Sven Curtze

Characterization of the Dynamic Behavior and Microstructure Evolution of High Strength Sheet Steels

Tampere 2009
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Thesis for the degree of Doctor of Technology to be presented with due permission for public examination and criticism in Konetalo Building, Auditorium K1702, at Tampere University of Technology, on the 9th of October 2009, at 12 noon.
ABSTRACT

The use of modern high strength steels enables automotive producers to notably reduce the vehicle body mass and this way meet today’s fuel efficiency and green house gas emission requirements without compromising structural integrity and passenger safety. With new steel generations under development, further mass reductions as well as reductions in production costs are in sight. The superior mechanical properties of these steels, which are classified under the name advanced high strength steels (AHSS), are a result from advanced microstructure design, which in many cases utilizes complex deformation and/or phase transformation processes in the material. Some of the most prominent representatives of the group of AHSS are dual phase (DP) and transformation induced plasticity (TRIP) steels, as well as the more recently developed high manganese twinning induced plasticity (TWIP) steels.

The stress-strain response of these materials at high strain rates is particularly interesting and important, on the one hand from a practical point of view, e.g., in a car crash event or in high rate forming processes, and on the other hand from a purely scientific point of view. The same argument basically also applies to the effects of temperature on the mechanical properties of theses steels. While the deformation mechanisms of materials deforming solely by dislocation glide are widely understood, as well as their dependence on temperature and strain rate, the mechanical behavior of materials additionally undergoing phase transformations and/or twinning is much more complex.

In order to better understand the role of phase transformations and/or twinning during plastic deformation of AHSS under varying deformation conditions, a DP 600 and a TRIP 700 steel were studied by mechanical tests in a strain rate range of roughly $10^{-3} \text{ s}^{-1}$ $\leq \dot{\varepsilon} \leq 10^{3} \text{ s}^{-1}$ while simultaneously varying the temperature between -100 °C and + 235 °C. X-ray diffraction measurements were performed on the as-delivered and deformed specimens. The mechanical behavior of fully austenitic high manganese TWIP steels is strongly dependent on the stacking fault energy $\gamma_{SFE}$, which determines the ease of cross slip and the twinning propensity. Five experimental high manganese TWIP steels were produced based on thermodynamic stacking fault energy calculations and tested in a similar manner as the DP and TRIP steels. The microstructure and texture evolution of the TWIP steels during plastic deformation was studied on some of the specimens by means of electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM).

The low and intermediate strain rate tests ($10^{-3}…1\text{s}^{-1}$) were carried out with a conventional commercial Instron 8800 servohydraulic materials testing machine, while the Hopkinson split bar technique was utilized for the high strain rate tests (~500…1250 s$^{-1}$). Various techniques were used to vary the test temperature. To guarantee the best possible reliability of the test results, extensive studies on the tensile Hopkison pressure bar technique, in particular the specimen geometry, were included in the present study.
The strain hardening behavior of the TRIP steel showed strong dependence on both temperature and strain rate, to which the DP steel seemed to be less sensitive. The mechanical behavior of the TWIP steels varied strongly with temperature and applied strain rate. In many cases the results revealed a good correlation with the calculated values of the stacking fault energy. Reduced elongations at low temperatures exhibited by TWIP steels with higher SFE’s could be attributed mostly to the underdeveloped microstructure and crystal orientations, and not as usually proposed in the literature, to extensive twinning at low strain levels. The measuring techniques, detailed SFE calculations, and the results of the mechanical tests and microstructure characterization are presented and discussed in details in this thesis.
PREFACE

The work presented in this thesis was carried out at the Department of Materials Science, Tampere University of Technology during the years 2004-2009, initiated in a project financed by the Academy of Finland. The work was carried out under the guidance of Professor Veli-Tapani Kuokkala to whom I am indebted and wish to express my sincere gratitude for his guidance, inspiration, and enthusiasm.

