Linköping Studies in Science and Technology, Dissertation No. 1894

Fatigue of Heavy-Vehicle Engine Materials

**Damage Mechanisms, Laboratory Experiments and
Life Estimation**

Viktor Norman

Division of Engineering Materials
Department of Management and Engineering
Linköping University, SE-581 83, Linköping, Sweden
www.liu.se

Linköping, March 2018

Opponent: Prof. Dr.-Ing. Tilmann Beck, Institute of Materials Science and Engineering, University of Kaiserslautern, Germany
Date: 10.15, March 16, 2018
Room: ACAS, Linköping University

Cover:
Fatigue of an internal combustion engine
Design and drawings by Viktor Norman

Printed by:
LiU-Tryck, Linköping, Sweden, 2018
ISBN 978-91-7685-390-0
ISSN 0345-7524

Distributed by:
Linköping University
Department of Management and Engineering
SE-581 83, Linköping, Sweden

© 2018 **Viktor Norman**

This document was prepared with L^AT_EX, February 7, 2018

Abstract

Due to increasing demands on sustainability exerted by end-costumers and policy makers, heavy-vehicle manufacturers are urged to increase the engine efficiency in order to reduce the exhaust gas emission. However, increasing the efficiency is also associated with an elevated fatigue rate of the materials constituting the engine parts, which consequently reduces the engine service life. The aim of the present thesis is therefore to confront the expected increase by studying the fatigue behaviour and damage mechanisms of the materials typically employed in heavy-vehicle diesel engines. With this knowledge, this work seeks to guide the development of new heavy-vehicle engine materials, as well as to develop improved life estimation methods designated to assist the mechanical design of durable heavy-vehicle engines.

In essence, a large set of thermo-mechanical fatigue (TMF) and combined thermo-mechanical and high-cycle fatigue (TMF-HCF) tests is conducted at engine load conditions on laboratory specimens of lamellar, compacted and spheroidal graphite iron. In this way, the fatigue performance and associated damage mechanisms are investigated. In particular, a new fatigue property is identified, the TMF-HCF threshold, which quantifies how resistant a material is to superimposed high-cycle fatigue.

The damage mechanism at low temperatures ($\lesssim 500^\circ C$) is confirmed to consist of the initiation, propagation and coalescence of numerous microcracks. Based on this, a successful fatigue life estimation model is formulated, allowing accurate estimations of TMF and TMF-HCF tests on smooth specimens, and TMF tests on notched specimens. In the latter case, the microcrack growth behaviour in non-uniform cyclic stress fields and its implications for life estimation are clarified. At elevated temperatures ($\gtrsim 500^\circ C$), surface oxidation is shown to govern the fatigue performance of cast iron grades intended for exhaust manifolds. It is observed that oxide intrusions are induced, from which surface fatigue cracks are initiated. Consequently, an optimal material at these conditions should have a low oxide growth rate and few casting defects at the surface, as these factors are found to stimulate the growth of intrusion.

Populärvetenskaplig sammanfattning

Men anledning av ökande krav på hållbarhet från slutkunder och beslutsfattare pressas tillverkare av tunga lastbilar att öka drivlinans verkningsgrad och minska utsläppen av luftföroreningar. För dieselmotorn som är den vanligaste motortypen i tunga lastbilar förväntas i dagsläget det största bidraget till att möta kraven vara att öka motorns verkningsgrad. Högre verkningsgrad innebär lägre bränsleförbrukning, vilket i sin tur står i direkt proportion till mängden utsläppt koldioxid. Denna eftertraktade förbättring åstadkoms generellt genom att öka kompressionsförhållandet, vilket medför en ökning i det maximala förbränningsstrycket och temperaturen i förbränningskammaren. Till följd av detta uppkommer även ett ökat slitage av motorkomponenterna, i tekniska sammanhang kallat utmattning, vilket ofrånkomligen begränsar motorns livslängd. Av denna anledning visar sig materialvalet sätta en övre gräns eftersom det valda motormaterialet måste uppfylla nya krav på slitstyrka samtidigt som det inte får leda till nämnvärt ökade framställnings- och produktionskostnader.

Avhandlingens övergripande mål är att möta den förväntade ökningen av materialutmattning genom att undersöka utmattningsegenskaperna hos de vanligaste konstruktionsmaterialen i motorer. Framförallt har målsättningen varit att öka förståelsen för de fysikaliska mekanismer som leder till materialutmattning under verklighetstroga lastfall. Förhoppningen med denna kunskap är att kunna vägleda utvecklingen av nya motormaterial samt att utveckla förbättrade metoder för livslängsuppskattning med syftet att underlätta designprocessen av nya slitstarka motorkomponenter.

I stora drag består arbetet av en omfattande serie med termomekanisk utmattningsprovning (TMF) och kombinerad termomekanisk och hög-cykel-utmattningsprovning (TMF-HCF). Proven har genomförts med belastningsförhållanden typiska i motorsammanhang på provstavar bestående av gjutjärn med lamellär, kompakt och nodulär grafitform. Tack vare denna provning har dessa tre grupper av gjutjärn blivit experimentellt jämförda och de tillhörande utmattningsmekanismerna dokumenterade. I synnerhet har en ny mekanisk egenskap identifierats, här kallad TMF-HCF-tröskeln, som kvanti-

fierar hur motståndskraftigt ett material är mot överlagrad hög-cykel-utmattning.

Vad gällande utmattningsmekanismerna vid låga temperaturer ($\gtrapprox 500^{\circ}\text{C}$) har det bekräftats att förloppet innefattar initiering, propagering och sammanlänkning av många små mikrosprickor. Tack vare dessa observationer har en livslängdsmodell utformats. Modellen har tillförlitligt kunnat uppskatta livslängden på släta provstavar i TMF- och TMF-HCF-provning samt på anvisade provstavar i TMF-provning. I det senare fallet som berör icke-uniforma lasttillstånd har även inverkan av mikrospricksförloppet på utmattningsegenskaperna och dess konsekvenser för livslängsuppskattning blivit klarlagda.

Vid höga temperaturer ($\gtrapprox 500^{\circ}\text{C}$) har det visats att utmattningsförfarandet i gjutjärn ämnade för avgassamlaren bör förknippas med en oxidationssprocess. Under dessa förhållanden bildas så kallade oxidationsintrång vid provmaterialets yta vilket leder till initiering av längre utmattningssprickor. Följaktligen har ett optimalt material för dessa lastförhållanden ett högt oxidationsmotstånd och ett lågt antal gjutdefekter vid ytan, eftersom dessa aspekter till synes bidrar till bildandet av oxidationsintrång.

Acknowledgement

The present thesis work was funded by the Swedish Governmental Agency for Innovation Systems (Vinnova), Scania CV AB, and the Swedish Foundation for Strategic Research (SSF), and are therefore greatly acknowledged.

I would like to express my gratitude to my main supervisor Johan Moverare, for encouraging my ideas and for always keeping his door open. Likewise, I want to thank all three supervisors, including co-supervising Peter Skoglund and Daniel Leidermark, for all the fruitful discussions, help with proof reading and suggested improvements to the paper manuscripts. In addition, I am very grateful for Peter's organisational effort to realise the funding for the second half of the project.

Many thanks are addressed to all project collaborators for their contributions and feedback. This includes the project group at Scania CV AB, namely Fredrik Wilberfors, Anders Tjernberg, Jessica Elfsberg, Lars Jacobsson, Daniel Bäckström, Madeleine Ekström, Baohua Zhu, Thommy Nilsson, Patrik Gustafsson, Ingegerd Annergren among other, as well as Falk Schoenfeld, Stefan Schmidt and Thomas Kempe from MAN Truck & Bus AG and Gaël Le Gigan, Roger Lundén, Tore V Verneresson and Johan Ahlström from Chalmers University of Technology. Special thanks are also addressed to Jessica Elfsberg for assisting in the chemical etching work and Patrik Härnman for his impeccable technical support in the mechanical testing lab.

During the course of the research underlying this thesis, I was enrolled in *Agora Materiae*, a multidisciplinary doctoral program at Linköping University, Sweden, which also have contributed to my scientific and professional progression.

I also want to express my appreciation to my colleague Mattias Calmunger for our collaborative work, from which Paper VI is a result. Similarly, all the eminent people at the engineering materials division are acknowledged for contributing to the best of working environments. Notably, Ingmari Hallkvist must be recognised in this regard for keeping this division together and for the invisible administrative work behind every Ph.D. student.

At last, I want to thank my family and friends for all their support and encouragement. In particular, my beloved wife and best friend Sara, and my dear son Valter, for their patience and loving support, especially during my last intense weeks as a Ph.D. student.

Viktor Norman
Linköping, February 2018

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 Metals Research* 29:1-2 (2016) 25-32.
- 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. *International Journal of Fatigue* 88 (2016) 121-131.
- IV. V. Norman, P. Skoglund, D. Leidermark and J. Moverare. Damage mechanisms in silicon-molybdenum cast irons subjected to thermo-mechanical fatigue. *International Journal of Fatigue* 99 (2017) 258-265.
- V. V. Norman, P. Skoglund, D. Leidermark and J. Moverare. The transition from micro- to macrocrack growth in compacted graphite iron subjected to thermo-mechanical fatigue. *Engineering Fracture Mechanics* 186 (2017) 268-282.
- VI. V. Norman and M. Calmunger. On the cyclic elastoplastic deformation behaviour of cast iron. *In manuscript*.

Own contribution to the included papers:

In the first five 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. Regarding the last paper, I have been the main contributor performing the

experimental work and manuscript writing in cooperation with Mattias Calmunger while performing the theoretical work single-handedly.

Papers not included in this thesis:

- VII. G. Gigan, V. Norman, J. Ahlström and T. Vernersson. Thermo-mechanical fatigue of grey cast iron brake discs for heavy vehicles. *Proceedings of the Institution of Mechanical Engineers, Part D: Journal of Automobile Engineering* (2015).

Contents

Abstract	iii
Populärvetenskaplig sammanfattning	v
Acknowledgement	vii
List of Papers	ix
Contents	xi
Abbreviation	xiii
 Part I Background and Theory	 xv
1 Introduction	1
1.1 Research questions and aims	3
1.2 Outline and scope of the thesis	4
2 Engine materials and load conditions	5
2.1 Engine materials	5
2.2 Engine load conditions	8
3 Fatigue damage mechanisms in cast iron	11
3.1 Damage mechanisms under monotonic load conditions	12
3.2 Effects of an elevated temperature	13
3.3 Damage mechanisms under cyclic load conditions	15
3.4 Fatigue life estimation	19
	xi

4	Deformation behaviour of cast iron	21
4.1	Micromechanisms	21
4.2	Constitutive modelling	22
5	Experimental and computational methods	25
5.1	Materials	25
5.2	Thermo-mechanical fatigue testing	25
5.3	Crack growth testing and modelling	27
5.4	Fatigue life estimation	28
5.5	Oxidation testing	29
5.6	Metallographic investigations	29
5.7	Digital image correlation	30
5.8	Micromechanical modelling	30
6	Discussion of appended papers	31
7	Outlook	35
	Bibliography	37
Part II	Included Papers	51
	Paper I: Damage evolution in compacted graphite iron during thermomechanical fatigue testing	55
	Paper II: Thermo-mechanical and superimposed high-cycle fatigue interactions in compacted graphite iron	65
	Paper III: The effect of superimposed high-cycle fatigue on thermo-mechanical fatigue in cast iron	77
	Paper IV: Damage mechanisms in silicon-molybdenum cast irons subjected to thermo-mechanical fatigue	91
	Paper V: The transition from micro- to macrocrack growth in compacted graphite iron subjected to thermo-mechanical fatigue	101
	Paper VI: On the cyclic elastoplastic deformation behaviour of cast iron	119

Abbreviation

CGI	Compacted graphite iron
DIC	Digital image correlation
FE	Finite element
HCF	High-cycle fatigue
IP	In-phase
LCF	Low-cycle fatigue
LGI	Lamellar graphite iron
OP	Out-of-phase
RVE	Representative volume element
SEM	Scanning electron microscopy
SGI	Spheroidal graphite iron
TMF	Thermo-mechanical fatigue
TMF-HCF	Combined thermo-mechanical and high-cycle fatigue

Part I

Background and Theory

1

Introduction

If only two words were to be selected to give the context of this 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 mechanical loads. 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 constitutes a significant part of the Swedish export, about 13 percent of the total exported value in 2016 [1], to which one sixth is contributed by the heavy-vehicle industry. Both globally and nationally, the need on heavy-vehicle transportation is extensive. For instance, 89 percent of the domestic cargo transported was made by heavy-vehicles in Sweden in 2014 [2] and the situation in other European countries is not much different. Meanwhile, the demands for sustainable and environmentally accepted transports exerted by end-costumers and policy makers are increasing. One example being the repeatedly tighten European emission standards which dictate the acceptable limits for exhaust emission. As a consequence, heavy-vehicle manufacturers have been urged to increase the engine efficiency which is motivated simply because the amount of carbon emission linearly scales with fuel consumption. Increasing the engine efficiency is generally achieved by increasing the compression ratio, which implies an increase in the maximum pressure and temperature in the combustion chamber. However, such power density amplification is also associated with an increased rate of material degradation due to the increased thermal and mechanical loads which consequently reduces the engine service life. Thus, there is an

emerging challenge for heavy-vehicle manufacturers to find material and design solutions from which high engine efficiency can be achieved without compromising the engine durability nor increasing the material or production costs noticeably.

