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世界学术研究前沿丛书

Material Science

材料科学

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世界图书出版公司

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中国出版集团公司
世界图书出版公司
广州 · 上海 · 西安 · 北京

图书在版编目（CIP）数据

材料科学：英文 / “世界学术研究前沿丛书”编委会编. —广州：世界图书出版广东有限公司，2017.8
ISBN 978-7-5192-2458-5

Ⅰ. ①材… Ⅱ. ①世… Ⅲ. ①材料科学—英文 Ⅳ. ①TB3

中国版本图书馆 CIP 数据核字(2017)第 040980 号

书　　名：材料科学
　　　　　Cailiao Kexue
编　　者：“世界学术研究前沿丛书”编委会
责任编辑：康琬娟
出版发行：世界图书出版广东有限公司
地　　址：广州市海珠区新港西路大江冲25号
邮　　编：510300
电　　话：（020）84460408
网　　址：http：//www.gdst.com.cn/
邮　　箱：wpc_gdst@163.com
经　　销：新华书店
印　　刷：广州市德佳彩色印刷有限公司
开　　本：787 mm×1092 mm　1/16
印　　张：42
插　　页：4
字　　数：1020千
版　　次：2017年8月第1版　　2017年8月第1次印刷
国际书号：ISBN 978-7-5192-2458-5
定　　价：598.00元

Preface

The interdisciplinary field of materials science, also commonly known as materials science and engineering, involves the discovery and design of new materials, with an emphasis on solids. The intellectual origins of materials science stem from the Enlightenment, when researchers began to use analytical thinking from chemistry, physics, and engineering to understand ancient, phenomenological observations in metallurgy and mineralogy. Materials science still incorporates elements of physics, chemistry, and engineering. In recent years, materials science has become more widely recognized as a specific and distinct field of science and engineering. Many of the most pressing scientific problems humans currently face are due to the limitations of the materials that are available and, as a result, breakthroughs in materials science are likely to have a significant impact on the future of technology.[1]

In the present book, numerous literatures about materials science published on international authoritative journals were selected to introduce the worldwide newest progress, which contains reviews or original researches on metallic materials, inorganic non-metallic materials, polymer materials and composite materials and so on. We hope this book can demonstrate advances in materials science as well as give references to the researchers, students and other related people.

March 9，2017

[1]From Wikipedia: https://en.wikipedia.org/wiki/Materials_science

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Contents

Chapter 1

Retardation of Plastic Instability via Damage-Enabled Microstrain Delocalization

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Abstract: Multi-phase microstructures with high mechanical contrast phases are prone to microscopic damage mechanisms. For ferrite-martensite dual-phase steel, for example, damage mechanisms such as martensite cracking or martensite-ferrite decohesion are activated with deformation, and discussed often in literature in relation to their detrimental role in triggering early failure in specific dual-phase steel grades. However, both the micromechanical processes involved and their direct influence on the macroscopic behavior are quite complex, and a deeper understanding thereof requires systematic analyses. To this end, an experimental-theoretical approach is employed here, focusing on three model dual-phase steel microstructures each deformed in three different strain paths. The micromechanical role of the observed damage mechanisms is investigated in detail by in-situ scan-

ning electron microscopy tests, quantitative damage analyses, and finite element simulations. The comparative analysis reveals the unforeseen conclusion that damage nucleation may have a beneficial mechanical effect in ideally designed dual-phase steel microstructures (with effective crack-arrest mechanisms) through microscopic strain delocalization.

1. Introduction

In the last decades, novel advanced high-strength steels (AHSS) with more and more complex microstructures have been introduced (e.g., twinning-assisted plasticity steels[1][2], quench and partition steels[3][4], and carbide-free bainite steels[5][6]) to achieve superior mechanical performance compared to existing grades. Yet, the connection between the microstructure and the overall mechanical behavior is still not fully set even for the more established AHSS grades, such as dual-phase (DP) steels that have been present for decades[7]. The martensitic-ferritic microstructures of DP steels provide excellent combinations of high strength and good ductility[7]–[9] at low cost (*i.e.*, low alloying content) and relatively simple thermo-mechanical processing (*i.e.*, intercritical annealing). Thus, DP steels are nowadays being used or considered for different automotive components, e.g., for crash box structures.

The development of DP steels was triggered in the early 1970s and intensive research has been done since then. A huge experimental literature exists, which has shown the influence of martensite volume fraction[10][11], grain size of the constituents, and grain refinement[9][12][13], as well as carbon content[14], on the ultimate strength and ductility of DP steels. Models that account for such effects have been proposed and widely used, e.g.,[15]–[17]. The influence of the morphology of the constituents has also been extensively studied, both from experimental and computational points of view, e.g.,[18]–[20].

