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Fatigue of Heavy-Vehicle Engine Materials

Experimental Analysis and Life Estimation

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Abstract

The heavy-vehicle automotive industry is constantly subjected to higher demands. In particular, new European emission standards are formulated with the intention of improving the environmental friendliness of newly-produced vehicles through reduced exhaust emission. In one way or another, this implies a successive improvement of the engine efficiency, which in turn, inevitably will require a higher combustion pressure and temperature. This is a respectable challenge for future engine constructions, but also for the engineering materials used to embody them. As higher thermal and mechanical loads must be sustained, there is a higher rate of wear, and consequently, a negative effect on the extent of the engine lifetime.

The aim of the present thesis is to confront the expected increase in rate of wear, henceforth referred to as fatigue, by studying the effect on materials typically employed in heavy-vehicle engines, namely cast irons. Foremost, the intention has been to improve the understanding of the physical mechanisms of fatigue in these materials, in order to develop a lifetime estimation method designated to assist the mechanical design of heavy-vehicle engines.

In essence, a large set of thermo-mechanical fatigue (TMF) and combined thermo-mechanical and high-cycle fatigue (TMF-HCF) tests has been conducted at engine load conditions on laboratory specimens of lamellar, compacted and spheroidal graphite irons. In this way, these three different material groups have been experimentally compared and the associated fatigue mechanism has been studied. In particular, a new property related to TMF-HCF conditions has been identified and measured, . Regarding the fatigue mechanism, it has been affirmed to consist of the initiation, propagation and coalescence of numerous microcracks. Based on this, a successful lifetime assessment model was formulated, allowing good estimations of the fatigue life of laboratory specimens subjected to both TMF and TMF-HCF conditions.

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List of Papers

The following papers have been included in this thesis:

- I. V. Norman, P. Skoglund and J. Moverare. Damage evolution in compacted graphite iron during thermomechanical fatigue testing. *International Journal of Cast Metal Research*. *Article in press*.
- II. V. Norman, P. Skoglund, D. Leidermark and J. Moverare. Thermo-mechanical and superimposed high-cycle fatigue interactions in compacted graphite iron. *International Journal of Fatigue* 80 (2015) 381-390.
- III. V. Norman, P. Skoglund, D. Leidermark and J. Moverare. The effect of superimposed high-cycle fatigue on thermo-mechanical fatigue in cast iron. *In manuscript*.

Own contribution to the papers included:

In all the three papers I have been the main contributor, performing all the experimental and theoretical work, as well as the manuscript writing, under the supervision of Peter Skoglund, Daniel Leidermark and Johan Moverare.

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Abbreviation

CGI	Compacted graphite iron
HCF	High-cycle fatigue
IP	In-phase
LCF	Low-cycle fatigue
LGI	Lamellar graphite iron
OP	Out-of-phase
TMF	Thermo-mechanical fatigue
TMF-HCF	Combined thermo-mechanical and high-cycle fatigue
SEM	Scanning electron microscopy
SGI	Spheroidal graphite iron

Part I

Background and Theory

1

Introduction

If only two words were to be selected to give the context of this licentiate thesis, they would probably be *mechanical design*. To most people, “design” stands for the conception of a product, or component, having a certain function and appearance, while “mechanical” signify that the main purpose of the component is to carry some kind of load. Inevitably, a mechanical component will have a mass; it will conduct heat; it has to be manufactured somehow; and it will wear out as it has to withstand the test of time within a given environment. To narrow it down, this thesis will be about the role of *materials* in the mechanical design of heavy-vehicle internal combustion diesel engines.

The automotive industry constitute a significant part of the Swedish export, about 10 percent of the total exported value in 2013¹, to which one forth is contributed by the heavy-vehicle industry. Today, the need of heavy-vehicle transportation is extensive, for instance 86 percent of the domestic cargo transported was made by heavy-vehicles in Sweden in 2010², but also the demands on environmentally accepted transport are increasing. The European union regularly formulates new European emission standards to which automotive manufacturers have to conform. Thus, there is an emerging challenge for heavy-vehicle manufacturers to fulfil the upcoming demands on emission, which in the long run will require a significant increase in engine efficiency implying higher combustion pressures and temperatures.

The above demands have called for more sophisticated engine design methods, as well as improved methods to evaluate any conceived design solution. A mechanical designer might not know how well his or hers solution qualifies until a prototype physically exists; a wait which can be very

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¹The foreign commerce of Sweden in 2013, issued by Kommerskollegium in 2014

²Rapport 2012:8, issued by Trafikanalys in 2012

long, not to mention the additional prototype production cycles eventually required due to the iterative nature of the design process. With the advent of modern computational power, this inconvenient situation is hoped to be circumvented. If the mechanical designer was to be given reliable simulation tools, the dependence on prototype evaluation would be less striking, since mistakes related to the physical behaviour of the components could be avoided already at the conceptual design level. Thus, an idealistic goal would be to be able to construct an entire vehicle virtually and to evaluate it before manufacturing an optimal real version.

This thesis will focus on the life of engine materials, *i.e.* how long the material can endure the loads to which it is exposed before it fails, which will be referred to as the *fatigue life* of the material. This thesis is not the first to address this issue. It is an entire field of science called fatigue of materials which deals with the permanent and successive degradation of materials due to a repeated pattern of loading and unloading [1]. Consider a metallic paper clip for instance, which inevitably will break if it is repeatedly bended. In a similar but more complicated manner, the engine material will be fatigued due to the complex thermal and mechanical loads to which it is subjected.

Today, there are already many computer tools commercially available, such as computer-aided design (CAD) and computer-aided manufacture (CAM) software, finite element (FE) software to perform mechanical analyses and more, which often are used in the design process. However, when it comes to life estimation of engine components, there is much less to choose from. This is a consequence of the complex behaviour of materials and the fact that different materials act very differently regarding their fatigue and failure mechanisms. Therefore, there is a need to investigate and characterise the fatigue behaviour of the materials employed in heavy-vehicles in order to develop accurate virtual evaluation tools. In addition, the present investigation also aids the ongoing engine development directly since a better understanding of the limits of the materials alleviates the understanding of the components. These limitations imposed by the material will remain a matter of relevance, especially since higher combustion pressures and temperatures eventually are demanded which will enforce even stricter requirements in the future.

1.1 Research questions and aims

The research questions are summarised into the following three:

- i) How is the material degraded? What are the physical mechanisms responsible for the successive deterioration towards a state beyond operation ability?
- ii) How is the fatigue life affected by a variation in the load conditions? By which factors does the life of the materials decrease as the thermal and mechanical loads are increased?
- iii) Given a proper definition of material failure and the exact load situation, how can the critical point when the material fails be estimated with a reasonable accuracy?