There are at present two potential strategies to prolong the service life of engine components. Either the material is replaced with a more resistant substitute, or the component geometry is optimised in order to reduce the intensity of the thermo-mechanical loads at critical locations in the engine. The former solution is relatively straightforward, however is often limited by cost aspects and how compatible it is with the current production line.

The latter strategy is associated with the development of more sophisticated engine design methods, and in particular methods to evaluate any conceived design solution. Foremost, the reason for this has been the uncertainty in knowing how well a particular design solution qualifies until a prototype physically exists; a wait which can be very long, not to mention the additional prototype production cycles eventually required due to the iterative nature of the design process. However, thanks to the advent of modern computational power, this inconvenient situation is expected to be circumvented. Given reliable simulation tools to predict the engine lifetime, the dependence on prototype evaluation becomes less significant, since mistakes related to the physical behaviour of the components can be avoided already at the conceptual design level. Consequently, improved simulation methods are anticipated to reduce the time to market, thereby speeding up the development of the demanded high-efficiency engines of tomorrow.

This thesis will deal with both strategies and is therefore concerned with the aspects influencing the life of engine materials. More precisely, the material life will be referred to as the *fatigue life*, which essentially signifies how long the material can endure the loads to which it is exposed before it breaks. 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 [3]. 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 commercial computer tools available to assist the design process, such as computer-aided design (CAD) and computer-aided manufacture (CAM) software, finite element (FE) software to perform mechanical analyses and more. 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. In addition, since the fatigue process often is relatively unstudied in engine materials, it is also unclear how the different microstructural aspects influence the fatigue resistance of the material. Due to these reasons, there is a need to investigate and characterise the fatigue behaviour and damage mechanisms of the materials employed in heavy-vehicle engines. Firstly, this is motivated since such an investigation sets the foundation for the development of accurate fatigue life estimation tools, and secondly since the knowledge about how microstructural parameters and the fatigue behaviour are related can reveal how engine materials should be developed in order to maximise the fatigue performance.

1.1 Research questions and aims

The research questions are summarised by the following four:

- i) How is the material degraded? What are the physical damage 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?
- iv) What material and microstructural parameters govern the damage mechanisms? In other words, what characterises a material with optimal fatigue resistance under typical engine load conditions?

The above research questions were formulated in order to fulfil the aim of the thesis, namely to provide the heavy-vehicle industry with experimental findings and computational methods in order to assist the design of more durable engine components. Evidently, the strategy to answer these questions has been to perform extensive experimental investigations in order to have a rich physical idea of the fatigue processes. In this way, the key aspects influencing the fatigue behaviour and the damage mechanisms have been identified which have been exploited to propose new engine material candidates as well as to estimate the fatigue life under the relevant load conditions.

1.2 Outline and scope of the thesis

This thesis is divided into two parts, starting with the *Background and Theory* which will briefly present the technical context and the current state of the corresponding academic field. This part is based on the Licentiate thesis *Fatigue of engine materials - experimental analysis and life estimation* [4], which was presented in September 2015. However, the content has been significantly modified and extended since then. The second part, *Included papers*, consists of enclosed academic papers previously produced which constitute the academic contribution done within this research project.

After the present chapter, the second chapter will present the particular material groups of interest, namely different grades of cast iron, and the load condition experienced by the materials due to the operation of an internal combustion diesel engine. Subsequently, a review of the fatigue mechanisms and deformation behaviour of cast iron are presented in the third and fourth chapter while the experimental and computational methods are summarised in the fifth. The sixth chapter discusses the included papers, and in the seventh and final chapter, an outlook is presented to see what is beyond this thesis.

The scope of the thesis mainly includes the experimental 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 estimation of real engine components. Consequently, also the performed modelling work is presently only applied and validated 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 engine components, as they set the basic conditions governing how the constitutive materials are fatigued. For this reason, the present chapter will deal with the typically employed materials as well as the load conditions in the components of interest.

The construction of a heavy-vehicle four-stroke internal combustion engine is displayed in a simplified schematic illustration in Figure 1. The central function is to transmit momentum to the 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 piston.

The diesel engine components of interest in this thesis are the *engine block*, the *cylinder head* and the *exhaust manifold*, which are schematically illustrated in Figure 1. The engine block, or cylinder block, houses the reciprocating motion of the pistons and the recurrent fuel ignition. On top of the engine block, the cylinder head is located, in which fuel and exhaust gas are interchangeably transported in intake and exhaust channels through regulated valves. In the exhaust manifold, the exhaust gas is collected and discharged from the cylinder head into subsequent units, which can be a turbocharger, air filters or catalytic converters.

2.1 Engine materials

Traditionally, the engine components in Figure 1 are constructed using *cast iron* and *cast steel*, depending on the component, engine type and manufacturer [5–14]. Both mentioned material groups are metallic alloys defined by

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

their key chemical components iron and carbon, as well as silicon in the case of cast iron, and the characteristic method of manufacture through casting.

In the heavy-vehicle engine context, cast iron is more common than cast steel due to superior castability, *i.e.* the ability of being cast with high quality, as well as due to better thermal conductivity and lower material cost. The use of cast steel is therefore often restricted to conditions where excellent resistance to elevated temperatures is required, such as for exhaust manifolds in gasoline engines [5, 15]. However, since the main focus of this thesis will be on diesel engines, most attention will be paid on cast iron henceforth.

As a surprise to most people, cast iron 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 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 which consists of graphite particles embedded in a steel-like matrix, see the microstructures in Figure 2. In contrast to the graphite particles which always consist of a graphite phase containing only carbon, the surrounding matrix can consist of many different phases, 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 iron may be varied through the variation of the matrix phases, however more significantly, also by altering the graphite shape, see Figure 2.

Cast iron grades are often categorised according to their corresponding

Figure 2: Examples of typical matrix-graphite microstructures seen in cast iron, (a) lamellar graphite iron, EN-GJL-250 (b) compacted graphite iron, EN-GJV-400 and spheroidal graphite iron, (c) EN-GJS-SiMo5-1 and (d) SiMo1000.

graphite shape. The most common group of cast iron is the lamellar graphite iron (LGI), often called grey cast iron, in which the graphite particles are elongated and flake-like giving the appearance shown in Figure 2a. 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 [16]. On the contrary, in compacted, or vermicular, graphite iron (CGI) the graphite particles are coarser and rounder as seen in Figure 2b. As a consequence, the CGI grades have superior mechanical properties but inferior thermal conductivity. Even more superior regarding mechanical properties is the spheroidal graphite iron (SGI), also called nodular or ductile cast iron, which is associated with a microstructure consisting of spherically shaped graphite particles, see Figures 2c and 2d. As a result, SGI grades have high strength and ductility, but also reduced machinability, thermal conductivity and damping compared to LGI and CGI. Other groups of cast iron, which are outside the scope of this thesis, are white cast iron and malleable cast iron.

At present, different cast iron grades are used depending on the engine component and its typical service load conditions. The engine block and

cylinder head have conventionally been cast in LCI regardless of engine type [8, 10]. However recently, replacing this material with CGI has become increasingly more common [8, 10, 17–20], due to the higher demands on durability [8, 10]. Moreover, the exhaust manifold has to endure higher temperatures than the engine block and cylinder head, which is due to the higher heat absorption from the exhaust gas and the less efficient cooling, and is therefore often made of SGI or cast steels. For diesel engines, ferritic SGI is frequently used, such as different grades of high silicon-molybdenum SGI [5–7, 12, 15, 19, 21–23].

2.2 Engine load conditions

Generally, the extent of the fatigue life is significantly dependent on the intensity of the loads to which the material is exposed. In this context, a load carried by the material can be purely mechanical, *i.e.* when stressing the material due to the application of forces, or purely thermal, which is when a high temperature is applied under free thermal expansion conditions. Conversely, if the thermal expansion is mechanically impeded or amplified, then the load is said to be *thermo-mechanical*. An aggressive chemical environment, such as in the presence of fuel and exhaust gases, can also be regarded as a load due to its potential to reduce the fatigue life.

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, see the schematic illustration in Figure 3. Consequently, the material closest to the combustion chamber and exhaust channels will become hot and heat will be conducted to the far off located regions, resulting in thermal gradients 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 is thermo-mechanically loaded as hot exhaust gas recurrently enters the exhaust channels and heats up the component. In this way, the exhaust manifold is typically subjected to a maximum temperature of 700 to 1000°C [5, 7, 13, 21], while the engine block and cylinder head are subjected up to about 400 to 500°C [18], depending on the engine type and material.

Each time the engine is turned on and off the above course of events is repeated. Thus, the entire engine is heated up and cooled down repeatedly as the driver stops for red lights and stops for pauses. The time period elapsed between a state of inactivity to a steady state of full operation, and then back to a stabilised state of inactivity, defines a characteristic *load cycle*, often referred to as the start-operate-stop cycle. Typically, the length of this

Figure 3: Schematic illustration of the thermo-mechanical load conditions encountered in a material point close to the combustion chamber. The mechanical strain, as defined by Equation 1, is most likely negative because the thermal expansion is anticipated to be constrained.

cycle ranges from a couple of minutes to a couple of hours depending on the driving circumstances. Moreover, on top of the slow heating and cooling cycle, there is also an additional mechanical load resulting from the high-frequency back-and-forth motion of the pistons and the reoccurring pressure bursts. This load situation is often referred to as high-cycle fatigue (HCF) which can be significant for the fatigue life, even though the load amplitude is generally small. Thus, due to all effects combined, the material inside the engine components is cyclically exposed to complex thermo-mechanical loads as schematically illustrated in Figure 3.

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 the name is also used when referring to the associated standard fatigue test procedure, which will be described later in Section 5.2.

Introducing a varying temperature in a fatigue test complicates the deformation analysis since the thermal expansion is mixed up with the deformation caused by the applied forces. For this reason, *mechanical strain* is often used as a deformation measure rather than the observable strain, where the latter includes the thermal expansion while the former does not. In an uniaxial

Figure 4: Schematic illustration of how (a) the TMF strain and HCF strain component are added to engender (b) the total mechanical strain. The differences of the maximum and minimum strain values, 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 marked out.

case, the two strain measures are related as

$$\epsilon = \epsilon_{Th} + \epsilon_{Mech} \quad (1)$$

where ϵ is the observable strain measured by, for instance, an extensometer, ϵ_{Th} is the thermal strain representing the thermal expansion and ϵ_{Mech} is the mechanical strain. Thus, when dealing with thermo-mechanical loads, it is most convenient to relate the mechanical load with the mechanical strain since this strain measure is independent of the thermal load.

Furthermore, the mechanical strain is anticipated to include a superimposed high-frequent component due to the vibrations of the engine, as mentioned above and illustrated in Figure 3. Therefore, it is expected that the fatigue failure is caused by a combination of thermo-mechanical and high-cycle fatigue. For the same reason, Equation 1 is 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 heating and cooling cycle, and ϵ_{HCF} represents the strain oscillation relative to the TMF strain, see Figure 4. In this case when the mechanical strain ϵ_{Mech} is composed of two mechanical strain signals, it will be denoted as the *total mechanical strain*, in order to distinguish from thermo-mechanical loads without superimposed cycling.