I would like to thank the members, past and present, of the Department of Materials Science and especially of the Hopkinson group for creating such a pleasant working atmosphere. Special thanks should go to Dr. Mikko Hokka for the prosperous cooperation and discussions, professional in nature and beyond. Thanks are also due to Dr. Taina Vuoristo for her support, especially in the early stages of my work. M.Sc. Jari Rämö, M.Sc. Jari Kokkonen, and M.Sc. Matti Isakov are gratefully acknowledged for their collaboration and fruitful discussions. Major parts of the microscopic studies were enabled on account of Lic. Tech. Kati Rissa and M.Sc. Tuomo Saarinen, whom I would like to thank. I also wish to express special gratitude to Dr. Pasi Peura, Rautaruukki Oyj., for the industry cooperation as well as the valuable contributions in terms of comments and discussions.

Mr. Ari Varttila is gratefully acknowledged, especially for the machining of the specimens and Mr. Terho Kaasalainen for the technical support and creativity with the construction of the testing equipment.

I am much obliged for the financial support by the Academy of Finland, the Graduate School on New Materials and Processes, Tekniikan Edistämissäätiö (TES), Rautaruukki Oyj, and the Graduate School CE Tampere.

My parents are most warmly thanked for their endless support and encouragement throughout my life, as well as my brother Sami. Finally, a thousand thanks to Katja for being by my side during this path and believing in me.

Tampere, September 2009

Sven Curtze
LIST OF INCLUDED PUBLICATIONS


AUTHOR’S CONTRIBUTION

The author performed the calculations and numerical simulations presented in the thesis and was involved in designing and modifying of the tensile Hopkinson split bar test system and the high and low temperature test systems. He planned the experiments, conducted or supervised the experimental tests and electron microscopy studies, partly in teamwork, and evaluated and interpreted the results.

The results and all publications were commented by and revised by the supervisor, Professor Kuokkala. The original manuscript was written by the author and commented by the supervisor. All the attached papers were also commented by the co-authors.
LIST OF SYMBOLS AND ABBREVIATIONS

<table>
<thead>
<tr>
<th>Symbol</th>
<th>Description</th>
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</thead>
<tbody>
<tr>
<td>A</td>
<td>Interfacial area</td>
</tr>
<tr>
<td>A_b</td>
<td>Cross sectional area of the bar in a HSB device</td>
</tr>
<tr>
<td>A_inc</td>
<td>Cross sectional area of the incident bar</td>
</tr>
<tr>
<td>A_s</td>
<td>Cross sectional area of the specimen</td>
</tr>
<tr>
<td>A_trans</td>
<td>Cross sectional area of the transmitted bar</td>
</tr>
<tr>
<td>a</td>
<td>Lattice parameter</td>
</tr>
<tr>
<td>b, bS</td>
<td>Burgers vector</td>
</tr>
<tr>
<td>C_0</td>
<td>Sound velocity</td>
</tr>
<tr>
<td>C_inc</td>
<td>Sound velocity in the incident bar</td>
</tr>
<tr>
<td>C_trans</td>
<td>Sound velocity in the transmitted bar</td>
</tr>
<tr>
<td>c</td>
<td>Specific heat capacity</td>
</tr>
<tr>
<td>c_EL</td>
<td>Longitudinal sound velocity</td>
</tr>
<tr>
<td>c_PL</td>
<td>Plastic wave velocity</td>
</tr>
<tr>
<td>D</td>
<td>Diameter of the specimen</td>
</tr>
<tr>
<td>d</td>
<td>Coefficient (Gibbs energy power series expression for pure elements)</td>
</tr>
<tr>
<td>e</td>
<td>Engineering strain</td>
</tr>
<tr>
<td>F</td>
<td>Force</td>
</tr>
<tr>
<td>f^p</td>
<td>Polynomial function (in the magnetic contribution to the Gibbs energy)</td>
</tr>
<tr>
<td>E</td>
<td>Young’s modulus</td>
</tr>
<tr>
<td>E_inc</td>
<td>Young’s modulus of the incident bar</td>
</tr>
<tr>
<td>E_trans</td>
<td>Young’s modulus of the transmitted bar</td>
</tr>
<tr>
<td>G</td>
<td>Shear modulus</td>
</tr>
<tr>
<td>G^g</td>
<td>Gibbs energy of pure elements</td>
</tr>
<tr>
<td>G_mg</td>
<td>Magnetic contribution to the Gibbs energy</td>
</tr>
<tr>
<td>g</td>
<td>Reflecting plane in TEM images</td>
</tr>
<tr>
<td>if</td>
<td>Interface</td>
</tr>
<tr>
<td>L_g</td>
<td>Linear function of T (in the expression of the excess free energy)</td>
</tr>
<tr>
<td>L_g</td>
<td>Constant (in the expression of the excess free energy)</td>
</tr>
<tr>
<td>L_s</td>
<td>Length of the specimen</td>
</tr>
<tr>
<td>L_str</td>
<td>Length of the striker bar</td>
</tr>
<tr>
<td>M</td>
<td>Orientation (Schmid) factor</td>
</tr>
<tr>
<td>M_D</td>
<td>Temperature above which austenite is stable</td>
</tr>
<tr>
<td>M_f</td>
<td>Temperature where all austenite has transformed to martensite</td>
</tr>
<tr>
<td>M_S</td>
<td>Martensite reaction start temperature</td>
</tr>
</tbody>
</table>