A wider application of DP steels is hampered by the limited understanding regarding their failure mechanisms. For example, it is beneficial for weight reduction purposes to employ higher strength DP grades in automotive body-in-white structures, as it would allow sheet thickness to be reduced. However, in such higher strength grades (with higher martensite content), activity of microstructural damage mechanisms may often lead to unpredicted failures during forming operations or upon crash[21][22]. The limited understanding of the macroscopic fracture processes in DP steel arises from the presence of multiple microstructural damage mechanisms that exhibit complex interactions[23]–[32]. As a consequence, the applicability of state-of-the-art damage models that aim at modeling multiple, interacting, damage nucleation mechanisms, e.g.,[33][34], is limited by the possibilities for experimental characterization, see e.g.,[29][31].

The challenge is thus clear: developing optimized martensite-ferrite microstructures that enable higher strengths in DP steels, while preserving good toughness. To this end, a vast variety of microstructure variations can be introduced in DP steels by small changes in the composition and/or thermomechanical processing[18][35]–[42]. To guide this microstructure design process, micromechanics-based foundations and design guidelines are needed that would ensure damage-prone microstructures. This research aims to provide an improved understanding in this direction.

There are many investigations in the literature on damage and failure mechanisms in DP steels[23]–[27][29]. These reports reveal three general observations:

- Aside from the rarely seen damage incidents at ferrite grain boundaries (DFGB), ferrite grain interiors (DFGI), or around inclusions (DINC), two main damage mechanisms are dominant in DP steel microstructures: martensite cracking (DMC) and martensite-ferrite interface damage (D_{MFI}).[1]

- The relative activity of these two mechanisms, their activation regimes, and their role on the overall mechanical response are strongly microstructure and strain path dependent.

- While its mechanical effect is critical, the overall damage fraction is difficult to detect as it is in the order of few percent even at high deformation levels.

Given these points, it is clear that generic microstructure design guidelines

[1]Note that the latter is often referred to in the literature as martensite-ferrite decohesion mechanism.

cannot be provided through qualitative analysis of a single microstructure deformed in a single strain path, as is done in most previous works. Therefore, in this research, we aim to improve on this by employing an experimental-numerical approach that has various novelties: (i) Experiments focus on quantitative characterization of ductile damage evolution up to failure, at different strain paths and strain levels; (ii) For these experiments, a recently designed miniaturized Marciniak setup[43] and a novel image post-processing methodology are employed for statistically sound quantification of damage evolution; (iii) Different model DP microstructures (with variation in only a single microstructural variable at a time) are investigated using these techniques; and (iv) For a deeper understanding of the most relevant damage nucleation mechanisms, follow-up in-situ scanning electron microscope (SEM) deformation experiments and finite element simulations are also carried out.

In what follows, first the employed methodology is introduced in detail. The results are presented, starting with the identification of active damage mechanisms and quantification of their activity, followed by focusing on the factors determining the relative activity of the damage mechanisms through a discussion of the numerical results and in-situ damage nucleation images. The report is finalized with the conclusions.

2. Methodology

2.1. Materials

To investigate systematically the influence of ferrite grain size and martensite volume fraction, different DP model microstructures are produced where a single microstructural parameter is changed at a time. These microstructures, referred in the text as fine-grained (μ_{FG}), coarse-grained (μ_{CG}), and high martensite (μ_{HM}) microstructures, are designed by thermal processing of non-commercial DP600 and DP800 steel grades of 1mm thickness from Tata steel IJmuiden (**Figure 1**). These base steels are chosen specifically, as they have almost equal (typical) concentrations of Mn, Si, and Cr, while differing only in C (0.092wt% vs. 0.147wt%, respectively). The μ_{CG} is produced by reaustenization of DP600 alloy at 960°C for 10min, followed by air cooling to room temperature, then intercritical