The answers to these questions are supposedly not independent. Rather, they are likely correlated, implicating that the outcome of one is applicable on the others. The strategy has been to perform an extensive experimental investigation in order to have a rich physical idea of the fatigue processes which in turn has been used to estimate the fatigue life as the load condition is varied.

The above formulated research questions are expected to lead to the fulfilment of the purpose of the thesis, which is to characterise the fatigue behaviour and mechanisms in order to also develop an accurate life estimation method for the heavy-vehicle industry.

1.2 Outline and scope of the thesis

The thesis is divided into two parts, starting with the *Background and Theory* which will briefly present the technological context and the current state of the corresponding academic field. The second part, *Included papers*, consists of enclosed academic papers previously produced which constitute the academic contribution done within this research project.

After the introduction of the first part, the second chapter will present the particular material groups of interest, namely different grades of cast irons, and the loading condition experienced by the materials due to the operation of an internal combustion diesel engine. Subsequently, a review of the fatigue mechanisms in cast irons is presented in the third chapter and the experimental methods in the fourth. The fifth chapter discusses the included papers, and in the sixth and final chapter, an outlook is presented to see what is beyond this thesis.

The scope of the thesis mainly involves the analysis of fatigue testing and the microstructural investigation of laboratory specimens. Thus, the investigation has been limited by excluding the study of the fatigue behaviour and life of entire engine components. Consequently, also the life assessment work performed is presently only applicable on laboratory specimens.

2

Engine materials and load conditions

This thesis will not treat heavy-vehicle engine design, but rather how it is influenced by the choice of materials. Even so, something must be said about the considered mechanical components, as they will set the frame in which the constitutive materials will be fatigued. Therefore, the present chapter will deal with the concerned engine components and the nature of the associated load condition, as well as the typical material group employed.

The construction of a heavy-vehicle four-stroke diesel engine is displayed in a simplified schematic illustration in Figure 1. The central function is to transmit a momentum to a driving shaft which in turn generates the forward motion of the vehicle, by a periodic work conversion cycle involving fuel ignition and the resulting power transfer through the back-and-forth motion of the pistons.

The three components, namely the *engine block*, the *cylinder head* and the *exhaust manifold*, illustrated in Figure 1 are those of main interest; however there are other engine components to which the outcome of the thesis is relevant as well. The engine block, or cylinder block, houses the reciprocating motion of the pistons and the recurrent fuel ignition. In the cylinder head, which is located on top of the cylinder chamber, fuel and exhaust gas are transported interchangeably within intake and exhaust channels through regulated valves. In the third illustrated component, the exhaust manifold, the exhaust gas is collected and discharged from the cylinder head into subsequent units, which could be a turbocharger, air filters or catalytic converters.

2.1 Engine materials

Presently, all three components are produced from materials categorised as *cast irons*, meaning the metallic alloy associated with the three main compo-

Figure 1: A schematic illustration of the engine components of interest.

nents, iron, carbon and silicon; as well as its characteristic method of manufacture through casting. When a metal is cast, it is poured into a mould in a melted state and thereafter cooled such that the melt gets solidified into the particular shape determined by the mould. This is a convenient way to manufacture components and many engineered objects are produced in this way. Thus, the main advantage of cast irons is the ability to produce the complex geometries often required in the automotive industry, but also the good thermal conductivity and the low cost of the raw materials.

As a surprise to most people, cast irons should not be regarded as a single material with only one set of properties. Rather, it is a versatile alloy whose properties can be widely modified by slightly changing the chemical composition, the cooling rate during solidification or by applying a post heat treatment. The grand diversity can be deduced from the variability of the two-fold structure of cast irons which is the most prominent feature of this material group, namely a structure consisting of embedded graphite particles in a steel matrix, see the examples in Figure 2. In contrast to the graphite particles which always consist of a graphite phase containing only carbon, the constituting phases of the matrix can take on different forms, such as ferrite, austenite, cementite, martensite or any combination, in a similar manner as the phases can be varied in conventional carbon steels. Thus, the overall properties of cast irons may be varied through the variation of the matrix phases, however more significantly, the properties are also varied through the variation of the graphite shape; see some examples in Figure 2. Whatever the case may be, there must always be an occurrence of the graphite phase

Figure 2: Examples of different graphite structures embedded in a steel matrix exhibited in cast irons, (a) lamellar graphite iron (b) compacted graphite iron and (c) spheroidal graphite iron.

in order to be included in the cast iron family. Consequently, compared to steels, which also are iron-carbon alloys, there is a clear distinction marked by a carbon content of about $2.11wt\%$ below which there is not enough carbon atoms to engender the graphite phase. By the same reason, silicon was included in the definition of cast irons as it stabilises the formation of graphite during the solidification process.

Cast irons are often categorised according to their corresponding graphite shape. The most common group of cast irons is the lamellar graphite iron (LGI), often also called grey cast iron, in which the graphite particles are elongated and flake-like giving the appearance shown in Figure 2(a). In general, this is the most employed cast iron in industry, from sewer pipes and automotive parts to kitchen cookware. LGI is easy to cast, has low strength and ductility but also good machinability, thermal conductivity and vibration damping [2]. On the contrary, in compacted, or vermicular, graphite iron (CGI) the graphite particles are coarser and rounder compared to LGI as seen in Figure 2(b). As a consequence, the CGI grades have superior mechanical properties. Even more superior is the spheroidal graphite iron (SGI), also called ductile cast iron owing to the enhanced mechanical properties, where the graphite particles exhibit a spherical shape, see Figure 2(c). SGI grades have a high strength and ductility, but also reduced machinability, thermal conductivity and damping compared to the LGI and CGI. Other groups of cast irons which are not within the scope of this thesis are white cast iron and malleable cast iron.

Currently, each of the above presented components is most commonly produced in one of the mentioned group of cast irons. The engine block and the cylinder head are often cast in a LGI or CGI grade. Conversely, the exhaust manifold has to endure higher temperatures than these two components, which is due to the higher heat absorption from the exhaust gas and the less efficient cooling, and is therefore made of SGI.

2.2 Engine load conditions

To measure the service lifetime of a component is not enough. The life must be related to the corresponding load level to which it is exposed, in order to be informative. In this regard, a load can be purely mechanical, *i.e.* an applied force or deformation, or thermal, denoting elevated temperatures attained while allowing free thermal expansion. Conversely, if the thermal expansion is mechanically impeded or amplified, then the loading is said to be thermo-mechanical. An aggressive chemical environment, such as in the combustion chamber or in the exhaust gas channels, could also be regarded as a load due to its potential to reduce the life of the components.