v


\( M_\text{s}' \)  
Temperature above which the parent phase yields before transformation

\( m \)  
Strain rate sensitivity factor

\( M \)  
Mass

\( N \)  
Avogadro’s constant

\( n \)  
Strain hardening exponent

\( p \)  
Structure dependent constant

\( R \)  
Gas constant

\( s \)  
Fillet radius

\( T \)  
Engineering stress

\( T \)  
Temperature

\( T_m \)  
Structure

\( T \)  
Specimen thickness

\( T_{\text{Néel}} \)  
Melting temperature

\( t \)  
Néel temperature

\( u \)  
Time

\( U' \)  
Mechanical energy required by the martensitic reaction

\( u_{\text{inci}} \)  
Displacement along the bar axis

\( u_{\text{trans}} \)  
Displacement of the end of the incident bar

\( V \)  
Displacement of the end of the transmitted bar

\( V_{\gamma} \)  
Volume fraction of the austenite phase

\( V_{\text{a} \rightarrow \text{a}'} \)  
Volume fraction of austenite phase transforming into martensite

\( v_{\text{inci}} \)  
Velocity of the end of the incident bar

\( v_s \)  
Shear wave velocity

\( v_{\text{trans}} \)  
Velocity of the end of the transmitted bar

\( W \)  
Specimen width

\( x \)  
Coordinate along the longitudinal bar axis

\( Z \)  
Acoustic impedance

\( \alpha', \alpha_{\text{bec}}^{\text{Ms}} \)  
Martensite phase

\( \beta \)  
Fraction of mechanical energy converted to heat

\( \beta^0 \)  
Magnetic moment

\( \gamma, \gamma_{\text{fcc}} \)  
Austenite phase

\( \gamma, \gamma_{\text{SFE}} \)  
Stacking fault energy

\( \gamma_T \)  
Twinned structure

\( \gamma^\ast \)  
Interfacial energy per unit area of interface

\( \gamma^\ast \)  
Ideal stacking fault energy

\( \gamma^\ast \)  
Effective stacking fault energy

\( \gamma \)  
Shear strain rate

\( \Delta G \)  
Thermal energy

\( \Delta G_i \)  
Change in Gibbs energy

\( \Delta G_i \)  
Energy required to overcome obstacle i
ΔGγ→ε  Change in Gibbs energy upon γ→ε phase transformation
ΔGγ→εφ  Chemical contribution to the change in Gibbs energy
ΔGγ→εmg  Magnetic contribution to the change in Gibbs energy