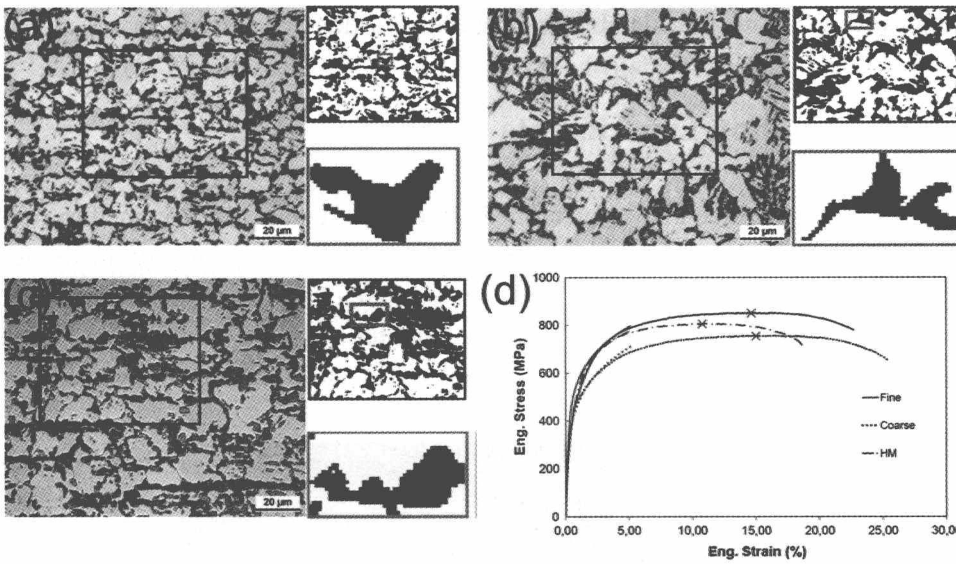

Figure 1. Optical microscopy images and SEM images (not shown) of a the μ_{FG}, b the μCG, and c the μHM microstructures, recorded at the center cross section (sheet thickness in vertical direction) of the specimen; no dependence of the generated microstructures on the prior sheet rolling direction was observed. Each optical image a-c was first converted into black (martensite) and white (ferrite) images (top right subfigure of a-c), and then converted to a representative volume element (RVE) for FEM analysis (each SEM pixel is converted to a finite element to a total of ~1000 × 800 elements), of which a zoom with martensite colored red (bottom right subfigure of a-c) shows the fine mesh used. Shown in d are the global stress-strain curves under uniaxial tension of the μ_{FG}, μ_{CG}, and μ_{HM} microstructures, with the point of plastic instability marked with a red cross. In dark red are shown the simulated stress-strain curves for each microstructure, which were fitted to the experimental curves by adapting the plastic model parameters (given in **Table 1**) (Color figure online).

annealing at 775°C for 30min, and finally quenching to room temperature. To produce the μ_{FG}, the re-austenization duration of the same alloy is decreased to 1min,[2], keeping the other conditions of the treatment identical. Decrease in austenization duration limits the growth of austenite grains but identical intercritical annealing treatment ensures largely unaffected martensite volume fraction (~33%) and morphology. To produce the μ_{HM} microstructure, DP800 steel is heat treated in the same manner as the μ_{CG}. For the same intercritical annealing temperature, the DP800 steel with higher carbon content produces a higher martensite volume percentage (~41%) compared to the DP600 steel with lower carbon content, while the

[2]Note that accompanying dilatometry experiments reveals that the austenite transformation is completed within this duration.

martensite carbon contents in both are, on average, identical (**Figure 1**).

2.2. Deformation Experiments

Each of the three above-mentioned DP microstructures is deformed to fracture in three different strain paths: uniaxial tension (UAT), plane strain tension (PST), and biaxial tension (BAT). To carry out these deformation experiments, the miniaturized Marciniak setup with a punch diameter of 40mm[43], shown in **Figure 2(a)**, is employed. A finite element analysis of this Marciniak test showed

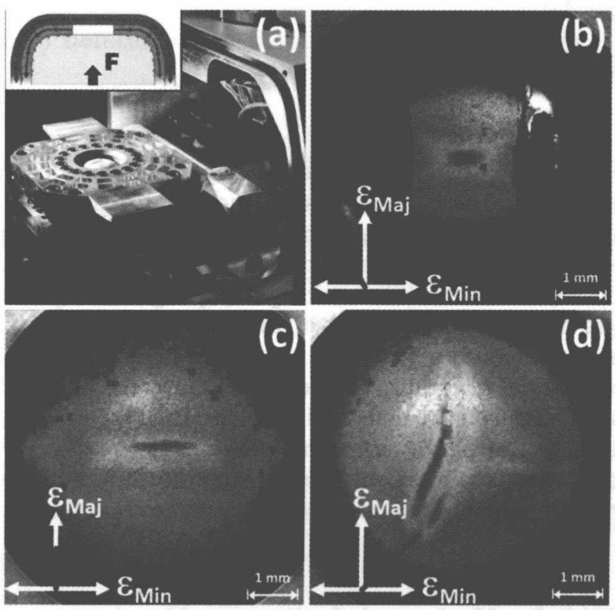

Figure 2. A Photograph of the miniaturized Marciniak setup[43], mounted in the door of the SEM, and used to perform all deformation experiments. The inset shows a schematic representation of the working principle of a Marciniak test on a specimen and so-called 'washer' (with central hole), where the red arrows show the in-plane displacement and the blue arrows the friction direction[43]. b-d The Real-time recorded optical images of the DP steel with fine-grained microstructure loaded under b UAT, c PST, and d BAT up to the first point of failure, and overlaid with the von Mises strain field obtained through digital image correlation. For each of the 3 strain paths, the strain fields have been used to calculate the evolution of the major strain, ε_{Maj}, as a function of the minor strain, ε_{Min}, which is shown in the inserts by the red curves. Note that at these critical strains the correlation of some subsets was lost due to large out-of-plane rotations at the specimen edge and detachment of the spray paint pattern at the specimen center, however, this did not interfere with the analysis (Color figure online).