Thus, in the running engine, the material will experience mechanical, thermal and corrosive loads due to the combustion and the resulting heat and pressure bursts. Consequently, the parts closest to the combustion chamber, in the engine block and the cylinder head, will become hot and heat will be conducted away to the far off located parts, resulting in the development of a thermal gradient across the components. Since a heterogeneous thermal expansion will follow, thermal stresses develop as the expansion in each point is differently constrained depending on the surrounding. Similarly, the exhaust manifold becomes thermo-mechanically loaded as hot exhaust gas recurrently enters the exhaust channels and heats up the component.

The attained temperatures are very likely to be more or less stabilised, however, due to the cyclic back-and-forth motion of the pistons and the reoccurring pressure bursts, there is supposedly an additional high-frequent mechanical load felt by the material. In addition, a vibrational load could also be induced by the roughness of the road surface. On the other hand, the amplitude is believed to be small, however significant when considering the life of the components.

Each time the engine is turned on and off, the above course of events are repeated. Thus, the entire engine is heated up and cooled down repeatedly as the driver stops for red lights and stops for pauses. The period elapsed between a state of inactivity to a steady state of full operation and then back to a cooled state of inactivity defines a characteristic *load cycle*, which in the present context often is referred to as the start-operate-stop cycle. Typically, the length of this load cycle could range from a couple of minutes to a couple of hours depending on the driving circumstances. Figure 3 shows a schematic example of how the load cycle could be in a point close to the cylinder chamber regarding the temperature and the mechanical strain; the latter being properly defined below. A negative mechanical strain is anticipated, notably in the cylinder head, because the thermal expansion of the hottest part is believed to be impeded by the surrounding less hotter parts. The

Figure 3: Schematic representation and diagrams of the load conditions encountered at the hot parts of engine components.

high-frequency oscillation represents the above mentioned vibrational loading. It must also be mentioned that this is a simplification to an uni-directional case, *i.e.* where one only considers loading in a single direction. In the real case, the loading is multi-axial meaning that the local stress and strain state at different parts of the component may likely include non-zero shear components, in contrast to the uni-axial case.

The identification of the start-operate-stop cycle has led to the introduction of the concept of *thermo-mechanical fatigue* (TMF), which refers to the fatigue damage accumulation due to the combination of a conventional cyclic mechanical load and a varying temperature. The concept involves the additional damage mechanisms induced by the changing temperature, but it is also used when referring to the associated standard fatigue test set-up, which has been employed in this work and therefore will be described later in Chapter 4.

Introducing a varying temperature in a fatigue test complicates the deformation analysis since there will also be a thermal expansion which is added up together with the deformation imposed by the applied forces. For convenience, one therefore prefers to talk about *mechanical strain* instead of just strain as a measure of deformation, where the latter includes the thermal expansion and the former does not. In an uni-axial case, the two strain measures are related as

$$\varepsilon = \varepsilon_{Th} + \varepsilon_{Mech} \quad (1)$$

Figure 4: Schematic illustration of how (a) the TMF strain and HCF strain are added to engender (b) the total mechanical strain.

where ϵ is the strain measured by an extensometer, ϵ_{Th} is the thermal strain representing the thermal expansion and ϵ_{Mech} is the mechanical strain which is related to the deformation induced by mechanical forces alone. Thus, when dealing with combined mechanical and thermal loading, it is more convenient to relate the mechanical load part to the mechanical strain since this strain measure is not explicitly dependent on the thermal load part.

Furthermore, the mechanical strain is believed to have a superimposed vibrational component, as illustrated in Figure 3. Thus, the anticipated load condition is more complex than what is originally contained in the TMF concept. Rather, the load cycle corresponds to a combination of thermo-mechanical and high-cycle fatigue (HCF), where the latter is the term for low-amplitude fatigue. Due to this, Equation 1 is slightly modified to separate the mechanical strain into a TMF and a HCF component,

$$\epsilon = \epsilon_{Th} + \epsilon_{TMF} + \epsilon_{HCF} \quad (2)$$

where ϵ_{TMF} represents the slow-varying mechanical strain corresponding to the start-operate-stop cycle and ϵ_{HCF} represents the strain oscillation relative to the TMF strain. The TMF and HCF strain signals, as well as their combination, are illustrated in Figure 4. In order to emphasize that the combined signal ϵ_{Mech} is composed of two mechanical strain signals, it will henceforth be denoted as the *total mechanical strain*. Furthermore, the magnitude of the corresponding strain ranges, *i.e.* the difference of the maximum and minimum strain value, namely the TMF strain range $\Delta\epsilon_{TMF}$, the HCF strain range $\Delta\epsilon_{HCF}$ and the total mechanical strain range $\Delta\epsilon_{Mech}$, are also illustrated in Figure 4.

3

Fatigue crack initiation and propagation in cast irons

As thermal and mechanical cyclic loads are applied to the engine materials, irreversible physical processes occur which render the material in a damaged state. These fatigue mechanisms involve the initiation and propagation of *fatigue cracks* which eventually will lead to the final failure of the considered component. The intention of this chapter is to present what is currently known about the damage mechanisms, in particular regarding crack propagation, associated with the monotonic and cyclic deformation of cast irons. The following review is restricted to lamellar (LGI), compacted (CGI) and spheroidal graphite iron (SGI) which were introduced in the previous chapter.

3.1 Crack initiation and propagation during monotonic loading

Before dealing with the fatigue mechanisms associated with cyclic loading, it is relevant to start with a review of the damage processes occurring during monotonic loading, since these are similar to the mechanisms observed at cyclic loading conditions.

The initial damage processes are highly related to the interaction between the graphite phase and the steel matrix mentioned in Chapter 2. As many authors have observed on polished surfaces of cast iron specimens, small deformations result in the delamination, or *debonding*, of the graphite-matrix phase boundary [3–11]. He et al. [9] reported that the tensile strain required for such damage initiation can be less than 0.03% for ferritic LGI and SGI, which later also was verified by Dong et al. [10] reporting a stress level of

Figure 5: (a) A schematic illustration of the different damage and fatigue mechanisms occurring in cast irons; (i) graphite-matrix debonding (ii) graphite cleavage (iii) graphite-initiated microcracks (iv) microcrack coalescence. (b) SEM image demonstrating the damage processes in CGI due to a large tensile load. Graphite-matrix debonding and graphite cleavage are indicated with white arrows and the resulting large deformation of the adjacent matrix is indicated with black arrows.

80 MPa ($\approx 0.04\%$) corresponding to graphite debonding for ferritic SGI. In addition, investigators have observed cracking within the graphite phase in LGI and CGI as the first sign of damage [4, 6, 12–14], resulting in *graphite cleavage*. The cleavage cracks have been proposed by Glover and Pollard [12], to propagate along the basal plane in the crystallographic structure of graphite, since the interatomic bonding between these planes is weaker than the bonds within the planes. Thus, the graphite phase is clearly the weak point of cast irons as it is the first to fail, either through graphite-matrix debonding or graphite cleavage; both schematically illustrated in Figure 5.