ΔGγ→α′  Energy required initiating the γ→α′ phase transformation
ΔGγ→εM  Energy required initiating the γ→ε phase transformation
ΔGγ→α′T1  Thermal driving energy for the martensitic reaction at temperature T1
ΔQ  Energy converted to heat
ε  Logarithmic (true) strain, ε = ln(e + 1)
εM  Epsilon martensite
εi  Incident stress pulse
εr  Reflected stress pulse
εt  Transmitted stress pulse
ε  Longitudinal strain rate
Θ  Strain hardening rate
μ  Shear modulus
μB  Bohr magneton
ν  Dislocation velocity
νS  Striker velocity
ν  Poisson’s ratio
vs  Shear wave velocity
ρ  Density of the bar material
ρm  Molar surface density
Average dislocation density
σ  True stress, σtrue = s(e + 1)
γ/ε  Interface energy
σ0  Thermal stress component
σG  Athermal stress component
τ  (Resolved) shear stress
τc  Critical (Resolved) shear stress
Twinning stress
τi  Shear stress needed to overcome obstacle i
τφ  Scaled Néel temperature: T/TNéel
φ, Φ  Elements
Phase
Ω, Ωφ→ε  Excess free energy
<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
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<tbody>
<tr>
<td>ABAQUS</td>
<td>Commercial finite element analysis software</td>
</tr>
<tr>
<td>AISI</td>
<td>American Iron and Steel Institute</td>
</tr>
<tr>
<td>ASTM</td>
<td>American Society for Testing and Materials</td>
</tr>
<tr>
<td>BCC</td>
<td>Body centered cubic structure</td>
</tr>
<tr>
<td>BF</td>
<td>Bright field (image)</td>
</tr>
<tr>
<td>CR</td>
<td>Cold rolled</td>
</tr>
<tr>
<td>DIC</td>
<td>Digital Image Correlation</td>
</tr>
<tr>
<td>DP</td>
<td>Dual phase steel</td>
</tr>
<tr>
<td>EBSD</td>
<td>Electron back-scatter diffraction</td>
</tr>
<tr>
<td>FCC</td>
<td>Face centered cubic structure</td>
</tr>
<tr>
<td>FEM</td>
<td>Finite element method</td>
</tr>
<tr>
<td>FFT</td>
<td>Fast Fourier Transform</td>
</tr>
<tr>
<td>HCP</td>
<td>Hexagonal close-packed</td>
</tr>
<tr>
<td>HKL</td>
<td>Commercial EBSD acquisition and evaluation software</td>
</tr>
<tr>
<td>HR</td>
<td>Hot rolled</td>
</tr>
<tr>
<td>HSB</td>
<td>Hopkinson Split Bar</td>
</tr>
<tr>
<td>HT</td>
<td>High temperature</td>
</tr>
<tr>
<td>DMS/TUT</td>
<td>Department of Materials Science / Tampere University of Technology</td>
</tr>
<tr>
<td>IR</td>
<td>Infrared</td>
</tr>
<tr>
<td>LT</td>
<td>Low temperature</td>
</tr>
<tr>
<td>LVDT</td>
<td>Linear variable differential transformer</td>
</tr>
<tr>
<td>Matlab</td>
<td>Programming language for technical computing</td>
</tr>
<tr>
<td>PID</td>
<td>Proportional-integrative-differential (controller)</td>
</tr>
<tr>
<td>RD</td>
<td>Rolling direction</td>
</tr>
<tr>
<td>RT</td>
<td>Room temperature</td>
</tr>
<tr>
<td>SADP</td>
<td>Selected area diffraction pattern</td>
</tr>
<tr>
<td>SFE</td>
<td>Stacking fault energy</td>
</tr>
<tr>
<td>SFP</td>
<td>Stacking fault phase</td>
</tr>
<tr>
<td>TD</td>
<td>Transverse direction</td>
</tr>
<tr>
<td>TE</td>
<td>Total elongation</td>
</tr>
<tr>
<td>TEM</td>
<td>Transmission electron microscopy</td>
</tr>
<tr>
<td>THSB</td>
<td>Tensile Hopkinson Split Bar</td>
</tr>
<tr>
<td>TRIP</td>
<td>Transformation induced plasticity</td>
</tr>
<tr>
<td>TT</td>
<td>Through thickness direction</td>
</tr>
<tr>
<td>TWIP</td>
<td>Twinning induced plasticity</td>
</tr>
<tr>
<td>UE</td>
<td>Uniform elongation</td>
</tr>
<tr>
<td>UTS</td>
<td>Ultimate tensile strength</td>
</tr>
<tr>
<td>Vol</td>
<td>Volume</td>
</tr>
<tr>
<td>XRD</td>
<td>X-ray diffraction, X-ray diffractometer</td>
</tr>
<tr>
<td>YS</td>
<td>Yield strength</td>
</tr>
</tbody>
</table>
CONTENTS

