgers vector magnitude, and q_c is the critical stress required to raise local stress to the fracture stress level).

The yield stress follows a Hall-Petch type relationship: kd^(-1/2) (where d is grain size and k is a constant related to crystal type).

Assuming twin size p is: then the critical twin size required to generate critical stress q_c is Eγb/1.

According to reference [24], in the twin-induced cleavage region, k = 2.48$\times$10$^-10m, d = 2\times10^7 N/m^2, and = 5.6\times10^-5m. Substituting into equation (5) yields p_c =$ 3.78×$10^-7 m. [Figure 17: see original paper] shows an SEM image of deformation twins in the fracture surface of ductile iron after impact at -80°C. The measured average thickness of deformation twins is approximately 500 nm, which matches the calculated value and exceeds the critical twin size, indicating that the formation of a certain number of deformation twins larger than the critical size leads to cleavage crack initiation. As temperature decreases, the number of deformation twins in the matrix increases. This crack initiation and propagation mechanism caused by deformation twins results in the dramatic decrease in both crack initiation energy and crack propagation energy during impact at -80°C.

Conclusions

(1) Above the ductile-brittle transition temperature, the impact fracture process in ferritic ductile iron proceeds as follows: First, decohesion occurs at graphite-matrix interfaces, followed by non-uniform plastic deformation of the ferrite matrix, then microcrack initiation, and finally connection of

multiple microcracks into a main crack. These four processes overlap and intersect throughout fracture.

(2) In the ductile-brittle transition temperature range, cleavage cracks are initiated by graphite nodules and inclusions. The matrix around graphite nodules exhibits mixed ductile-brittle fracture characteristics, with protruding step morphologies present near the notch region and obvious cracking observed in the matrix near the fracture surface.

(3) Below the ductile-brittle transition temperature, deformation twins appear in cleavage fractures. The measured average thickness of deformation twins is approximately 500 nm, exceeding the critical twin size. Under these conditions, forming a certain quantity of deformation twins larger than the critical size is the prerequisite for crack initiation. As temperature decreases, the number of deformation twins in the matrix increases. This crack initiation and propagation mechanism caused by deformation twins leads to the dramatic decrease in both crack initiation energy and crack propagation energy during impact at -80°C.

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Note: Figure translations are in progress. See original paper for figures.