3

Fatigue damage mechanisms in cast iron

As engine materials are subjected to cyclic thermo-mechanical loads, irreversible physical processes occur which render the material in a damaged condition. These physical fatigue mechanisms involve the initial accumulation of damage on a microscopic length scale, which eventually result in the initiation and subsequent propagation of *fatigue cracks* on a macroscopic length scale. Inevitably, continued thermo-mechanical cycling will maintain a steady increase in fatigue crack length, and in due time, cause complete failure of the considered engine component.

The intention of this chapter is to present what is currently known about the fatigue damage mechanisms in cast iron when subjected to cyclic thermo-mechanical load conditions. To this end, this chapter will start by reviewing associated mechanisms under simple conditions, namely when exposed to monotonic mechanical loads and at elevated temperature conditions in absence of mechanical loads. The subsequent section will deal with the damage mechanisms observed under cyclic load conditions, including isothermal, thermo-mechanical (TMF) and combined thermo-mechanical and high-cycle (TMF-HCF) fatigue test conditions. In the last section, a number of fatigue life estimation approaches based on the observed damage mechanisms are presented.

The following review is restricted to cast iron grades relevant to the heavy-vehicle engine context, namely lamellar (LGI), compacted (CGI) and spheroidal graphite iron (SGI) which were introduced in the previous chapter. Furthermore, the present study will mainly focus on ideal load conditions, *i.e.* uniaxial mechanical loads and spatially uniform temperature distributions.

3.1 Damage mechanisms under monotonic load conditions

Before dealing with the fatigue damage mechanisms associated with cyclic loads, it is of relevance to start with the damage processes occurring during monotonic loads, *i.e.* when the mechanical load is applied at a single and constant rate. These damage mechanisms are naturally related to the mechanisms observed under cyclic load conditions, simply because a cyclic test can be considered a periodic sequence of monotonic load segments.

In general, the initial damage processes in cast iron are highly related to the interaction between the graphite phase and the steel-like matrix mentioned in Chapter 2. As many authors have observed on polished surfaces of cast iron specimens at room temperature, small deformations result in the delamination, or *debonding*, of the matrix-graphite phase boundary [24–34]. More precisely, it has been reported that the tensile strain required for such damage initiation can be less than 0.03–0.04% for ferritic LGI and SGI [30, 31]. In addition, investigators have also observed cracking within the graphite phase in LGI, CGI and SGI as the first sign of damage [25, 27, 32, 35–39], resulting in *graphite cleavage*. The cleavage cracks have been reported to propagate along the basal plane of the hexagonal crystallographic structure of graphite [32, 35] or along boundaries between graphite crystallites [25, 27] in LGI and CGI. Moreover, in SGI, cracks have been demonstrated to propagate along layer boundaries generated by the different graphite growth stages during the casting solidification process [37, 39]. The two graphite failure modes are schematically illustrated and experimentally observed in Figure 5.

As the monotonic tensile load is increased to higher levels, the cracks start to enter the matrix at graphite tips oriented perpendicular to the load direction and eventually coalesce as multiple cracks propagate simultaneously [29, 31–34, 36, 40, 41], see Figure 5a. 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*. The subsequent growth and coalescence of such microcracks have been observed to be dependent on the cast iron type [29, 36]. In particular, microcrack coalescence was seen to result in the formation of a clearly visible dominant crack in SGI, while in LGI and CGI, such a dominant crack could not be definitively distinguished from the individual growth of microcracks. Clearly, crack growth is a microstructure-dependent and complex process in cast iron as it involves the evolution and interaction of microcrack networks.

Figure 5: (a) A schematic illustration of the different damage mechanisms occurring in cast iron; (i) matrix-graphite debonding (ii) graphite cleavage (iii) graphite-initiated microcracks (iv) microcrack coalescence, from [42] (Paper III). (b) SEM image demonstrating the damage processes in CGI due to a large tensile load. Matrix-graphite debonding and graphite cleavage are indicated with white arrows and the resulting large deformation of the adjacent matrix is indicated with black arrows.

3.2 Effects of an elevated temperature

Even in the absence of mechanical loads, the application of an elevated temperature during an extended period of time might have a substantial effect on the state of the material. Therefore, it is also appropriate to review the detrimental effects associated with elevated temperatures under free thermal expansion conditions before proceeding to situations including both high temperatures and mechanical loads.

When cast iron specimens are exposed to elevated temperatures for a long time in ambient air, they grow in size and experience surface oxidation, or *scaling* [19, 43–47]. 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 [19, 46, 47]. 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* [45, 48]. 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 [48]. Regarding the mentioned possible increase of volume, it is often associated with *matrix decomposition*, *i.e.* the structural breakdown

Figure 6: (a) Schematic illustration of processes at elevated temperatures; (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 a LGI due to a TMF test with a temperature cycle of 100-500°C and a TMF strain range of 0.35%.

of bainite and pearlite into ferrite and graphite [43, 47, 49]. The reason is the meta-stable nature of the cementite phase which 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 6a and an example of how the surface of a LGI material is affected by scaling and decarburisation can be seen in Figure 6b.

The kinetics of the scaling in cast iron 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 iron grades 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 [19, 44, 45, 50]. The scaling of the Fe-Si system, which arguably should be similar to the matrix in cast iron, was reviewed by Birk et al. [51] by reference to the work of Adachi and Meier [52]. According to these, the presence of silicon allows the formation of silica (SiO_2), which in turn may react with wüstite (FeO) to form fayalite (Fe_2SiO_4). If the silicon content is high enough, a continuous layer of fayalite, silica, or a combination of the

two, is formed [52] 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 [19, 45–47, 50]. 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 [19, 44, 53].

3.3 Damage mechanisms under cyclic load conditions

The fatigue damage mechanisms under cyclic isothermal and thermo-mechanical load conditions can favourably be categorised into mechanisms dominant at low and elevated temperatures. The underlying reason for this distinction is that surface oxidation and creep deformation, which are effects known to contribute to fatigue, only occur at elevated temperatures. The limit between the two temperature regimes is not completely distinct since it is likely to depend on multiple aspects, such as the material and the characteristics of the cyclic mechanical load. Nevertheless, based on a number of studies on SGI [21, 42, 53–57], the limit temperature is presumed to be in the range of 500–600 °C in cast iron.

Low temperatures

At generic cyclic load conditions at room temperature and up to 500 °C, microcracks have been observed to initiate at multiple graphite tips on polished samples of LGI, CGI and SGI, propagating transversally into the matrix [49, 58–66], including the work of Norman et al. [42, 67, 68] (Paper I, III and V), as similar to when under monotonic load conditions. An example of a graphite-initiated microcrack caused by thermo-mechanical cycling is shown in Figure 7a. Some authors [58, 59, 61] 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 is not surprising in view of the observed damage mechanisms occurring during monotonic load conditions, see Section 3.1, under which the graphite phase is likely to fracture already at low mechanical loads. Consequently, it can easily be argued that some graphite particles lose their load-carrying ability during the first load cycle which thereby generates microcrack initiation sites due to the resulting stress concentrations at the edges of the graphite particles.

In the case of SGI, it has been seen that the graphite phase is not the

Figure 7: Optical microscope image of a microcrack emanating from (a) a graphite particle in CGI and (b) a microshrinkage void in SGI.

only microcrack initiation site. Several investigators have reported of crack initiation at microshrinkage porosities [59, 69–72], 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 7b. It is however important to note that cracks not necessarily have to initiate at porosities, as graphite spheroids have been shown to be distinct crack initiation sites even in the presence of porosities [72]. 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 [71, 72].

As the material is subjected to an increasing number of load cycles, microcracks grow longer, which eventually results in the possible encounter of another graphite particle or microcrack. For this reason, the microcrack growth in LGI, CGI and SGI have been observed to progress incrementally, *i.e.* stepwise, as microcracks occasionally link-up [58, 59]. Typically, this growth mode ends with the failure of the specimen at a point when the final crack length is about 1 to 2mm, regardless of cast iron type. Similar fatigue crack growth involving microcrack coalescence has also been reported in the case of thermal fatigue [49, 63], as well as for TMF and TMF-HCF tests [65, 66]. Thus, regardless of the particular cyclic load condition employed, the fatigue process in cast iron seems to be following a general pattern consisting of a brief crack initiation phase and a subsequent crack propagation phase involving the growth and coalescence of multiple microcracks.

Norman et al. [42, 67] (Paper I and III) confirmed this general fatigue process using the concept of the unloading modulus, following the previously

mentioned hypothesis put forth by Hanney and Zambelli [73, 74]. It was observed that regardless of cast iron type and cyclic load conditions, including isothermal low-cycle fatigue (LCF), TMF and TMF-HCF, the unloading modulus always decreased linearly with the number of cycles until a consistent critical value was attained. Conversely, the only factors affecting the slope of this line were the load variables, *i.e.* the temperatures and mechanical strain cycles. Thus, it was concluded that the physical fatigue process is the same regardless of cast iron type and cyclic load conditions.

As a further generalisation, it was also demonstrated by Norman et al. [68] (Paper V) that the microcracking process also occurs in non-uniform cyclic stress fields. More precisely, microcracks were observed to initiate within a limited volume ahead of a notch in CGI specimens, resulting in a rapid microcrack coalescence event identified as the instant of macro-scale fatigue crack initiation. In other words, the damage process appears to be similar to smooth specimen testing, however progresses locally at different rates due to the point-wise difference in the maximum cyclic stress field.

Regarding TMF testing with superimposed HCF, it has been reported that the microcrack growth characteristic, including individual microcrack growth and subsequent coalescence, was unchanged when applying a superimposed HCF strain range to a TMF test of LGI, CGI and SGI [65, 75, 76]. More precisely, the measured crack length profile as a function of number of cycles, normalised to the number of cycles to failure, was demonstrated to be the same regardless of the value of the HCF strain range [75, 76]. These observations are in direct accordance with the general unloading modulus behaviour of cast iron mentioned above, *i.e.* that the only effect caused by the addition of a HCF strain range was a decrease in the negative slope of the linear unloading modulus curve [42, 67] (Paper I and III).

Elevated temperatures

As the temperature rises, surface oxidation and creep deformation become significant. This has been seen to alter the overall fatigue process in cast iron, which therefore will be the subject of this subsection. Moreover, since LGI and CGI are seldom used in applications at temperatures above 500 °C, below which these effects are negligible, the following review is restricted to SGI.

Under cyclic load conditions at elevated temperatures, fatigue cracks have been observed to be initiated at the surface through an oxidation-assisted process, in isothermal [50, 53, 55, 56], OP TMF [57, 77] (Paper IV) and IP TMF [77, 78] tests. This process is much similar to what has been observed in high temperature LCF of stainless steels [79–81], for which it has been

Figure 8: Oxide intrusion in SGI due to OP-TMF cycling with 300-750 $^{\circ}\text{C}$ at (a) 250 cycles and (b) 500 cycles. The images are acquired from two different specimens.

demonstrated that small cracks, often denoted as *oxide intrusions* [57, 82, 83], are initiated through the fracture of the oxide scale formed due to the high temperature oxidation. These oxide intrusions extend circumferentially for smooth cylindrical specimens and produce oxide ridges above and beneath the crack mouth; resulting in a resemblance to furrows of a ploughed field on the specimen surface [79, 80]. As a consequence, the oxide intrusions are capable of becoming prominent fatigue crack initiation sites as they penetrate the surface through repeated cycling. Examples of oxide intrusions in a SGI at different progress states are given in Figure 8. Similarly, oxides intrusions have also been reported in CGI in a low temperature regime [46, 67], see Figure 6b. However, it was not seen to be the main cause for fatigue failure [67] (Paper I).

Besides oxidation-assisted crack initiation, it is also expected that damage mechanisms seen under creep test conditions can be present in TMF tests at elevated temperatures on SGI [56, 78]. The creep damage mechanisms generally involve grain boundary sliding and the formation of grain boundary cavities [3, 84], which also have been reported for an austenitic SGI subjected to TMF tests [78]. On the other hand, in a study on ferritic and austenitic SGI subjected to high-temperature creep deformation, it has rather been observed that the creep damage processes are more similar to the processes under monotonic load conditions involving microcracking at graphite spheroids and carbides [84]. Thus at present, the importance of creep damage and its effect on fatigue are not completely clear in SGI and are therefore subjects of further investigation.