As the monotonic tensile load is increased to higher strains, the cracks start to enter the matrix at graphite tips oriented perpendicular to the load direction and eventually coalesce as multiple cracks propagate simultaneously [8, 10, 11, 13, 15]. These graphite-initiated cracks which have managed to extend into the matrix, possibly connecting several isolated graphite particles, will in the continuation be referred to as *microcracks*. Microcrack propagation and coalescence in LGI, CGI and SGI were extensively studied by Voigt et al. [8, 13] using in-situ scanning electron microscopy (SEM). In SGI, they could observe the formation of a larger dominant crack at an early stage due to the coalescence of microcracks initiated from graphite nodules. As the main crack successively grew longer, new microcracks were also observed to initiate in front of the crack tip, eventually becoming linked-up with the main crack. The process was similar in LGI and CGI, however the formation of a dominant crack could not be definitively distinguished from

the individual growth of microcracks. Nevertheless, it can be generally concluded that the crack growth leading to failure is a complex process in cast iron, involving the evolution of interlinked microcrack networks.

As the graphite phase fails, many investigators have reported observation of severe deformation of the adjacent matrix [8, 13, 16–19]. This kind of deformation can be seen between graphite particles in Figure 5(b) indicated by the black arrows. The local matrix deformation is readily explained by the stress concentrations generated either due to the significant difference in stiffness of the two phases, or more likely, due to the lost load-carrying ability of the graphite phase as it fails in tensile deformation. The effect is clearly visible at low macroscopically applied strain levels [8, 13, 16], because locally, the intensified deformation surpasses the yield limit resulting in permanent plastic deformation. Such local plastic flow has been observed experimentally using acoustic emission methods by Sjögren and Svensson [17] and digital image correlation techniques by Sjögren et al. [18, 19]. Starting with the hypotheses proposed by Gilbert [3, 20, 21], local plastic flow in combination with graphite failure, have for a long time been assumed to be the underlying reasons for the complicated constitutive stress-strain behaviour of cast irons, involving non-linear elasticity, tension-compression asymmetry and varying elastic properties. Many investigators have attempted to capture these uncommon mechanical features into a constitutive model by considering the micro-mechanical interaction of the graphite and matrix phases [22–26].

In particular regarding the varying elastic properties, Haenny and Zambelli [27, 28] made a couple of investigations where they measured the elastic modulus during partial unloading, referred to as the unloading modulus, and its variation due to the stress level, annealing treatments and the strain history. They argued that microcrack propagation in the matrix, in combination with localised plastic yielding and the resulting residual stress development upon complete unloading, rationalises the observed anomalous behaviour of the unloading modulus. Thus, it is reasonable to believe that the occurrence of microcracking does not only govern the fracture and fatigue behaviour, but also the constitutive mechanical behaviour of cast irons to some extent.

Figure 6: Optical microscope image of a microcrack emanating from (a) a graphite particle in CGI and (b) a microshrinkage void in SGI.

3.2 Crack initiation and propagation during cyclic loading

Fewer investigations have been done on crack propagation during cyclic loading. On the other hand, the processes observed during cyclic loading have been observed to be very similar to those present under monotonic loading, which will be seen in the present section.

In cyclic loading, microcracks have been observed to initiate at multiple graphite tips on polished samples of LGI, CGI and SGI, propagating transversally into the matrix [29–37], including the work of Norman et al. [38, 39] (Paper I and III), as similar to the monotonic loading case, see Figure 5(a). Some authors [29, 30, 33] have also claimed that the instant of crack initiation occurs during the first 10% of the fatigue life. Thus, the major part of the fatigue life consists of crack propagation in contrast to other metals where a large number of cycles are required before the first crack is initiated. This does not come as a surprise in view of the previously presented damage mechanisms occurring during monotonic tensile deformation of cast irons. Accordingly, it is reasonable to believe that some graphite particles lose their load-carrying ability during the first load cycle, either by debonding or cleavage, thereby generating fatigue crack initiation sites due to the resulting stress concentrations at the edge of the graphite particles, see Figure 5(a). A real example of a graphite-initiated microcrack is shown in Figure 6(a).

In the case of SGI, it has been seen that graphite nodules are not the only microcrack initiation site. Several investigators have reported of crack initiation at microshrinkage porosities [30, 40–43], which are casting defects

that may develop during solidification due to the volume difference associated with the liquid-solid transformation. An example of a microshrinkage and a microcrack initiated at its boundary are shown in Figure 6(b). It is however important to note that cracks not necessarily have to initiate at porosities, as graphite nodules have been shown to be distinct crack initiation sites even in the presence of porosities [43]. Nevertheless, it has been concluded that porosities play an important role in determining the fatigue resistance of SGI, especially depending on the size and location of the defect [42, 43].

The initiated fatigue microcracks propagate due to the recurrent application of load cycles, which, as in the case of monotonic loading, eventually results in the possible encounter of another graphite particle or microcrack. Consequently, the microcrack growth has been observed to progress incrementally, *i.e.* stepwise as microcracks occasionally are linked-up, increasing the total crack length abruptly [29, 30]. Fatigue crack growth involving microcrack coalescence has also been reported in the case of thermal fatigue [31, 35], as well as for TMF and TMF-HCF tests [37]. Thus, regardless of the particular fatigue mode employed, the fatigue process in cast irons seems to be following a general pattern consisting of a brief fatigue crack initiation phase and a fatigue crack propagation phase involving the growth and coalescence of multiple microcracks.

Norman et al. [38] (Paper I) further affirmed the general fatigue process using the concept of the unloading modulus, following the previously mentioned hypothesis put forth by Hanney and Zambelli [27, 28]. It was observed that regardless of the fatigue test, either TMF or isothermal low-cycle fatigue (LCF), and the magnitude of the load amplitudes, the unloading modulus always decreased linearly with the number of cycles until a consistent critical value, corresponding to a total loss of 6-8% of the initial unloading modulus value, was attained. Therefore, it was proposed that the fatigue life is consumed following a general process, in which only the rate-of-process is dependent on the load variables.

Similar trends in the stress response due to cyclic strain loading have been observed in LGI by Fash [29], who observed a linear decrease in stress amplitude with increasing number of cycles until a point of rapid loss shortly before the final failure of the specimen. He related this mechanical behaviour to the development of microcracks using replica techniques and noticed that the drop in the stress amplitude coincided with the rapid coalescence of microcracks becoming the final crack responsible for failure of the specimen. The mentioned evolution of the stress amplitude and the unloading modulus, as well as the crack length, are displayed schematically in Figure 7. Furthermore, Socie and Fash [30] reported in a subsequent investigation that the longest microcrack found on the verge to the final coalescence was around

Figure 7: Schematic representation of the results presented by Fash [29] and Norman et al. [38] (Paper I) showing how the stress amplitude, unloading modulus and crack length change with increasing number of cycles. Crack length is reported as the largest portion of the microcrack eventually becoming the final crack responsible for failure.