ABSTRACT ....................................................................................................................................................... i
PREFACE ........................................................................................................................................................... iii
LIST OF INCLUDED PUBLICATIONS ..................................................................................................................... iv
AUTHOR’S CONTRIBUTION .............................................................................................................................. iv
LIST OF SYMBOLS AND ABBREVIATIONS ......................................................................................................... v
CONTENTS ..................................................................................................................................................... ix
1 INTRODUCTION ........................................................................................................................................... 1
2 MECHANISMS OF PLASTIC DEFORMATION IN METALS ........................................................................... 4
  2.1 Stacking faults, \(\varepsilon/\alpha’\)-martensite and deformation twins in fcc metals ......................................................... 4
  2.1.1 Stacking faults ........................................................................................................................................ 4
  2.1.2 \(\varepsilon\)-martensite ..................................................................................................................................... 5
  2.1.3 \(\alpha’\)-martensite .................................................................................................................................... 6
  2.1.4 Deformation twins .................................................................................................................................. 6
  2.2 Advanced high strength steels (AHSS): DP, TRIP, and TWIP ........................................................................... 8
  2.2.1 Dual phase (DP) steel ....................................................................................................................... 8
  2.2.2 Transformation induced plasticity (TRIP) steel ............................................................................... 9
  2.2.3 High manganese twinning induced plasticity (TWIP) steel ............................................................ 11
  2.3 Dislocation dynamics .................................................................................................................................. 12
  2.3.1 Thermally activated dislocation motion ........................................................................................... 12
  2.3.2 Strain rate dependent mechanical behavior due to thermally activated dislocation motion .......... 13
  2.3.3 Dislocation drag ................................................................................................................................... 16
  2.3.4 Effects of strain rate and temperature on phase transformations and twinning ................................ 17
  2.3.5 The strain rate sensitivity factor ....................................................................................................... 19
  2.4 Stacking fault energy .................................................................................................................................. 20
  2.4.1 Thermodynamic Stacking Fault Energy Calculations .................................................................... 21
3 MECHANICAL MATERIALS TESTING ........................................................................................................ 27
  3.1 Inertia and wave propagation effects ........................................................................................................ 28
  3.2 Adiabatic heating ......................................................................................................................................... 30
  3.3 Hopkinson split bar technique ................................................................................................................ 31
  3.3.1 Elastic waves in cylindrical bars ......................................................................................................... 31
  3.3.2 Compression and tensile Hopkinson split bar ................................................................................... 32
  3.3.3 Elastic wave dispersion ....................................................................................................................... 35
  3.3.4 Effects of specimen geometry .............................................................................................................. 36
4 EXPERIMENTAL PROCEDURES .................................................................................................................. 37
  4.1 Materials ..................................................................................................................................................... 37
  4.1.1 TRIP and DP steels ............................................................................................................................ 37
  4.1.2 TWIP steels ......................................................................................................................................... 38
  4.2.1 Compression test specimens ............................................................................................................. 38
  4.2.2 Tensile test specimens ....................................................................................................................... 38
  4.3 Hopkinson split bar equipment ................................................................................................................. 44
  4.3.1 Compression Hopkinson Split bar ...................................................................................................... 44
  4.3.2 Tensile Hopkinson Split bar ................................................................................................................ 44
4.3.3 Specimen heating/cooling techniques in mechanical testing................. 46
4.4 Microstructure and texture characterization by EBSD and TEM............... 48
5 RESULTS AND DISCUSSION........................................................................ 49
  5.1 TRIP and DP steels ............................................................................... 49
    5.1.1 Mechanical behavior................................................................. 49
  5.2 TWIP steels......................................................................................... 59
    5.2.1 Thermodynamic Stacking Fault Energy Calculations ................. 59
    5.2.2 Mechanical behavior of the high manganese TWIP steels............ 64
    5.2.3 Microstructure and texture evolution....................................... 71
    5.2.4 Defect structures ................................................................. 75
6 SUMMARY AND CONCLUSIONS .................................................................. 77
APPENDIX.................................................................................................. 80
References................................................................................................ 82
1 INTRODUCTION