The fatigue mechanisms under TMF-HCF test conditions at elevated temperatures have only been scarcely investigated. Nevertheless, it has been

demonstrated that the HCF cycling does not accelerate the oxide intrusion growth significantly in high silicon-molybdenum SGI [57] (Paper IV). Instead, the reduction in number of cycles to failure caused by HCF cycling is due to the increased growth of microcracks initiated at graphite spheroids within the interior of the specimen. Thus, the dominating fatigue mechanism is not only distinguished by temperature regime, but can also be influenced by the type of mechanical loads employed.

3.4 Fatigue life estimation

Fatigue life estimation, or lifetime assessment, is about predicting the number of cycles to failure using information acquired from an isothermal or a thermo-mechanical load cycle, for example the cyclic maximum stress or the plastic strain range. The predictions are often based on fracture mechanics models consisting of a growing fatigue crack for which there is a critical crack length corresponding to the end of the fatigue life. Such an 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* [3]. Both approaches have previously been applied to cast iron in engine applications, however, this section will mainly focus on the former type of models, as briefly presented below.

With the microcracking process in mind, Metzger et al. [65] formulated a model for TMF and TMF-HCF tests following the work done by Seifert and Riedel [85] and Seifert et al. [86], where it was assumed that the crack growth rate 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 material constant. 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

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

Alternatively, Ghodrati et al. [64] successfully used Paris' law in order to estimate the fatigue life of notched CGI specimens subjected to TMF tests,

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

where C and n are material constants. Similarly, the model presented by Norman et al. [87] (Paper II) was also based on Paris' law, Equation 5, combined with Equation 4 to account for a superimposed HCF load. In contrast to previous models, the crack length was interpreted as an average of a 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.

Typically, the above models are reformulated into terms of variables deducible from hysteresis loops, such as the stress range $\Delta\sigma$ or the plastic strain range $\Delta\varepsilon_p$, in order to be fitted to TMF test series on smooth specimens. However, as remarked by Norman et al. [68] (Paper V), there are two important aspects when such fitted models are transferred to engine components. Firstly, the microcracking process occurs over an extended volume, which infers that the calculated stress field and resulting fatigue failure should not be investigated point-by-point. Rather, an averaged stress measure is suggested, for which the stress is averaged over a volume corresponding to the characteristic length of microstructural variation. Secondly, it is identified that in the transition from the independent growth of microcracks to the initiation of a macroscopically-large crack, there is an associated crack retardation effect due to heterogeneous microcrack density resulting from the non-uniform stress field. If the two aspects are not taken into consideration, the life estimation will likely be severely underestimated, as demonstrated on notched specimens in [68] (Paper V).

As seen in the previous section, additional damage mechanisms other than the microcracking process become active at higher temperatures in cast iron, namely oxide intrusion formation [57] and creep damage [56, 78]. A general approach to include these effects has been to anticipate a linear summation of the respective damage mechanisms [56, 82, 88–91]

$$\frac{dD_{tot}}{dN} = \frac{dD_{fat}}{dN} + \frac{dD_{ox}}{dN} + \frac{dD_{creep}}{dN} \quad (6)$$

where D_i is a generic damage parameter for each present damage mechanism. Accordingly, the number of cycles to failure is defined as the cycle at which D_{tot} reaches a critical value. Formulations of the damage parameters intended for thermo-mechanical cycling of SGI have been proposed by Wu et al. [56, 91].

4

Deformation behaviour of cast iron

The constitutive behaviour of materials, *i.e.* the thermo-mechanical properties such as for instance the elastic modulus, yield strength and their temperature dependence, are a prerequisite for fatigue life estimation as they constitute the basis of stress-strain calculation of mechanically loaded geometries. In practice, the reason is simply that the fatigue life cannot be estimated if the stress-strain state is unknown since this is an input variable in most fatigue life estimation models. For this reason, the present chapter is devoted to the deformation behaviour of cast iron and how it can be modelled for fatigue life estimation.

4.1 Micromechanisms

In contrast to most conventionally used metallic alloys, cast iron has been demonstrated to exhibit non-standard mechanical behaviour under cyclic load conditions. This involves (i) an absence of a linear-elastic regime in monotonic tension, (ii) tension-compression stress asymmetry, (iii) varying elastic modulus, and (iv) an inflection in the tension-to-compression hardening curve [73, 74, 92]. As an example, Figure 9 illustrates the listed phenomena in a LGI.

The macromechanical behaviour is believed to be related to micromechanisms occurring in the characteristic microstructure of cast iron, namely the interaction between the graphite phase and surrounding steel-like matrix. In fact, many investigators have reported observations of severe deformation of the adjacent matrix under monotonic load conditions at room temperature [29, 34, 36, 93–97], which also can be seen in Figure 5b. The local matrix deformation is intuitively explained by the stress concentrations generated

Figure 9: Illustration of the general mechanical behaviour of cast iron, represented by a LGI, involving (i) an absence of a linear-elastic regime in monotonic tension, (ii) tension-compression stress asymmetry, (iii) varying elastic modulus, and (iv) an inflection in the tension-to-compression hardening curve.

either due to the significant difference in stiffness of the two constituents, or more likely, due to the failure of the graphite phase discussed in Section 3.1. Starting with the hypotheses proposed by Gilbert [24, 92, 98], such local plastic flow in combination with graphite opening in tension, have for a long time been assumed to be the underlying reasons for the mentioned complicated mechanical behaviour. More precisely, due to successive debonding or internal fracture, the graphite particles are anticipated to partially open in tension. This is presumably the origin of the decreasing elastic modulus (iii) and the origin of non-linear elastic behaviour (i) in monotonic tensile tests. Similarly, when going from tension to compression, the fracture and debonded surfaces supposedly recontact resulting in the recovery of the elastic modulus (iii), which in turn can account for the stress asymmetry (ii) and the inflection (iv). Thus, the process of opening and closing the graphite phase under cyclic load conditions intuitively explains the mentioned anomalies (i)-(iv), as also supported by micromechanical modelling (Paper VI).

4.2 Constitutive modelling

Due to the complicated mechanical behaviour, most standard constitutive models fail to accurately predict the constitutive behaviour of cast iron. For this reason, there has been an intense development of specialised models intended for cast iron over the past years. In most cases, these models can be categorised either as phenomenological or fundamental, whether if the constitutive behaviour is modelled based on experimental observations of the mechanical phenomena or derived from principles postulated at a more

fundamental level, *e.g.* such as deriving a model from the microstructural behaviour of the material.

Most phenomenological constitutive models are based on the foundation set out by classical plasticity [99] and thermodynamic theory [100]. However, as this traditional approach does not naturally infer an asymmetric behaviour seen in cast iron, a number of propositions within the theoretical framework have been suggested. A common approach has been to employ the Gurson model [101], later modified by Tvergaard and Needleman [102, 103], since this model incorporates the opening and closing of voids in a porous medium, in accordance with the anticipated behaviour of cast iron [31, 86, 104–108]. The model is derived from a rigid-plastic limit analysis of a solid approximated by a homogeneous spherical body containing a spherical void occupying a volume fraction of c_v of the whole solid. In this derivation, the global yield function f of this unit-cell becomes

$$f = \left(\frac{\bar{\sigma}_e}{\sigma_y} \right)^2 + 2q_1c_v \cosh \frac{3q_2\bar{\sigma}_h}{2\sigma_y} - (1 + q_3c_v^2) \quad (7)$$

where $\bar{\sigma}_e$ is the global von Mises stress, σ_y is the yield strength of the void-surrounding medium, $\bar{\sigma}_h$ is the global hydrostatic stress, and q_1 , q_2 and q_3 are constants.

Even though the Gurson-Tvergaard-Needleman model can be argued to be based on micromechanical concepts, the above yield function has often been used and extended within a phenomenological theoretical framework, *i.e.* in combination with classical plasticity theory. Effectively, the model must be associated with a hardening law and flow rule, *i.e.* equations dictating the plastic flow and hardening behaviour, as well as an equation describing the evolution of the porosity c_v , in order to be useful as a constitutive model.

Other examples of phenomenological constitutive models applied to cast iron are empirical approaches within classical plasticity theory, such as for instance the Drucker-Prager yield function [109–111], semi-empirical modelling approaches [112], state variable modelling approaches [113] and multilayer model approaches [114].

Regarding fundamental models, micromechanical modelling of cast iron has become increasingly more common. In most cases, the models are implemented in order to assist investigations on microstructural phenomena, such as void growth and local plastic deformation, or to predict macroscopic properties, *e.g.* the elastic modulus or yield strength, based on micromechanical considerations.

The micromechanical modelling approach relies on the concept of representative volume element (RVE), which is an envisioned small-scale volume representing the microstructure and its associated mechanical behaviour. In the case of cast iron, a suitable RVE typically consists of graphite particles embedded in a uniform matrix. Provided such a RVE, as well as suitable boundary conditions and anticipated constitutive properties of each constituent in the RVE, both the microstructural and the overall mechanical behaviour of the RVE are available. The latter is assessed by computing global or macroscopic load variables, typically defined as the volume averages of the stress and strain variation over the RVE. In this way, a global stress-strain behaviour can be derived which can be used to assess the macroscopic constitutive behaviour.

Different approaches to select suitable RVE for cast iron materials have been considered and investigated in the past. Regarding SGI, the most common way has been to use a unit-cell approach due to the favourable symmetry of a spheroidal graphite particle [31, 105, 106, 115–120]. In this approach, the RVE is modelled as a cubic or cylindrical cell in which a spherical graphite particle or void is situated. Generally, the relative size of the particle or void with respect to the cell is ascribed such that the experimentally graphite volume fraction is reflected.

A similar kind of approach was attempted on graphite shapes other than the spherical in a couple of investigations [121, 122], including the work of Norman and Calmunger (Paper VI). In this case, symmetry can not be exploited in the same extent as with spherical particles and the choice of the cell shape becomes complicated since elongated graphite shapes might not fit into a simply shaped unit-cell for a given graphite volume fraction. For these reasons, more sophisticated cell designs and boundary conditions were implemented. In addition, the approach also involved a methodology to account for the effect of the angular distribution of the graphite particle orientation with respect to the load direction, which arises due to the reduced symmetry of the particle shape.

An alternative approach to the unit-cell approach is to model a two-dimensional representation of the microstructure obtained either conceptually or through optical or scanning electron microscopy [32, 34, 97, 123–126]. The advantage is the possibility of directly comparing the model with experimental observations, such as strain measurements using digital image correlation (DIC) [34, 97, 126]. On the other hand, the models are often insufficient due to the negligence of the underlying bulk material. As a consequence, there have recently been a few investigations conducted to study three-dimensional RVE generated experimentally [127], or through statistical modelling of the spatial distribution of graphite particles [128].

5

Experimental and computational methods

In the present chapter, the materials and methods employed in order to answers the previously formulated research questions, will be described. For brevity, many details are excluded here but can be found in the corresponding research paper in Part II.

5.1 Materials

In the present work, different cast iron grades employed in different heavy-vehicle engine components have been investigated. The tested materials included one non-commercial lamellar grade, EN-GJL-250, with a composition optimised for high-temperature properties and one compacted grade, namely a commercial pearlitic EN-GJV-400. In addition, a commercial ferritic high silicon-molybdenum EN-GJS-SiMo5-1 alloy and a newly developed silicon-molybdenum cast iron with aluminium, SiMo1000 [129], have been investigated within this work. The typical microstructures of these cast iron grades were shown in Figure 2 in Chapter 2. The EN-GJL-250 and EN-GJV-400 grade are typically used in the engine block and cylinder head while the high silicon-molybdenum alloys are associated with the exhaust manifold.

5.2 Thermo-mechanical fatigue testing

The concept of thermo-mechanical fatigue 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

Figure 10: (a) TMF test set-up and (b) schematic illustration of the load cycle employed in the TMF and TMF-HCF tests.

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 and thereby studying the response of the material.