1mm to 2mm, for all studied cast irons, namely LGI, CGI and SGI.

Regarding TMF testing with superimposed HCF, only a handful of experimental studies have been carried out on cast irons including the work of this thesis [37–39, 44–46]. Nevertheless, it has been reported that there is an existence of a consistent HCF strain range, below which the fatigue life is unaffected by the superimposed HCF load, for LGI, CGI and SGI [39, 44, 46]. Regarding the fatigue crack mechanisms, extensive studies have only been conducted on SGI. For instance, Hammers et al. [47] categorised different fatigue cracks into three different types depending on the type of crack, namely surface-initiated, bridging two neighbouring nodules or nodule-initiated. More importantly, they also concluded that the crack growth characteristic was unchanged when applying a superimposed HCF strain range, *i.e.* the crack length profile as a function of number of cycle normalised to the number of cycles to failure appeared to be the same regardless of the value of the HCF strain range.

3.3 Elevated temperature in air environment

An elevated temperature might have a substantial effect on the state of the material even in the absence of a cyclic mechanical load. Similarly, the ambient environment may also have a certain impact, especially at elevated temperature. This section is therefore devoted to the detrimental effects associated with an elevated temperature in ambient air, and their interaction with a mechanical load cycle.

When cast iron specimens are exposed to elevated temperatures for a long time in ambient air, they can grow in size and experience surface oxidation, or *scaling* [48–53]. The scaling phenomenon is clearly reflected by preforming mass evolution measurements, where significant mass changes are observed after a long period of time at high temperatures due to the additional mass acquired through surface oxide formation [51–53]. However, not only the scale formation is supposed to affect the mass evolution. There is also a loss in mass because of the graphite removal due to the carbon-oxygen reactions producing carbon monoxide or dioxide; a process often referred to as *decarburisation* [50, 54]. As a consequence, decarburisation will lead to the excavation of the material at the surface and a resulting porosity, especially if the graphite structure is interconnected as in the case of LGI and CGI [54]. Regarding the mentioned possible increase of volume, it is often associated with *matrix decomposition*, *i.e.* the structural breakdown of bainite and pearlite into ferrite and graphite [31, 48, 53]. The reason is the metastable nature of the cementite phase which therefore decomposes at elevated temperatures. Since the resulting stable phases have a larger molar volume, the specimen will tend to increase in size.

The three mentioned processes are schematically illustrated in Figure 8(a) and an example of how the surface of a LGI material is affected by scaling and decarburisation in Figure 8(b).

The kinetics of the scaling in cast irons is not clear due to its complexity. It is particularly complicated by the presence of multiple alloying elements and their corresponding diffusional properties and oxygen affinities. Nevertheless, the scale will most frequently consist of iron oxides, such as wüstite (FeO), hematite (Fe_2O_3) and magnetite (Fe_3O_4), as the iron content dominates the composition. However, the significant content of silicon in cast irons has frequently been argued to play an important role in the formation of the surface scale. In particular, it is believed to contribute to an improved oxidation resistance [49–51, 55]. The scaling of the Fe-Si system, which arguably should be similar to the matrix in cast irons, was reviewed by Birk et al. [56] by reference to the work of Adachi and Meier [57]. According to these, the presence of silicon allows the formation of silica (SiO_2), which in

Figure 8: (a) Schematic illustration of the processes; (i) decarburisation, the removal of carbon atoms from the graphite phase; (ii) oxide scaling on the surface and (iii) matrix decomposition involving the phase transformation of cementite (Fe_3C) into ferrite (α) and graphite (G); and (b) an optical microscope image of a cross-sectional view of the oxidised surface in LGI due to a TMF test with a temperature cycle of 100-500°C and a TMF strain range of 0.35%.

turn may react with wüstite (FeO) to form fayalite (FeO_2SiO_4). If the silicon content is high enough, a continuous layer of fayalite, silica, or a combination of the two, is formed [57] which supposedly impedes further oxidation of the matrix. Effectively, silicon is often seen to accumulate between the iron oxide scale and the matrix in LGI, CGI and SGI [50–53, 55]. Several investigators have also reported that a higher content of silicon in SGI results in a slower growth of the iron oxide film, which is consistent with the above argument [49, 51].

The presence of an interactive environment is known to accelerate fatigue crack growth for most metallic materials [1], and the effect is often supposed to be even more important at TMF conditions [58]. This is most likely also the case for cast irons. For instance, Nadot et al. [41] compared crack growth in SGI in air and under vacuum in order to see the difference between cracks propagating within the bulk and surface-initiated cracks. They observed that the crack growth rate measured in crack propagation tests in air is about ten times greater than in vacuum, which suggests that crack growth is accelerated due to the presence of air. On the other hand, there are only a few observations on how high temperature oxidation affects the microcrack initiation and coalescence process mentioned in Section 3.2. It is possible that environmentally-assisted surface cracks are initiated simultaneously as microcracking is occurring within the bulk. Oxide intrusions, possibly tightly related to the decarburisation process mentioned above, have been observed in TMF tested CGI specimens [38, 52] and have been proposed

to be fatigue crack initiation sites [52]. Moreover, Ekström observed both internodular cracks, similar to the crack systems reported by Socie and Fash [30], and oxidised surface cracks propagating through the matrix in SGI after high temperature LCF tests. Thus, the importance of environment-assisted surface crack compared to graphite-initiated cracks is clearly a subject of further investigation.

3.4 Life assessment

Fatigue life assessment is often, but not necessarily, based on the consideration of propagating cracks, where a critical final crack length corresponds to the end of life of the material. This approach is often called a *defect-tolerant approach*, while an approach which ignores the existence of cracks often is labelled as a *total-life approach* [1].

As it has been demonstrated previously, the fatigue life of cast iron is strongly associated with the propagation of microstructurally small cracks. As a consequence, many estimation approaches which have been developed for cast irons over the years are defect-tolerant, *i.e.* relating the life to the length of cracks using concepts formulated within fracture mechanics. Essentially, the interest of this thesis is limited to loading conditions which are typical for heavy-vehicle engines. Therefore, some of the most well-known life assessment models addressed to TMF and TMF-HCF conditions applied to cast irons will be briefly presented below.