The development of new steel grades has resulted mostly from the demands set by the industry, in the case of sheet steels mainly by the automotive industry. In the early 1970s, growing interest in DP steels arose as a response to the worldwide oil crises, which forced car manufacturers to produce lighter and more fuel-efficient vehicles. The good workability and high mechanical strength of DP steels allowed to gauge down the thickness of structural car members at simultaneously reduced production costs without compromising the impact safety of the final parts. As a direct descendant of the DP steel, the TRIP (Transformation Induced Plasticity) steel was developed later. The TRIP steel takes advantage of the martensitic phase transformation occurring in grains of retained austenite during mechanical deformation of the material, which results in even larger elongation values at more steady strain hardening rates when compared to the DP steel. From the late 1980s on, the development of new steel grades was primarily driven by the energy conservation policy set by the United Nations, aiming at the mitigation of global warming. The requirements that steels were facing became at the same time more diversified due to the competition from alternative structural materials such as aluminum and various composite materials. One of the more recent automotive steel innovations is the high manganese TWIP (Twinning Induced Plasticity) steel, which complies with the demand for even lower yield strength combined with high maximum strength and total elongation. The first commercial TWIP steel grades have entered the market only a few years ago.

The aforementioned steels belong to the group of advanced high strength steels (AHSS), further prominent representatives of which are the complex-phase (CP) steels as well as martensitic steels. The term ‘high strength’ refers to tensile strength levels of 340 MPa or more, which traditionally is considered a tough value for press forming operations [1]. AHSS’s exhibit also often properties that make them superior to conventional steel grades, like enhanced ductility and excellent strain hardening capability in many cases brought about by complex deformation and/or phase transformation processes in the material.

For the high-speed material forming processes as well as for the crash safety performance of vehicles, high strain rate material behavior is an important issue. The temperature and strain rate dependence of the mechanical properties of most crystalline materials can be explained by the theory of thermally activated dislocation motion; the material’s strength usually increases with increasing strain rate and with decreasing temperature, strain rate and temperature thereby being coupled by an Arrhenius type equation. The strain rate range where thermally activated dislocation motion dominates the mechanical behavior, however, breaks down when entering the strain rate range where dislocation drag mechanisms start to control the deformation. This happens, depending on the material, usually at strain rates around 1000 s⁻¹ and is observed experimentally as a dramatic increase in the strength levels. For materials exhibiting phase transformations and/or twinning during mechanical deformation, the temperature and strain rate dependences...
become much more complex, and no generally accepted comprehensive model that includes also these effects has yet been presented in the literature.

The most common technique to study the high strain rate response of materials is the Hopkinson Split Bar (HSB) technique, also called the Kolsky bar technique, which finds its origins in the late 19th and early 20th centuries, when John Hopkinson [2] and his son Bertram [3] laid the foundation for this technique with their pioneering experimental work on the wave propagation in slender metal rods. Later in the 1940s, Kolsky modified this technique into a three bar (striker-, incident-, and transmitted bar) technique [4], which basically still today is the most widely used setup for dynamic material testing all over the world. The details of the equipment design may vary according to specific needs, e.g., whether testing is conducted in compression, tension, torsion, or under uni-, bi-, or triaxial loading conditions, etc. [5-15]. Experimentally obtained data from dynamic mechanical tests have nowadays become an absolute necessity whether it involves the assessment of a material’s performance in a car, where it is supposed to protect the passengers from a possible impact, or for example the simulation of forming processes, where high strain rates are becoming increasingly common. In each case the evaluation of material models or their implementation into existing codes requires accurate experimental data.