A standard TMF test is performed in a uniaxial configuration on axisymmetric elongated specimens, where the deformation and load force are applied in the axial direction. The equipment used to conduct these tests was an Instron 8801 servo hydraulic test machine, which is displayed in Figure 10a, and the associated test control software provided by Instron. During a test, the test specimen is subjected to a prescribed periodic temperature and mechanical load cycle, see the example in Figure 10b. Conventionally, the cycle length of the thermal and mechanical cycle are the same, however the phase angle between the two is often 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. Irrespective of the cycle type applied, the desired output of TMF tests is the number of load cycles required to fatigue the specimen and its dependence on the different load variables such as the maximum temperature and mechanical strain range.

Two different OP TMF cycles were used in this thesis work; one motivated by the cylinder head operation cycle and the other by the exhaust manifold operation cycle. Both cycles consisted of periods of ramping up and down in temperature, as well as hold times at each load reversal point, during which

Table 1: Presentation of the two employed OP TMF cycles corresponding to the load situation of two components of interest, including the cycle time, temperatures and materials.

Component	Cycle time	Minimum temperature	Maximum temperature	Materials
Cylinder head	200s ramp-up 25s hold 200s ramp-down 25s hold	100°C	400°C, 450°C, 500°C	EN-GJL-250, EN-GJV-400, EN-GJS-SiMo5-1
Exhaust manifold	150s ramp-up 90s hold 170s ramp-down 10s hold	300°C	750°C	EN-GJS-SiMo5-1 SiMo1000

the temperature and mechanical strain were held constant. Details about the tested cycles including cycle times, minimum and maximum temperatures and the tested materials are given in Table 1. In fact, the selected cycles are motivated by their resemblance to some of the component tests regularly performed on engine components. However, it should be noted that the cycles are significantly shortened compared to the real load cycles since it would require unreasonable long test times otherwise.

A TMF test is easily transformed into a TMF-HCF test by adding a HCF strain range, as presented in Figure 4 in Chapter 2. However importantly, the maximum and minimum *total* mechanical strain values, ε_{Max} and ε_{Min} , were always kept unchanged when performing a test series with increasing HCF strain ranges, not to change the R_e -value, *i.e.* the $\varepsilon_{Min}/\varepsilon_{Max}$ ratio. Comparisons between tests of different HCF strain ranges would otherwise have been ambiguous, as discussed in Paper II. Regarding the HCF frequency, it was given a fixed value of 15Hz since it is a reasonable value corresponding to the stroke frequency of a heavy-vehicle diesel engine.

5.3 Crack growth testing and modelling

Generally, TMF testing of smooth test specimens does not fully reflect the entire fatigue life of engine components since the material in the component is not uniformly loaded. Rather, the material fails locally at critical locations, which often results in the generation of a fatigue crack whose growth rate may have a substantial influence on the overall fatigue performance. For this reason, it was also of interest to study the crack growth characteristics under thermo-mechanical load conditions through a crack growth measurement

procedure, see Paper V.

For these tests, the same experimental set-up was used as in the TMF tests described in Section 5.2, except of the employment of a notched specimen design instead of the smooth cylindrical specimens. The added notch acts as a stress concentration from which a steadily growing fatigue crack is generated. For this purpose, the specimens were designed with a shallow circular notch with a radius of $1mm$ and a notch depth of $0.5mm$. These specimens were subjected to the cylinder head load cycle, see Table 1, and monitored by a camera mounted to view the notch from the side. The captured images were used to measure the crack length and the strain field at the notch using digital image correlation, see Section 5.7.

The crack growth rate is typically controlled by the current load condition at the crack tip which often is related to different fracture mechanics parameters such as the stress-intensity factor [3]. The reason is simply that the growth rate is empirically observed to be well correlated to such parameters using a power law, or Paris' law as often denoted in this context. To comply to this theory, the notched specimen was modelled using the finite element (FE) software Abaqus according to the procedure described in Paper V. In this way, a couple of selected fracture mechanics parameters, namely the stress-intensity factor (K_I), the cyclic J-integral (ΔJ) and the crack-tip opening displacement ($CTOD$), and their dependence on the crack length were calculated in order to correlate them with the crack growth rate.

5.4 Fatigue life estimation

Life estimation of TMF and TMF-HCF tests on smooth specimens and estimation of the number of cycles to crack initiation in TMF tests on notched specimens described above, were carried out using a model developed within in this thesis work, see Paper II, III and V. The model is based on the experimental observation of the growth of numerous microstructurally-small cracks and the assumption that the average microcrack length increases according to Paris' law. Under these premises, the following equation was derived to assess the fatigue life in TMF and TMF-HCF tests performed below $500^\circ C$, see Paper II,

$$D \int_0^{N_f} [\sigma_{Max}(N) - \sigma_{op}]^n dN = 1 - D\rho[E\Delta\varepsilon_{HCF}]^n N_f \quad (8)$$

where D and n are material constants, N_f is the number of cycles to failure, σ_{Max} is the maximum cyclic stress as a function the number of cycles

N , σ_{op} is the microcrack opening stress level, ρ is the number of HCF cycles applied at a stress larger than σ_{op} , E is the elastic modulus and $\Delta\varepsilon_{HCF}$ is the HCF strain range.

5.5 Oxidation testing

As described in Section 3.3, oxidation has a significant influence on the materials selected for the exhaust manifold. Therefore, a number of oxidation tests were conducted in order to assess and compare the oxidation resistance of the investigated exhaust manifold materials, see Paper IV.

The tests were conducted in accordance with the ISO 21608:2012(E) standard by discontinuously measuring the mass gain of small disc-shaped specimens under high-temperature corrosion conditions in laboratory air. More precisely, the specimens were subjected to a temperature of 750°C and recurrently weighted outside the furnace at time intervals of one to three hours.

5.6 Metallographic investigations

Since this thesis focuses on the microstructural influence on the fatigue performance of engine materials, it is of relevance to visualise the microstructure and the microscopic damage in order to support the hypotheses formulated. Such visualisation has been done using a Nikon Optiphot optical microscopy and a scanning electron microscope (SEM), namely a HITACHI SU-70 field emission gun. Furthermore, additional analyses have been conducted to characterise the microstructure of the considered materials. This includes systematic measurements of the graphite morphology, *i.e.* the size and shape distributions, using the image process software Axiovision, manual measurements of the distribution of casting defect sizes and chemical etching to visualise and measure the eutectic cell sizes.

As mentioned previously, interrupted TMF tests were conducted in order to measure the average microcrack and oxide intrusion length 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 while the intrusion lengths were assessed by measuring the average depth of oxidation penetration following the ISO 26146:2012(E) standard, see Paper III and IV.

5.7 Digital image correlation

Digital image correlation (DIC) is an image analysis technique which correlates the position in a sequence of images and thereby allows the acquisition of experimentally measured displacement and strain fields [130]. This technique was essentially used for two purposes; to validate FE calculations and to measure microscopically-small deformations in SEM images. The software used for this was an open-source Matlab code written by Eberl et al. and distributed by mathworks [131].

Prior to the instant of crack initiation, image correlation was made on the images captured in the crack growth tests presented in Section 5.3. The images were taken at the point of maximum tensile load and the acquired strain field was compared with FE calculations, see Paper V.

As mentioned, DIC was also employed to measure microscopically-small deformations in SEM images of EN-GJV-400 following a specially-designed test procedure, see Paper VI. In essence, this procedure consisted of a uniaxial tensile test followed by subsequent microstructural investigation of the unloaded condition to measure the residual strain field close to graphite particles caused by the heterogeneous plastic straining of the microstructure. To this end, image correlation of images captured with the above mentioned SEM equipment between loading and unloading a miniature dogbone specimen was carried out according to the procedure outlined in Paper VI.

5.8 Micromechanical modelling

In order to gain a better understanding of the cyclic mechanical behaviour of cast iron, which was reviewed in Chapter 4, a micromechanical modelling investigation was performed using Abaqus, as presented in Paper VI.

The modelling work was based on a unit-cell approach in which the graphite phase was modelled as disc-shaped ellipsoids consisting of a transversally isotropic medium, while the matrix was considered as a standard isotropic elastoplastic medium. The particular behaviour of the former was motivated by the anisotropic behaviour of pure graphite originating from its crystallographic structure. Moreover, since the microstructure is anticipated to consist of randomly oriented graphite ellipsoids with reference to the load axis, the variation seen over multiple unit-cells implemented with different ellipsoids orientations was considered. In this way, the macroscopic constitutive behaviour was defined as the volume average of the stress and strain in each unit-cell, and then again averaged over the angular distribution of the possible graphite ellipsoid orientations with respect to the load direction.

6

Discussion of appended papers

Part II includes the scientific papers produced within this thesis work and contains all the obtained results and drawn conclusions. In this chapter, these results will be summarised and collectively discussed with respect to the research questions formulated in Chapter 1.

i) How is the material degraded? What are the physical damage mechanisms responsible for the successive deterioration towards a state beyond operation ability?

This research question has been considered the most fundamental and a large effort has therefore been put into understanding the microscopic fatigue processes in cast iron, which were reviewed in Chapter 3. One key finding resulting from this endeavour, was the identification and distinction of two fatigue damage regimes demarcated by a critical temperature limit. Beneath this temperature, the fatigue process consists of the growth of microstructurally-small fatigue cracks initiated at naturally occurring initiation sites such as from graphite phase particles and casting defects (Paper I and III). However at elevated temperatures, surface scaling by oxidation becomes of relevance which in combination with the mechanical load cycling, assists the formation of surface oxide intrusions (Paper IV). These intrusions were seen to act as the starting point for fatigue cracks responsible for the final failure of the test specimens. By the experiments conducted in this work and by others [21, 53–56], the limit temperature separating these two regimes is expected to be in the range of 500-600°C for cast iron. This particular temperature range is likely due to the thermodynamic stability of the iron oxide phase wüstite, which causes high scale growth rates in iron-based alloys; however the oxide is not stable below 570°C [51].

Through the investigations carried out within this thesis work, the universality of the fatigue process in the low-temperature regime has been demonstrated. It was experimentally observed that regardless of the load conditions, whether it was isothermal or thermo-mechanical with or without superimposed high-frequent cycling, there was a consistent decline in the stiffness of the test specimen (Paper I and III). Thus, since the progressive growth of microcracks can be directly related to the observed decrease in stiffness, it is clear that the physical damage process is general. Moreover, it was also seen that the fatigue process was similar when the test materials were subjected to non-uniform cyclic thermo-mechanical load conditions (Paper V). As similar to uniform load conditions, microcracks emerged and coalesced, however only in the region subjected to the highest cyclic stress. In contrast to smooth specimens where coalescence coincided with the specimen failure, this test condition resulted in the initiation of a microstructurally-large crack whose growth characteristics were different compared to microcracks. In conclusion, these laboratory observations imply that the deterioration of engine components operating below 500°C is very likely related to the growth of microcracks.

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?

This research question was mainly addressed by performing a large set of fatigue tests, which consisted of measuring the variation in the fatigue life of test specimens when varying the load variables, *e.g.* the maximum temperature or the mechanical strain range. The obtained data has been presented as strain-life curves found in the first four papers (Paper I-IV). For instance, the measured dependence of the maximum temperature and mechanical strain range on the fatigue life of EN-GJV-400 is seen in Figure 8a in Paper II.

Much attention was given to how the fatigue life is affected by a superimposed HCF strain range. This influence was studied by performing TMF-HCF test series for which the HCF strain range was varied but the temperature cycle and total mechanical strain range were held constant, as explained in Chapter 5. By drawing a plot with the number of cycles to failure obtained from these experiments as a function of the HCF strain range, it was observed that the resulting curve always had a step-like appearance (Paper II-IV). This can be seen for instance in Figure 8 in Paper III where the TMF-HCF results of EN-GJL-250, EN-GJV-400 and EN-GJS-SiMo5-1 are compared. The critical HCF strain range at which the fatigue life abruptly decreases is clearly an important limit, and was therefore identified as a new material property denoted as the *TMF-HCF threshold*. For engine manufac-

turers, knowing this limit is of essence as it sets a natural tolerance limit to the superimposed mechanical load.

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?