Metzger et al. [37] formulated a model for TMF and TMF-HCF tests following the work done by Seifert and Riedel [59] and Seifert et al. [60], where it was assumed that the crack growth rate $\frac{da}{dN}$ is proportional to the cyclic crack-tip opening displacement $\Delta CTOD$,

$$\frac{da}{dN} = \beta \Delta CTOD \quad (3)$$

where β is a constant. Using a couple of other fracture mechanical relations, the $\Delta CTOD$ can be related to variables deducible from a TMF hysteresis loop, such as the stress range $\Delta\sigma$ and the plastic strain range $\Delta\varepsilon_{pl}$, in order to make life estimates of laboratory specimens. Furthermore, to include the effect of a superimposed HCF load, it was hypothesised that crack propagation due to the TMF and HCF cycles are separable as

$$\frac{da}{dN} = \left(\frac{da}{dN} \right)_{TMF} + \sum \left(\frac{da}{dN} \right)_{HCF} \quad (4)$$

Ghodrat et al. [36] successfully used a Paris law in order to estimate the lifetime of notched CGI specimens fatigued in a TMF condition,

$$\frac{da}{dN} = C\Delta K^n \quad (5)$$

where C and n are constants. Similarly, the model presented by Norman et al. [46] (Paper II) was also based on a Paris law, Equation 5, combined with Equation 4 to account for a superimposed HCF load. In contrast to previous models, they interpreted the crack length as an average of the large number of microcracks and the point of failure as the instant when the average microcrack length gets in the same order as the average distance between microcracks.

Finally, it should be mentioned that accurate constitutive models are a prerequisite for lifetime assessment modelling as they constitute the basis of stress calculation of mechanically loaded geometries. The reason is that the lifetime cannot be estimated if the stress state is unknown since this is an input variable in most lifetime assessment models. However, as it was mentioned in Section 3.1, formulating such constitutive models for cast irons is not a simple task because of their complicated mechanical behaviour, which therefore, remains a topic of research. Nonetheless, it is important to also consider the influence of constitutive models when evaluating different lifetime assessment models as they will depend on the accuracy of the preceding constitutive model.

4

Experimental methods

The experimental work is based on thermo-mechanical fatigue tests and metallographic investigations. In this chapter, the materials and methods employed will be described in detail.

4.1 Materials

As indicated in Chapter 2, one cast iron grade from each cast iron category, namely LGI, CGI and SGI, has been investigated. The tested LGI grade was a non-commercial EN-GJL-250 material with a composition optimised for high-temperature properties. Conversely, the CGI material was a commercial pearlitic EN-GJV-400, and the SGI a commercial ferritic high-silicon and molybdenum EN-GJS-SiMo5-1 alloy. The typical microstructures of these three cast irons were shown in Figure 2 in Chapter 2.

4.2 Thermo-mechanical fatigue testing

The concept of thermo-mechanical fatigue (TMF) and its relevance to the characteristic start-operate-stop cycle were introduced in Chapter 2. As it was mentioned, the TMF concept also refers to the particular fatigue test set-up in which mechanical and thermal loads are applied simultaneously. The purpose of such a TMF test is to simulate the load condition of a mechanical component subjected to varying temperatures, in a controlled lab environment; thereby studying the corresponding TMF behaviour of the material.

A standard TMF test is performed in an uni-axial configuration on axisymmetric elongated specimens, where the deformation and load force are

Figure 9: (a) The load cycle employed in all the TMF and TMF-HCF tests, where T_{Max} , $\Delta\epsilon_{Mech}$ and $\Delta\epsilon_{HCF}$ are the studied variables, and (b) the TMF test set-up.

applied in the axial direction. More precisely, a prescribed temperature and uni-axial mechanical strain cycle are applied periodically, which results in a cyclic stress response. Eventually, after a certain number of such load cycles, the test specimen will fail and the number of cycles to failure is acquired, whose value inevitably will depend on the magnitude of the mechanical strain and temperature amplitudes.

Conventionally, the cycle length of the thermal and mechanical cycle are the same, however the phase angle between the two might be varied. The two most commonly employed phase angles are the extreme cases, namely *in-phase* (IP) and *out-of-phase* (OP) testing in which the phase angle is 0° and 180° respectively. In the former, the maximum temperature coincides with the maximum mechanical strain, while for the latter, it coincides with the minimum mechanical strain. Consequently, the considered start-operate-stop cycle seen in Figure 3 can therefore be considered to be in an out-of-phase configuration.

The load cycle selected for the TMF tests is displayed in Figure 9(a). It consisted of 200s of ramping up and down in temperature and mechanical strain and a 25s hold time at each turning point, during which the maximum and minimum values were held constant; giving a total cycle length of 450s. The cycle was significantly shorter compared to the anticipated real load cycle since a test with the latter would require unreasonable long test times. The selected cycle was also motivated by its close resemblance to some of the component tests regularly performed on engine components. Furthermore,

Table 1: Overview of all the conducted tests within the project.

	EN-GJL-250	EN-GJV-400	EN-GJS-SiMo5-1
LCF	x	x	
TMF-OP 100-400°	x	x	x
TMF-OP 100-500°	x	x	x
TMF-OP 300-750°			x
interrupted TMF-OP	x	x	x

the minimum temperature was fixed to 100°C and the maximum total mechanical strain to zero, implying a R_{ϵ} -value, *i.e.* $\epsilon_{Min}/\epsilon_{Max}$, always equal to $-\infty$.

TMF tests were easily transformed into a TMF-HCF test by adding a HCF strain range, as presented in Figure 4 in Chapter 2. However, the maximum *total* mechanical strain value ϵ_{Max} was still always kept at zero, not to change the R_{ϵ} -value. The HCF frequency was given a fixed value of 15Hz ; a reasonable value corresponding to the stroke frequency of a diesel engine. Thus, each TMF-HCF test performed was defined by a variable set including a maximum temperature T_{Max} , a total mechanical strain range $\Delta\epsilon_{Mech}$ and a HCF strain range $\Delta\epsilon_{HCF}$; the latter being zero in the case of a regular TMF test. In addition, a few low-cycle fatigue (LCF) tests were conducted, which can be considered a special case of a TMF test where T_{Max} equals T_{min} .

The TMF experimental set-up was an Instron 8801 servo hydraulic test machine, which is displayed in Figure 9(b). The specimen was heated by induction heating, and cooled by a compressed air flow distributed onto the specimen through three nozzles. All the tests were conducted using an Instron extensometer to measure the strain and a dedicated TMF software developed by Instron for automatic test control.

The specimen geometry was cylindrical with a 6.3mm diameter, 25mm parallel length, 12.5mm extensometer gauge length and 30mm transition radius. The total length of the specimen was 145mm and both ends were threaded for gripping with the thread size M12.

In order to have a consistent measure of the number of cycles to failure, a failure criterion was used in accordance with the code-of practice [61], defined as both a 10% and a 2% drop in the stress range from a tangent line drawn at the last point of zero curvature in the stress range versus number of cycles-plot.