The unstable oil price and even more drastically the economic and financial crises of 2008/2009 have hit the automotive industry hard and are forcing the car manufacturers to improve the fuel efficiency of their vehicles. Therefore, in order for steel to remain the car industry’s dominant material, it is vital to understand the deformation mechanisms of steels as thoroughly as possible, make their behavior more predictable and controllable, and in that way exploit their full potential. To better understand the contribution of phase transformations and twinning to the mechanical behavior of advanced high strength steels, selected DP, TRIP and TWIP steels were studied within the framework of the present thesis. The DP steel represents a steel grade which deforms solely by dislocation glide, where the deformation is concentrated mainly in the soft ferrite matrix while hard martensite islands dispersed in the matrix give rise to the high strength levels. The TRIP steel contains, in addition to the soft ferrite matrix and initial hard phases (usually bainite and possibly small amounts of martensite), a third phase, namely retained austenite, which is responsible for the continuous strain hardening up to very high deformation levels due to the gradually proceeding martensitic phase transformation. Five experimental fully austenitic high manganese TWIP steel grades, all varying slightly in their composition and stacking fault energy (SFE), were produced to study the deformation behavior of steels characterized by extensive deformation twinning, and also to better understand the correlations between stacking fault energy and plasticity mechanisms.

Mechanical tests were performed over a wide range of strain rates and temperatures, utilizing a servohydraulic materials testing machine for the low strain rate experiments and the Hopkinson split bar technique for the high strain rate tests. The temperature during the tests was varied using specially designed temperature control systems. In order to study the interdependence between the mechanical behavior and path dependent
microstructure evolution, electron backscatter diffraction (EBSD), transmission electron microscopy (TEM), and X-ray diffraction were used with emphasis on the TWIP steels. Since no standardized specimen geometry exists for the tensile Hopkinson split bar technique, extensive studies were performed in order to better understand the effects of inertia, wave propagation, and specimen geometry on the test results.
2 MECHANISMS OF PLASTIC DEFORMATION IN METALS

In most cases plastic deformation in metals comes about by ordinary dislocation glide. Dislocation interactions also cause strain hardening of the material, i.e., increase its strength during deformation. However, especially in the case of metastable austenite, whether it is retained austenite as a microstructural minor component or a fully austenitic microstructure, mechanical twinning of crystal planes and various phase transformation processes can make a major contribution to the deformation process or, under specific deformation conditions, even dominate it. From a crystallographic point of view, a close relationship exists between mechanical twinning, martensitic phase transformations, and stacking faults [57].

Even though dislocation glide is the primary mechanism responsible for the change in shape during plastic deformation of most metals, both twinning and phase transformations can produce additional obstacles to dislocation glide and thus reduce the free mean path of dislocations. A reduction in the free mean path of dislocations, in turn, gives rise to higher strain hardening rates. In TRIP and TWIP steels advantage is taken of this phenomenon where martensitic phase transformation and twinning, respectively, occurs during the progress of straining. A certain amount of stress/strain is required to initiate the phase transformation or twinning of crystal planes. Some of the fundamental nucleation parameters for twinning and displacive (martensitic) phase transformations are the stacking fault energy (SFE), grain size, and critical shear stress.

The strain rate and temperature dependent mechanical behavior of metals is mainly controlled by dislocation dynamics. In practice, however, both temperature and strain rate often also affect the competition between dislocation glide, twinning and phase transformations, and thus the overall mechanical behavior as a function of temperature and/or strain rate can be much more complex.

2.1 Stacking faults, ε/α’-martensite and deformation twins in fcc metals

Although in practice the distinction between mechanical twins, martensite phases, and stacking faults via characterization methods may not always be unambiguous, a rather clear crystallographic formalism exists for each of these cases.