In view of the experimental observations of the fatigue damage mechanisms mentioned above, it is argued that fatigue life estimation is suitably based on fracture mechanics models, which were reviewed in Section 3.4. To demonstrate this, a model developed within this thesis work was proposed, which considers the growth of the average microcrack in an attempt to simulate the observed mechanisms (Paper II). For the model, it was also assumed that the average microcrack growth rate follows Paris' law, Equation 5, and that the coefficients and exponents associated with this relation are the same for the TMF cycle and HCF cycles. It is important to note that this is a highly simplified approach, however it has the advantage of resulting in simple analytical expressions which allow a convenient manipulation and comparison with experimental data. Despite the considered simplifications, the model was shown to accurately fit TMF tests below 500°C (Paper III), as well as to be able to predict TMF-HCF tests in the same temperature range when only fitted to TMF data (Paper II and III). Furthermore, the model was also successfully used to estimate the number of cycles for macroscopically-large crack initiation in TMF tests of notched EN-GJV-400 (Paper V).

Some important aspects were identified regarding life estimation of non-uniformly loaded structures. Firstly, it must be well understood that the microcrack process progresses within an extended volume with sufficient microstructural variation. In addition, it must also be noted that the domain of validity of a model based on this fatigue process is exceeded as soon microcracks start to coalesce. For these reasons, it was shown that life estimation was improved if the stress value required as input to the model, was taken as the average value over a volume of a size comparable with the characteristic length of the microstructure; rather than an isolated point value (Paper V). Furthermore, it was also demonstrated that the model deviated significantly from experiments as soon as a microstructurally-large crack was generated through coalescence.

As pointed out in the beginning of Chapter 4, constitutive modelling is a prerequisite for life estimation. However, cast iron does not mechanically behave according to standard metallic models, which has been identified as an important issue. For this reason, an effort was put into understanding and modelling the constitutive behaviour of cast iron using a micromechanical approach, which was shown to successfully capture both the macroscopic

and microscopic cyclic behaviour at room temperature (Paper VI). In view of these results, it is thus argued that micromechanical modelling is a highly potential method to accurately model the thermo-mechanical constitutive behaviour of cast iron.

iv) What material and microstructural parameters govern the damage mechanisms? In other words, what characterises a material with optimal fatigue resistance under typical engine load conditions?

This research question is highly related to the first one. Clearly, at temperatures below 500°C , the graphite phase and casting defects appear to be the origin to the fatigue failure of cast iron. Furthermore, it was also demonstrated that SGI is superior to CGI and LGI when subjected to comparable maximum cyclic stress (Paper III), which suggests that a more compact graphite shape is favourable. This is intuitively understood, since a more elongated graphite shape results in higher microscopic stress concentrations at the edges and thereby a greater propensity to induce microcracks. In fact, no microcracks at all were observed at the graphite spheroids under these conditions in SGI. Rather, they were initiated at casting defects in contrast to LGI and CGI. Thus, an optimised cast iron in this regard should have a compact and finely dispersed graphite phase, and a low density of defects.

As the damage mechanism involving oxide intrusions becomes active at elevated temperatures, another property was found to influence the fatigue performance at these conditions, namely the oxide scale growth rate (Paper IV). A low scale growth rate was correlated to a slower growth of oxide intrusions, which were observed to govern the TMF life. For this reason, the recently developed high silicon-molybdenum cast iron SiMo1000 [129] was distinguished as a superior material owing to its low scale growth rate at elevated temperatures.

An useful outcome of the proposed model mentioned earlier was the possibility to generally predict the value of the TMF-HCF threshold from a simple expression requiring a number of input values, see Equation 9 in Paper III. The indications made by this expression is in agreement with what is intuitively expected and can therefore arguably be used to find materials which have optimised resistance to superimposed HCF. In this regard, an optimised material has a low elastic modulus and preferably also a high resistance to creep deformation, if the material is designed for OP thermo-mechanical cycling. The latter is motivated as follows; if the material significantly stress-relaxes during the compressive and hot part of the load cycle, it results in a shake-up of the hysteresis loop, see Figure 6 in Paper II. Consequently, a larger number of HCF cycles are applied at tensile stress levels which therefore reduces the threshold according to the threshold expression.

7

Outlook

Through this thesis work, the research questions given in Chapter 1 have been addressed and discussed, as summarised in Chapter 6. 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 the fatigue life of cast iron is controlled by microstructurally-small cracks when subjected to load conditions typically encountered in the engine block and cylinder head. In contrast to most metals, where the free surface is the most prominent fatigue crack initiation site, these microcracks are rather initiated homogeneously within the material. In view of these facts, it becomes clear that measures to enhance the service lifetime of these component through surface treatments, such as reducing surface roughness, 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.

If one wishes to prolong the lifetime, one should rather regard the matrix-graphite microstructure, which clearly governs the initiation and growth of microcracks. As discussed in the previous chapter, the results indicate that the more compact and finely dispersed the graphite phase is, the better is the fatigue performance. Thus, since the graphite structure is typically controlled by the melt composition and the cooling rate during solidification, optimisation of the metallurgical parameters appears to be a far better strategy to achieve this endeavour.

Post heat treatments, which in principle only alter the properties of the matrix, is another potential method to increase 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 also suggested that the effect of post heat treatments on microcrack growth should be investigated if one seeks to improve the fatigue performance without replacing the material.

For the fatigue performance of the exhaust manifold, the situation is different due to the higher operation temperature. In this case, it was demonstrated that the resistance to oxide scale growth has a significant effect on the fatigue life. For this reason, the investigated material SiMo1000 manifested a superior fatigue performance owing to the addition of aluminium in its composition, which is known to improve the high-temperature oxidation resistance of iron-based alloys [132]. Thus, in contrast to when at low temperatures, improvements are very likely to be achieved by altering the surface properties in exhaust manifold materials.

Another important remark is that this thesis work has mainly considered the intrinsic fatigue performance of the materials. In fact, it must be noted that other material properties affect the overall fatigue performance as well, for instance the thermal conductivity. The thermal conductivity of the material controls how efficiently heat is dissipated and thereby determines the maximum temperature, which severely affects the lifetime. Consequently, materials must be selected in regard to many of its different properties, such as the thermal expansion, thermal conductivity and elastic modulus etc. For this reason, it is emphasised that future research dealing with material development must include investigations on the eventual trade-off which may exist between the properties. Clearly, a gain in the intrinsic fatigue resistance has no benefit if it also would result in a more significant decrease in thermal conductivity.

Regarding life estimation, the model proposed here has given satisfying estimates; thereby fulfilling one of the main purposes of this thesis work. However, it was also demonstrated that there is a great potential in micromechanical modelling of simulating both the micro- and macromechanical behaviour of cast iron simultaneously. Therefore, it would be of great interest to investigate the combination of these two modelling approaches. The micromechanical modelling clearly demonstrates the importance of the stress-concentrations generated by the graphite phase for the constitutive behaviour. Similarly, this is also anticipated to be the origin of the microcrack initiation, in view of the experimental observations. Thus, it is believed that an improved and more general model can be acquired if the microcrack growth aspects are included in the micromechanical model. In particular, this can bring light to the observed temperature dependence of the microcrack growth, which up to now, only is incorporated as an unpredictable temperature dependence in different model coefficients.

Bibliography

- [1] Kommerskollegium. Sveriges utrikeshandel med varor och tjänster samt direktinvesteringar - Helåret 2016, 2017.
- [2] Trafikanalys. Godstransporter i Sverige - en nulägesanalys, Rapport 2016:7, 2016.
- [3] S. Suresh. *Fatigue of Materials*. Cambridge university press, Cambridge, 2nd edition, 1998.
- [4] Viktor Norman. *Fatigue of Heavy-Vehicle Engine Materials: Experimental Analysis and Life Estimation*. Licentiate thesis, Linköping university, 2015.
- [5] Peter Y P.Y. Chen, Richard Lyons, Michael Rakoczy, H. Yamanaka, and T. Mimata. Development of the 6.8L V10 heat resisting cast-steel exhaust manifold. In *SAE Technical Paper*, 1996.
- [6] Masaaki Takanezawa, Yo Tomota, and Yoshinori Kobayashi. A new heat resistant cast iron for thin wall exhaust manifold. *The Iron and Steel Institute of Japan*, 38(1):6–8, 1998.
- [7] E. Charkaluk and A. Constantinescu. Energetic approach in thermo-mechanical fatigue for silicon molybdenum cast iron. *Materials at High Temperatures*, 17(3):373–380, 2000.
- [8] Franz Langmayr, Franz Zieher, and Milan Lampic. The thermo-mechanics of cast iron for cylinder heads. *MTZ worldwide*, 65(4):17–20, 2004.
- [9] Sung Hwan Park, Jong Moon Kim, Hak Jin Kim, Se Jin Ko, Hyoun Soo Park, and Jong Dae Lim. Development of a Heat Resistant Cast Iron Alloy for Engine Exhaust Manifolds. *SAE Technical Paper*, (724), 2005.

- [10] Steve Dawson. Compacted Graphite Iron - A material solution for modern diesel engine cylinder blocks and heads. *China Foundry*, 6(3): 241–246, 2009.
- [11] Taner Gocmez and Udo Deuster. An Integral Engineering Solution for Design of Exhaust Manifolds. *SAE Technical Paper*, (2009-01-1229), 2009.
- [12] Delin Li and Clayton Sloss. Ferrous High-Temperature Alloys for Exhaust Component Applications. *SAE International Journal of Materials and Manufacturing*, 3(1):391–404, 2010.
- [13] Pierre-olivier Santacreu. Ferritic Stainless Steel Grade with Improved Durability for High Temperature Exhaust Manifold. *Society of Automotive Engineers*, 2011-01-01, 2011.
- [14] Erhu Chen, Xuelin Wang, and Chengjia Shang. A Low Cost Ferritic Stainless Steel Microalloyed by Higher Nb for Automotive Exhaust System. In *HSLA Steels 2015, Microalloying 2015 and Offshore Engineering Steels 2015 Conference Proceedings*, pages 613–619, 2015.
- [15] Andrew Halfpenny, Robin Anderson, and Xiaobin Lin. Isothermal and Thermo-Mechanical Fatigue of Automotive Components. *SAE Technical Paper*, (2015-01-0548), 2015.
- [16] Willy Leijon (red.). *Karlebo Materiallära*. Liber AB, 15 edition, 2014.
- [17] F Zieher, F Langmayr, A Jelatancev, and K Wieser. Thermal Mechanical Fatigue Simulation of Cast Iron Cylinder Heads. In *SAE world congress*, number 724, 2005.
- [18] S. Trampert, T. Gocmez, and S. Pischinger. Thermomechanical Fatigue Life Prediction of Cylinder Heads in Combustion Engines. *Journal of Engineering for Gas Turbines and Power*, 130(012806):1–10, 2008.
- [19] F. Unkic, Z. Glavas, and K. Terzic. The influence of elevated temperatures on microstructure of cast irons for automotive engine thermo-mechanical loaded parts. *RMZ - Materials and Geoenvironment*, 56(1): 9–23, 2009.
- [20] Mario Metzger, Marc Leidenfrost, Ewald Werner, Hermann Riedel, and Thomas Seifert. Lifetime Prediction of EN-GJV 450 Cast Iron Cylinder Heads under Combined Thermo-Mechanical and High Cycle Fatigue Loading. *SAE International Journal of Engines*, 7(2):2014–01–9047, 2014.