In total, about 150 LCF, TMF and TMF-HCF tests have been conducted within this project on the three presented materials, EN-GJL-250, EN-GJV-

400 and EN-GJS-SiMo5-1. To get an overview, the conducted tests are categorised according to the tested temperature cycle and material in Table 1. Some tests were interrupted TMF tests which were stopped before specimen failure, in order to be metallographically investigated, see Section 4.3 below.

4.3 Metallographic investigations

Since this thesis focuses on the effect of fatigue damage in materials, it is relevant to visualize the damage in order to support the hypothesis formulated. Such visualisation has been done using optical microscopy and scanning electron microscopy (SEM).

Before each microstructural investigation, all the specimens were ground and polished in order to have a flat and smooth surface. Specimens were first cut out and then polished following a standard program developed for cast irons in which the specimens were stepwise ground and polished.

The optical microscope used was a Nikon Optiphot optical microscopy capable of a 1000 times magnification. In addition, the image process software Axiovision was employed to systematically obtain measures of the graphite morphology, such as graphite shape and size distributions, fraction area of graphite and nodularity values.

As mentioned previously, interrupted TMF tests were conducted in order to measure the average microcrack size after different fractions of the total fatigue life using the optical microscope mentioned above. The microcrack lengths were measured as their orthogonal projection in the traversal direction relative to the tensile axis. Dozens of such cracks were measured for each specimen and the average and standard deviation were determined.

A few specimens were analysed using SEM equipment, namely a HITACHI SU-70 field emission gun (FEG)-SEM, equipped with a solid state 4-quadrant BSD detector.

Several specimen of each tested material group was also chemically etched in order to obtain information about the metallurgical parameters, in particular the eutectic cell size. The etching agent was based on picric acid and sodium hydroxide, applied at a temperature of about 80°. An example of how the surface is coloured can be seen in Paper II, Figure 2b. A representative measure of the eutectic cell size was measured as the average of the nine largest cells measured over three different locations.

5

Discussion of appended papers

Part II of this thesis includes the scientific publications produced during the project, and therefore contains all the significant results. In this section, these results will be discussed as a whole, relating them to the research questions formulated in Section 1.1.

Even though it has not been mentioned so far, the first important outcome of the work is the large set of fatigue data generated. Foremost, this includes the measured variation of the fatigue life, *i.e.* the number of cycles to failure, as the load variables T_{Max} , $\Delta\varepsilon_{Mech}$ and $\Delta\varepsilon_{HCF}$ are varied, which most commonly is presented as strain-life curves, see Figure 5a in Paper III as an example. Strain life curves are presented in all three papers (**Paper I-III**) and are very useful in a comparative analysis. However, on their own, the academic value is low since drawing strain life curves will only indicate the fatigue resistance without giving the physical explanations. Nevertheless, the strain life curves obtained for the tested LGI, CGI and SGI have allowed the investigation to answer the research question *ii*).

The research question *i*) was mainly addressed in **Paper I**, and partly in **Paper III**. The main results presented in the former concerned the fatigue process of microcracking, which has been treated in Chapter 3, and the supposed relation with the measurable quantity defined as the elastic modulus during partial unloading, *i.e.* the unloading modulus. This relationship, connecting the instantaneous value of the unloading modulus with the concurrent state of progressed microcracking, was anticipated by Haenny and Zambelli [28] and supported by the metallographic investigations presented in Paper I and III. Accordingly, it could be established in the two papers that all fatigue tests conducted, including both TMF and isothermal LCF tests on LGI, CGI and SGI, manifested an identical response in the percentage change of the unloading modulus for each cast iron type, see Figure 10 in

Paper III. Consequently, it is clearly indicated that the fatigue process can be generalised to include both LCF and TMF conditions. More importantly, this also suggests that the existence of a general model, capable of covering both fatigue modes with the same set of equations, is possible.

The main purpose in **Paper II**, and later also in **Paper III**, was the identification of the seldom mentioned fatigue property which relates the ability of a material to resist a superimposed HCF load, namely the *TMF-HCF threshold*. This should be an important property in all applications which are subjected to superimposed low-amplitude loads. The threshold is most easily visualised by drawing a plot with the number of cycles to failure as a function of the HCF strain range, with the temperature cycle and the total mechanical strain range fixed; denoted as a *TMF-HCF plot*. In this way, the TMF-HCF resistance of different cast irons could be conveniently compared, see Figure 7 in Paper III. Effectively, the studied SGI is observed to be the most resistant to superimposed HCF loading compared to CGI, followed by LGI. Equally important, the TMF-HCF threshold has been observed to be almost independent of the underlying TMF cycle, *i.e.* it does not depend on variables such as the maximum temperature and the total mechanical strain range, which is a convenient property. As a consequence, it is reasonable to believe that it could take on the same value for most TMF conditions, perhaps even for LCF conditions. In particular, it is possible that the threshold has a similar value in both in-phase and out-of-phase configurations.

An expression to estimate the lifetime of specimens subjected to TMF and TMF-HCF loading were developed in **Paper II** and **Paper III**; its final form presented in Equation 6 in Paper III. From this equation, a simple expression of the threshold value could be deduced, see Equation 8 in Paper III, from which its influence on other parameters is estimated. As expected, it is noted that a combination of a low elastic modulus and a low Paris exponent is favourable for a good TMF-HCF resistance. In addition, Equation 8 in Paper III also make an estimate of the HCF frequency dependence of the threshold. Thus, through this expression, it is possible to estimate the TMF-HCF threshold of other untested cast irons grades solely from static and TMF tests.

The proposed model mentioned above, which was developed to deal with the research question *iii*), considers the propagation of an average microcrack in an attempt to simulate the experimental observation made here and by other investigators. Furthermore, the model supposes that the average crack propagation rate follows Paris law, Equation 5, and that the coefficients and exponents associated with this relation are the same for the TMF cycle and all the HCF cycles. It is important to note that this is a highly simplified approach, since it is very likely that at least the Paris coefficient C should

Figure 10: Schematic illustration of the different crack propagation regimes in metallic materials.

have a temperature dependence. The advantage on the other hand, is the acquisition of analytical expressions which allow a convenient manipulation and comparison with experimental data, which has been shown to be more than acceptable for the tested LGI and CGI materials, see Figure 12 in Paper II and 6a in Paper III. However, in the case of the SGI material, the threshold is slightly underestimated, see Figure 6b in Paper III.

Some of the assumptions made have been verified by experiments in order to further justify the model. For instance, the effect of crack closure was investigated in Paper II and it was verified that an assumption of zero crack closure level is applicable. Furthermore, regarding the growth of microcracks, the average crack length was measured after different fractions of the total fatigue life to be compared with the crack lengths estimated by the model, see Figure 9 in Paper III. Even though there are experimental difficulties to obtain representative average crack lengths, the comparison affirms that the evolution of the average crack length is not far from the evolution predicted by a Paris law.