-
- [21] M. Ekström and S. Jonsson. High-temperature mechanical- and fatigue properties of cast alloys intended for use in exhaust manifolds. *Materials Science and Engineering A*, 616:78–87, 2014.
- [22] D. Li and C. Sloss. Comment on "High-temperature mechanical and fatigue properties of cast alloys intended for use in exhaust manifolds" by Ekström et Al. (Mater. Sci. Eng. A 616 (2014) 78). *Materials Science and Engineering A*, 624:90–91, 2015.
- [23] Katherine Avery, Jwo Pan, and Carlos Engler-Pinto. Effect of Temperature Cycle on Thermomechanical Fatigue Life of a High Silicon Molybdenum Ductile Cast Iron. *SAE Technical Paper*, (2015-01-0557), 2015.
- [24] G.N.J. Gilbert. Factors relating to the stress-strain properties of cast iron. *Journal of the British Cast Iron Research Association*, 6:546–588, 1957.
- [25] A. F. Hieber. Fracture in Compacted Graphite Iron. *American Foundry Society Transactions*, 78(31):143–154, 1978.
- [26] K.V. Prabhakar and H.L. Ewalds. An Experimental Technique to Study the Fracture Behaviour of Gray Cast Irons using Scanning Electron Microscopy (SEM). *International cast metals journal*, 5(1):47–59, 1980.
- [27] G. Zambelli. Microcracking in cast iron. In *Fracture and Fatigue: Elasto-Plasticity, Thin Sheet and Micromechanisms Problems*, pages 227–237, 1980.
- [28] L. Eldoky and R.C. Voigt. Fracture of Ferritic Ductile Cast Iron. *AFS Transactions*, 93(58):365–372, 1985.
- [29] R.C. Voigt and C.R. jun. Loper. Microstructural aspects of fracture in cast irons. In *Physical Metallurgy of Cast Iron IV, Proceedings 4th International Symposium*, pages 293–303, Tokyo, Japan, 1989.
- [30] Z. He, S. Ji, and G. Lin. Mechanical behaviour of graphite and its adjacent zone to matrix interface in cast iron. *Acta metallurgica sinica*, 6(3):364–368, 1993.
- [31] M. J. Dong, C. Prioul, and D. François. Damage effect on the fracture toughness of nodular cast iron: Part I. Damage characterization and plastic flow stress modeling. *Metallurgical and Materials Transactions A*, 28(11):2245–2254, 1997.

- [32] Walid M. Mohammed, Eu Gene Ng, and M. A. Elbestawi. On stress propagation and fracture in compacted graphite iron. *International Journal of Advanced Manufacturing Technology*, 56(1-4):233–244, 2011.
- [33] Francesco Iacoviello, Vittorio Di Cocco, Alessandra Rossi, and Mauro Cavallini. Damaging Micromechanisms Characterization in Pearlitic Ductile Cast Irons. *Procedia Materials Science*, 3:295–300, 2014.
- [34] Keivan A. Kasvayee, Ehsan Ghassemali, Kent Salomonsson, S. Sujakhu, S. Castagne, and Anders E.W. Jarfors. Strain localization and crack formation effects on stress-strain response of ductile iron. *Materials Science and Engineering A*, 702:265–271, 2017.
- [35] A.G. Glover and G. Pollard. Deformation and fracture of grey cast-iron structures. *Journal of The Iron and Steel Institute*, 209(2):138–141, 1971.
- [36] R.C. Voigt and S.D. Holmgren. Crack initiation and propagation in gray and compacted graphite iron (CG) cast irons. *American Foundry Society Transactions*, 91:213–225, 1990.
- [37] V. Di Cocco, F. Iacoviello, and M. Cavallini. Damaging micromechanisms characterization of a ferritic ductile cast iron. *Engineering Fracture Mechanics*, 77(11):2016–2023, 2010.
- [38] M. Lundberg, M. Calmunger, and R.L. Peng. In-situ SEM / EBSD Study of Deformation and Fracture Behaviour of Flake Cast Iron. In *13th International conference on fracture*, 2013.
- [39] F Iacoviello, O Di Bartolomeo, V Di Cocco, and V Piacente. Damaging micromechanisms in an as cast ferritic and a ferritized ductile cast iron. *Structural Integrity Procedi*, 3:201–207, 2017.
- [40] E. Hornbogen. Fracture toughness and fatigue crack growth of grey cast irons. *Journal of Materials Science*, 20:3897–3905, 1985.
- [41] M. Petre nec, H. Tesařová, P. Beran, M. Šmíd, and P. Roupcová. Comparison of low cycle fatigue of ductile cast irons with different matrix alloyed with nickel. *Procedia Engineering*, 2(1):2307–2316, 2010.
- [42] V. Norman, P. Skoglund, D. Leidermark, and J. Moverare. The effect of superimposed high-cycle fatigue on thermo-mechanical fatigue in cast iron. *International Journal of Fatigue*, 88:121–131, 2016.

-
- [43] S. Buni, N. Raman, and S. Seshan. The role of graphite morphology and matrix structure on low frequency thermal cycling of cast irons. *Sadhana*, 29(1):117–127, 2004.
- [44] D. Li, R. Perrin, G. Burger, D. Mcfarlan, B. Black, R. Logan, and R. Williams. Solidification Behavior, Microstructure, Mechanical Properties, Hot Oxidation and Thermal Fatigue Resistance of High Silicon SiMo Nodular Cast Irons. Technical Report 724, SAE International, Detroit, 2004.
- [45] D Jedrzejczyk, M Hajduga, and R Lorek. High temperature oxidation as the method of surface treatment of cast iron. *Metal*, 13:1–6, 2008.
- [46] S. Ghodrat, M. Janssen, Roumen H. Petrov, Leo a.I. Kestens, and Jilt Sietsma. Microstructural Evolution of Compacted Graphite Iron under Thermo-Mechanical Fatigue Conditions. *Advanced Materials Research*, 409:757–762, 2011.
- [47] Sepideh Ghodrat, Michael Janssen, Leo A I Kestens, and Jilt Sietsma. Volume expansion of compacted graphite iron induced by pearlite decomposition and the effect of oxidation at elevated temperature. *Oxidation of Metals*, 80(1-2):161–176, 2013.
- [48] Y Yamaguchi, S Kiguchi, H Sumimoto, and K Nakamura. Effects of graphite morphology on decarburized cast iron. *International Journal of cast metals research*, 16(1-3):137–142, 2003.
- [49] V. M. Yuzvak, I. P. Volchok, and V. I. Gontarenko. Micromechanism of fracture of cast irons. *Metal Science and Heat Treatment*, 25(8): 569–573, 1983.
- [50] Madeleine Ekström. *Oxidation and corrosion fatigue aspects of cast exhaust manifolds*. PhD thesis, KTH Royal Institute of Technology, 2015.
- [51] N. Birks, G.H. Meier, and F.S. Pettit. *Introduction to the high temperature oxidation of metals*. Cambridge university press, second edition, 2006.
- [52] T. Adachi and G. H. Meier. Oxidation of iron-silicon alloys. *Oxidation of Metals*, 27(5-6):347–366, 1987.
- [53] Yoon Jun Kim, Ho Jang, and Yong Jun Oh. High-temperature low-cycle fatigue property of heat-resistant ductile-cast irons. *Metallurgical and Materials Transactions A*, 40(9):2087–2097, 2009.

- [54] P. Matteis, G. Scavino, A. Castello, and D. Firrao. High Temperature Fatigue Properties of a Si-Mo Ductile Cast Iron. *Procedia Materials Science*, 3:2154–2159, 2014.
- [55] Cristiana Delprete and Raffaella Sesana. Experimental characterization of a Si-Mo-Cr ductile cast iron. *Materials & Design*, 57:528–537, 2014.
- [56] Xijia Wu, Guangchun Quan, Ryan MacNeil, Zhong Zhang, and Clayton Sloss. Failure Mechanisms and Damage Model of Ductile Cast Iron Under Low-Cycle Fatigue Conditions. *Metallurgical and Materials Transactions A*, 45(11):5085–5097, 2014.
- [57] V. Norman, P. Skoglund, D. Leidermark, and J. Moverare. Damage mechanisms in silicon-molybdenum cast irons subjected to thermo-mechanical fatigue. *International Journal of Fatigue*, 99:258–265, 2017.
- [58] J.W. Fash. Fatigue crack initiation and growth in gray cast iron. Technical report, College of Engineering, University of Illinois at Urbana-Champaign, 1980.
- [59] D.F. Socie and J.W. Fash. Fatigue behavior and crack development in cast iron. Technical report, Department of Mechanical and Industrial Engineering, University of Illinois at Urbana-Champaign, Urbana, 1982.
- [60] H. Nisitani and S. Tanaka. Initiation and propagation of fatigue crack in cast iron. *Transactions of the Japan Society of Mechanical Engineers, Part A*, 51(465):1442–1447, 1985.
- [61] D.J. Weinacht and D.F. Socie. Fatigue damage accumulation in grey cast iron. *International Journal of Fatigue*, 9(2):79–86, 1987.
- [62] S. Tanaka, H. Nisitani, and T. Teranishi. Microcrack propagation behavior of ferritic spheroidal graphite cast iron under rotating bending or reversed torsion. *Transactions of the Japan Society of Mechanical Engineers, Part A*, 60(571):613–618, 1994.
- [63] Xin Tong, Hong Zhou, Lu quan Ren, Zhi hui Zhang, Wei Zhang, and Ren dong Cui. Effects of graphite shape on thermal fatigue resistance of cast iron with biomimetic non-smooth surface. *International Journal of Fatigue*, 31(4):668–677, 2009.

-
- [64] S. Ghodrati, A. C. Riemsdijk, M. Janssen, J. Sietsma, and L. A. I. Kestens. Measurement and characterization of thermo-mechanical fatigue in compacted graphite iron. *International Journal of Fatigue*, 48: 319–329, 2013.
- [65] Mario Metzger, Britta Nieweg, Christoph Schweizer, and Thomas Seifert. Lifetime prediction of cast iron materials under combined thermomechanical fatigue and high cycle fatigue loading using a mechanism-based model. *International Journal of Fatigue*, 53:58–66, 2013.
- [66] M. X. Zhang, J. C. Pang, Y. Qiu, S. X. Li, M. Wang, and Z. F. Zhang. Thermo-mechanical fatigue property and life prediction of vermicular graphite iron. *Materials Science and Engineering A*, 698:63–72, 2017.
- [67] V. Norman, P. Skoglund, and J. Moverare. Damage Evolution in Compacted Graphite Iron during Thermo- Mechanical Fatigue Testing. *International Journal of cast metals research*, 29(1&2):25–32, 2016.
- [68] V. Norman, P. Skoglund, D. Leidermark, and J. Moverare. The transition from micro- to macrocrack growth in compacted graphite iron subjected to thermo-mechanical fatigue. *Engineering Fracture Mechanics*, 186, 2017.
- [69] M.S Starkey and P.E Irving. The influence of microstructure on fatigue crack initiation in spheroidal graphite cast irons. Technical report, GKN group technological centre, Wolverhampton, England, 1979.
- [70] Y. Nadot, J. Mendez, N. Ranganathan, and A. S. Beranger. Fatigue life assessment of nodular cast iron containing casting defects. *Fatigue and Fracture of Engineering Materials and Structures*, 22(4):289–300, 1999.
- [71] Y. Nadot, J. Mendez, and N. Ranganathan. Influence of casting defects on the fatigue limit of nodular cast iron. *International Journal of Fatigue*, 26(3):311–319, 2004.
- [72] N. Costa, N. Machado, and F. S. Silva. A new method for prediction of nodular cast iron fatigue limit. *International Journal of Fatigue*, 32(7):988–995, 2010.
- [73] L. Haenny and G. Zambelli. The stiffness and modulus of elasticity of grey cast iron. *Journal of Materials Science Letters*, 2:239–242, 1983.

- [74] L. Haenny and G. Zambelli. Strain mechanisms in grey cast iron. *Engineering Fracture Mechanics*, 18(2):377–387, 1983.
- [75] T. Hammers, A. Uihlein, K. Lang, and D. Lhe. Development of damage in cast iron during superimposed low frequency thermal-mechanical and higher frequency mechanical loading. In *Plasticity, Failure and Fatigue in Structural Materials-from Macro to Nano: Hael Mughrabi Honorary Symposium*, pages 253–258, New Orleans, 2008.
- [76] Tilmann Beck, Karl-heinz Lang, and Detlef Lhe. Interaction of thermally induced and mechanical fatigue. *Transactions of The Indian Institute of Metals Interaction of thermally induced and mechanical fatigue*, 63(2-3):195–202, 2010.
- [77] Katherine R Avery. *A Predictive Framework for Thermomechanical Fatigue Life of High Silicon Molybdenum Ductile Cast Iron Based on Considerations of Strain Energy Dissipation*. PhD thesis, The University of Michigan, 2016.
- [78] HJ Khn, B Rehmer, and B Skrotzki. Thermomechanical fatigue of heat-resistant austenitic cast iron EN-GJSA-XNiSiCr35-5-2 (Ni-Resist D-5S). *International Journal of Fatigue*, 99:295–302, 2017.
- [79] M W Brown and K J Miller. Initiation and Growth of Cracks in. *Fatigue of Engineering Materials and Structures*, 1:231–246, 1979.
- [80] S G S Raman, D Argence, and A Pineau. High temperature short fatigue crack behaviour in a stainless steel. *Fatigue & Fracture of Engineering Materials & Structures*, 20(7):1015–1031, 1997.
- [81] B. Fournier, M. Sauzay, C. Cas, M. Noblecourt, M. Mottot, A. Bougault, V. Rabeau, and A. Pineau. Creep-fatigue-oxidation interactions in a 9Cr-1Mo martensitic steel. Part I: Effect of tensile holding period on fatigue lifetime. *International Journal of Fatigue*, 30(4):649–662, 2008.
- [82] R.W. Neu and H. Sehitoglu. Thermomechanical fatigue, oxidation and creep: Part I. Damage mechanisms. *Metallurgical Transaction A*, 20(9):1755–1767, 1989.
- [83] Huseyin Sehitoglu. Thermo-mechanical fatigue life prediction methods. In *Advances in Fatigue Lifetime Predictive Techniques , ASTM STP 1122*, pages 47–76, Philadelphia, 1992.

-
- [84] E. Hug, C. Keller, J. Favergeon, and K. Dawi. Application of the Monkman-Grant law to the creep fracture of nodular cast irons with various matrix compositions and structures. *Materials Science and Engineering A*, 518(1-2):65–75, 2009.
- [85] T. Seifert and H. Riedel. Mechanism-based thermomechanical fatigue life prediction of cast iron. Part I: Models. *International Journal of Fatigue*, 32(8):1358–1367, 2010.
- [86] T. Seifert, G. Maier, A. Uihlein, K. H. Lang, and H. Riedel. Mechanism-based thermomechanical fatigue life prediction of cast iron. Part II: Comparison of model predictions with experiments. *International Journal of Fatigue*, 32(8):1368–1377, 2010.
- [87] 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*, 80:381–390, 2015.
- [88] R.W. Neu and H. Sehitoglu. Thermomechanical fatigue, oxidation, and creep: Part II. Life prediction. *Metallurgical Transaction A*, 20(9):1769–1783, 1989.
- [89] Robert Minichmayr, Martin Riedler, Gerhard Winter, Heinz Leitner, and Wilfried Eichlseder. Thermo-mechanical fatigue life assessment of aluminium components using the damage rate model of Sehitoglu. *International Journal of Fatigue*, 30(2):298–304, 2008.
- [90] Xijia Wu. A Model of Nonlinear Fatigue-Creep (Dwell) Interactions. *Journal of Engineering for Gas Turbines and Power*, 131(3):32101, 2009.
- [91] Xijia Wu, Guangchun Quan, Ryan MacNeil, Zhong Zhang, Xiaoyang Liu, and Clayton Sloss. Thermomechanical Fatigue of Ductile Cast Iron and Its Life Prediction. *Metallurgical and Materials Transactions A*, 46(6):2530–2543, 2015.
- [92] G.N.J. Gilbert. Stress/Strain Properties of Cast Iron and Poisson’s Ratio in Tension and Compression. *Journal of the British Cast Iron Research Association*, 9:347–363, 1961.
- [93] M.J. Dong, B. Tie, A.S. Béranger, C. Prioul, and Dominique François. Damage Effect on the Fracture Toughness of Nodular Cast Iron. *Advanced Materials Research*, 4-5(6):181–188, 1997.

- [94] T. Sjögren and I. L. Svensson. Studying elastic deformation behaviour of cast irons by acoustic emission. *International Journal of Cast Metals Research*, 18(4):249–256, 2005.
- [95] Torsten Sjögren, Per Eric Persson, and Peter Vomacka. Analysing the Deformation Behaviour of Compacted Graphite Cast Irons Using Digital Image Correlation Techniques. *Key Engineering Materials*, 457 (457):470–475, 2010.
- [96] Torsten Sjögren, Fredrik Wilberfors, and Monika Alander. Digital Image Correlation Techniques for Analysing the Deformation Behaviour of Compacted Graphite Cast Irons on a Microstructural Level. *Applied Mechanics and Materials*, 70:171–176, 2011.
- [97] Keivan A. Kasvayee, Kent Salomonsson, Ehsan Ghassemali, and Anders E.W. Jarfors. Microstructural strain distribution in ductile iron; comparison between finite element simulation and digital image correlation measurements. *Materials Science and Engineering: A*, 655: 27–35, 2016.
- [98] G.N.J. Gilbert. The components of strain due to deformation of the matrix and due to volume changes in a flake graphite cast iron under uniaxial stress. *Journal of the British Cast Iron Research Association*, 11:512–524, 1963.
- [99] R. Hill. *The mathematical theory of plasticity*. Oxford university press, 1950.
- [100] J. L. Chaboche. A review of some plasticity and viscoplasticity constitutive theories. *International Journal of Plasticity*, 24(10):1642–1693, 2008.
- [101] A. L. Gurson. Continuum Theory of Ductile Rupture by Void Nucleation and Growth: Part I - Yield Criteria and Flow Rules for Porous Ductile Media. *Journal of Engineering Materials and Technology*, 99 (1):2, 1977.
- [102] V Tvergaard. On the localization in ductile materials containing voids. *International Journal of Fracture*, 18(4):237–252, 1982.
- [103] Viggo Tvergaard. Material failure by void growth to coalescence. *Advances in applied mechanics*, 27:83–151, 1989.

- [104] W. Brocks, S. Hao, and D. Steglich. Micromechanical Modelling of the Damage and Toughness Behaviour of Nodular Cast Iron Materials. *Le Journal de Physique IV*, 6:43–52, 1996.
- [105] M. Dahlberg. Micromechanical modelling of nodular cast iron, a composite material. *International Journal of Cast Metals Research*, 9(6): 319–330, 1997.
- [106] D. Steglich and W. Brocks. Micromechanical modelling of damage and fracture of ductile materials. *Fatigue & Fracture of Engineering Materials & Structures*, 21:1175–1188, 1998.
- [107] C. Berdin, M. J. Dong, and C. Prioul. Local approach of damage and fracture toughness for nodular cast iron. *Engineering Fracture Mechanics*, 68(9):1107–1117, 2001.
- [108] Gaël Gigan, Viktor Norman, Johan Ahlström, and Tore Verneresson. Thermomechanical fatigue of grey cast iron brake discs for heavy vehicles. *Proceedings of the Institution of Mechanical Engineers, Part D: Journal of Automobile Engineering*, 2015.
- [109] D. C. Drucker and W. Prager. Soil Mechanics and Plastic Analysis or Limit Design. *Quarterly of Applied Mathematics*, 10(2):157–165, 1952.
- [110] Holm Altenbach, G. B. Stoychev, and K. N. Tushtev. On elastoplastic deformation of grey cast iron. *International Journal of Plasticity*, 17(5):719–736, 2001.
- [111] Mohammed Yusuf Ali, Wei-Jen Lai, Nikhil Kotasthane, Jagadish Sorab, Chari Sever, and Jwo Pan. Stress-Strain Relations for Nodular Cast Irons with Different Graphite Volume Fractions under Tension and Compression. *SAE International Journal of Engines*, 10(2):2017–01–0399, 2017.
- [112] Fabien Szmytka, Luc Rémy, Habibou Maitournam, Alain Köster, and Myriam Bourgeois. New flow rules in elasto-viscoplastic constitutive models for spheroidal graphite cast-iron. *International Journal of Plasticity*, 26(6):905–924, 2010.
- [113] C H Wang and M W Brown. Modelling of the deformation behaviour of flake graphite iron under multi-axial cyclic loading. *The Journal of Strain Analysis for Engineering Design*, 29(4):277–288, 1994.

- [114] Stephen Douglas Downing. *Modeling cyclic deformation and fatigue behavior of cast iron under uniaxial loading*. PhD thesis, University of Illinois at Urbana-Campaign, 1984.
- [115] F. Rabold and M. Kuna. Cell model simulation of void growth in nodular cast iron under cyclic loading. *Computational Materials Science*, 32(3-4):489–497, 2005.
- [116] Nicola Bonora and Andrew Ruggiero. Micromechanical modeling of ductile cast iron incorporating damage. Part I: Ferritic ductile cast iron. *International Journal of Solids and Structures*, 42(5-6):1401–1424, 2005.
- [117] G Nicoletto, L Collini, R Konecná, and E Riva. Analysis of nodular cast iron microstructures for micromechanical model development. *Strain*, 42:89–96, 2006.
- [118] Tito Andriollo and Jesper Hattel. On the isotropic elastic constants of graphite nodules in ductile cast iron: Analytical and numerical micromechanical investigations. *Mechanics of Materials*, 96:138–150, 2016.
- [119] Tito Andriollo, Jesper Thorborg, Niels Tiedje, and Jesper Hattel. A micro-mechanical analysis of thermo-elastic properties and local residual stresses in ductile iron based on a new anisotropic model for the graphite nodules. *Modelling and Simulation in Materials Science and Engineering*, 24(5):055012, 2016.
- [120] Tito Andriollo, Jesper Thorborg, and Jesper Hattel. Modeling the elastic behavior of ductile cast iron including anisotropy in the graphite nodules. *International Journal of Solids and Structures*, 100-101:523–535, 2016.
- [121] Wolfgang Gaudig, Roland Mellert, Ulrich Weber, and Siegfried Schmauder. Self-consistent one-particle 3D unit cell model for simulation of the effect of graphite aspect ratio on Young’s modulus of cast-iron. *Computational Materials Science*, 28:654–662, 2003.
- [122] J. C. Pina, V. G. Kouznetsova, and M. G.D. Geers. Thermo-mechanical analyses of heterogeneous materials with a strongly anisotropic phase: The case of cast iron. *International Journal of Solids and Structures*, 63:153–166, 2015.

- [123] E. S. Russell. Finite element simulation of the microstructure of gray cast iron. Technical report, College of Engineering, University of Illinois at Urbana-Champaign, 1979.
- [124] L Chuzhoy, R E DeVor, S G Kapoor, and D J Bammann. Microstructure-Level Modeling of Ductile Iron Machining. *Journal of Manufacturing Science and Engineering*, 124(2):162, 2002.
- [125] Gianni Nicoletto, Luca Collini, R. Konečná, and P. Bujnová. Strain Heterogeneity and Damage Localization in Nodular Cast Iron Microstructures. *Materials Science Forum*, 482:255–258, 2005.
- [126] J.C. Pina, S. Shafqat, V.G. Kouznetsova, J.P.M. Hoefnagels, and M.G.D. Geers. Microstructural study of the mechanical response of compacted graphite iron: An experimental and numerical approach. *Materials Science and Engineering: A*, 658:439–449, 2016.
- [127] M. Metzger and T. Seifert. Computational assessment of the microstructure-dependent plasticity of lamellar gray cast iron - Part III: A new yield function derived from microstructure-based models. *International Journal of Solids and Structures*, 87:102–109, 2016.
- [128] Felix Fritzen, Samuel Forest, Thomas Böhlke, Djimedo Kondo, and Toufik Kanit. Computational homogenization of elasto-plastic porous metals. *International Journal of Plasticity*, 29(1):102–119, 2012.
- [129] S. Kleiner and K. Track. SiMo1000 Ein aluminiumlegiertes Gusseisen für Hochtemperaturanwendungen. *Giesserei*, 97(10):28–34, 2010.
- [130] Bing Pan, Kemao Qian, Huimin Xie, and Anand Asundi. Two-dimensional digital image correlation for in-plane displacement and strain measurement: a review. *Measurement Science and Technology*, 20(6):062001, 2009.
- [131] Christophe Eberl, Rob Thompson, and Daniel Gianola. Matlab-based DIC code, version 2. Matlab’s File Exchange. <https://se.mathworks.com/matlabcentral/fileexchange/12413-digital-image-correlation-and-tracking>, 2015.
- [132] R. Prescott and M. J. Graham. The oxidation of iron-aluminum alloys. *Oxidation of Metals*, 38(1-2):73–87, 1992.

Papers

The papers associated with this thesis have been removed for copyright reasons. For more details about these see:

<http://urn.kb.se/resolve?urn=urn:nbn:se:liu:diva-145176>