One of the most intriguing outcome of the modelling work has been the exceptionally high values of the Paris exponent, *i.e.* values around 9-13, which normally are found to be much lower for most materials, usually around 3-5. However, there are a couple of plausible reasons which could answer to this, the most likely being the following. The dominant part of the fatigue life of cast irons has been verified to consist of small fatigue crack growth until a point of coalescence where the crack lengths are believed to rapidly increase in size. By reference to the logarithmic plots typically obtained

for metallic materials, relating the stable fatigue crack growth rate with the stress-intensity factor range ΔK [1], one could argue that the fatigue life of cast irons is spent almost solely in regime A, see Figure 10. Thus, as the small cracks would increase in size, they are quickly linked-up leading to the final failure of the specimen, instead of entering the conventional Paris regime (regime B) normally seen in single-crack propagation. Consequently, fitting a Paris expression to regime A, *i.e.* a linear slope in Figure 10, therefore results in a more steep slope and thus a higher n -value than what would be obtained in regime B.

Theoretically, it is important to note that a higher n -value not necessarily will imply a higher crack growth rate. On the contrary, a high n -value is more likely to characterise a slow small crack growth regime by the following argument. By considering the simplest form of the Paris law, Equation 5, it is seen that the growth rate is increased more slowly with increasing ΔK , the higher the n -value is, provided that $C^{1/n}\Delta K$ is smaller than unity. This can be understood by comparison with the function $y = x^n$ which is increased more slowly with increasing x for higher n -values, when x is small and less than unity. Thus, it can be argued that the microcrack growth rate is the slowest in CGI compared to LGI and SGI, which also is suggested by the measured crack growth profiles in Figure 9 in Paper III, by considering the corresponding n -values.

In the case of a loaded component, the situation is believed to be slightly altered since different points within the component might be very differently loaded. In contrast to laboratory specimens where microcracks are homogeneously initiated due to the uniform stress state, it is possible that different points in a component have attained different states of progressed microcracking depending on the local stress state. Consequently, it is likely that the fatigue process of an engine component also includes a regime of *macrocrack* formation and propagation, which possibly governs a significant part of late life of the component. It should be emphasised that the main difference to microcrack growth, which only have been addressed so far, would be the explicit independence of the microstructure on the fatigue crack growth.

The transition from microcrack to macrocrack growth has been identified as an issue in lifetime assessment modelling, since models only addressing microcrack growth are believed to underestimate the total fatigue life of components by neglecting the extent of the macrocrack regime [62]. On the other hand, considering a regime of long crack growth should most likely be very complicated, since the macrocrack propagation rate probably will be dependent on how far microcracks in the surrounding materials have extended. This aspect is yet to be addressed both experimentally as well as theoretically, and is therefore considered a challenge for future investigations.

6

Outlook

The research questions in Chapter 1 have been addressed and discussed both here and in the appended papers. In particular, the emphasis has been given to the first research question *i)*, since this question is argued to be the most important, *i.e.* being the key for a better understanding of the two more technologically relevant research questions, *ii)* and *iii)*. To look forward, some thoughts on technological implications and interesting paths to go further with this research topic, are presented in this chapter.

It has been clearly demonstrated that small cracks are induced in cast irons during the typical load conditions encountered in heavy-vehicle engines. Likewise, it has been shown that these cracks are responsible for the final failure as they propagate and successively are linked-up. In contrast to most metals, where the free surface is the most prominent fatigue crack initiation site, the small cracks are initiated homogeneously within the material.

In view of these facts, it becomes clear that measures that could be done to enhance the service lifetime of engine component through surface treatments, such as removing notches, protective coatings, shot peening, laser treatment etc., may not be the most efficient. Even though many such methods may impede surface crack initiation, microcracks are still initiated beneath the surface and the service lifetime would only be fairly improved. Similarly, corrosion issues should not be of a major concern, as long as the growth rate of environmentally-assisted cracks starting from surfaces exposed to a corrosive environment, is less or in the same order as the growth rate of internal microcracks.

If one wishes to optimise the lifetime of cast iron engine components, one should rather focus on the graphite-matrix microstructure, which clearly governs the rate of microcrack initiation. For instance, as it has been shown in this work, casting defects in the matrix formed during solidification are

very likely to induce such cracks. Thus, a good aim would be to avoid having defects larger than the typical size of the graphite structure, even though this might imply quite a metallurgical challenge. Regarding the graphite structure itself, it does not come as a surprise that different graphite morphologies result in different crack initiation propensities. As it has been demonstrated, compacted graphite iron seems to be advantageous in this regard, and most likely also spheroidal graphite iron if the influence of defects is attenuated.

Post heat treatments, which in principle only alter the properties of the matrix, are another potential method to prolong the lifetime. Hypothetically, it is likely that the microcrack growth rate can be modified by changing the structure of the matrix, *i.e.* switching between pearlite, ferrite, martensite, bainite or any combination. Therefore, it is suggested that the effect of post heat treatments on microcrack growth should be investigated if one seeks to optimise the fatigue resistance without replacing cast iron for another material.

Regarding life estimation, the lifetime assessment model proposed here has given satisfying estimates; thereby fulfilling one of the main purposes of this thesis work. On the other hand, it is undoubtedly a simplification of the complex fatigue mechanism, and therefore, cannot answer to all possible phenomenological aspects imposed by the underlying physical processes. For instance, it is not investigated how well it deals with in-phase TMF or isothermal LCF conditions. In order to further generalise, it is believed that a deeper understanding of the nature of the microcrack growth is required. In particular, better knowledge about the local stress field adjacent to the graphite phase and how it affects the growth of a microcrack is important. For instance, it could bring light to the observed temperature dependence of the crack growth, which up to now, only is incorporated as an unpredictable temperature dependence in different model coefficients.

Looking further, the next step would be to consider life estimation of entire components based on the findings presented here. In this regard, there are two challenges that need to be dealt with. Firstly, the formulation of constitutive models capable of assessing the thermo-mechanical constitutive behaviour in cyclic loading is of uttermost importance, as motivated in Section 3.4. This is quite a challenge since many different deformation mechanisms come into play, such as plasticity and creep effects, not to mention the already complicated behaviour of cast irons briefly mentioned in Section 3.1. Secondly, as discussed in Chapter 5, there is a need to study the transition from a microcrack to a macrocrack regime and the extent of the latter. Capturing these two physical aspects of cast iron materials in models, *i.e.* the constitutive and long crack growth behaviour, is believed to be the key for more reliable lifetime assessment of heavy-vehicle engine components.

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Part II

Papers Included

Papers

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