2.1.1 Stacking faults

In an fcc crystal, a stacking fault is formed as a consequence of the dissociation of a perfect a/2<110> dislocation into the energetically more favorable configuration of two Shockley partial dislocations of type a/6<211>. The stacking fault is a two dimensional defect formed between the partial dislocations, where the regular stacking sequence of the {111} planes is disturbed. If the stacking sequence of the {111} planes changes from the regular stacking to the ‘faulted’ stacking by removing part of a {111} plane, i.e., ABCABCABC→ABCACABCA, the fault is called an intrinsic stacking fault. In the case
of an extrinsic stacking fault, an additional part of a \{111\} plane is inserted in the crystal and the stacking sequence consequently changes for instance as ABCABCABC → ABCACBCAB, hence containing an excess plane with C stacking. In some cases it can be more convenient to think of an extrinsic stacking fault as the overlapping of two intrinsic stacking faults on successive \{111\} planes.

During the formation of a stacking fault, the Shockley partial dislocations are repelling each other due to a mutual interaction with a net repulsive force. However, with increasing separation \(r\) of the partials also an increase in energy \(E_{SF}\) of the stacking fault is associated,

\[
E_{SF} = \gamma_{SFE} \cdot L \cdot r
\]

where \(L\) is the length of the dislocation and \(\gamma_{SFE}\) the stacking fault energy (per unit area), so that a force opposing the repulsion of the partials is

\[
F_{SF} = -\frac{\partial E_{SF}}{\partial r} = -\gamma_{SFE} \cdot L
\]

The equilibrium separation of the dissociated dislocation is hence controlled by the minimum of the total stored energy in the stacking fault, corresponding to the force balance at the partial dislocations:

\[
F_{tot} = -\frac{\partial E_{tot}}{\partial r} = 0
\]

Based on these considerations Cottrell [48] derived an expression for the width \(w\) of the stacking fault as

\[
w = \frac{Gb_p^2}{4\pi\gamma_{SFE}}
\]

where \(G\) is the shear modulus and \(b_p\) the absolute value of the Burger’s vector of the Shockley partials. From the equation it is apparent that the width \(w\) of the stacking fault is essentially controlled by the stacking fault energy \(\gamma_{SFE}\). \(w\) in turn controls the ease of cross slip of screw dislocations and is thus largely responsible for the prevailing deformation mechanism or deformation type, i.e. planar glide, wavy glide, deformation twinning, or martensitic transformation.

### 2.1.2 \(\epsilon\)-martensite

During the \(\gamma_{fcc} \rightarrow \epsilon_{hcp}\) phase transformation, the close-packed planes and directions in the two structures are parallel, i.e., \(\gamma\{111\} \parallel \epsilon\{0001\}\) and \(\gamma<110> \parallel \epsilon<11\overline{2}0>\), and \(\epsilon\)-martensite forms a hexagonal close-packed (hcp) crystal structure [52]. In terms of \{111\} planes of the cubic crystal structure, the stacking sequence in the \(\epsilon\)-martensite can be expressed as CACA. Therefore, each single intrinsic stacking fault contains a thin layer
of ε-martensite phase. This consideration is also used in the SFE calculation approach proposed by Olson and Cohen [26] (see section 2.4). Brooks et al. [49] proposed that single stacking faults can be regarded as ε-martensite nuclei, while the growth of ‘perfect’ ε-martensite occurs by overlapping of the stacking faults on every second {111} plane [49-51], either by a specific mechanism or by random irregular overlapping processes [51]. The distinction between single stacking faults, bundles of overlapping stacking faults, and faulted or perfect ε-martensite is therefore not quite unambiguous.

2.1.3 α’-martensite

Through comparison of the close packed planes of the fcc and hcp lattice, i.e., the {111}- and the {0001}-plane, respectively, with the most closely packed plane {110} of the bcc lattice, it is evident that the close packed planes of the fcc and hcp lattices can be transformed into the {110} bcc-plane through little distortion. The lattice transformation can, e.g., be described through the Bain transformation, which transforms the fcc lattice through buckling and straining into the bcc lattice. The crystallographic orientation relationship between the γ and the α’ phase obeys the Bain transformation relationship described by γ{111} ≡ α’{001} and γ<100> ≡ α’<110> [53]. Another commonly observed transformation orientation relationship is the Kurdjumov-Sachs transformation, where close-packed planes of the fcc structure transform into the most closely packed planes of the bcc crystal structure, and close-packed directions into close-packed directions, i.e., γ{111} || α’{110} and γ<110> || α’<111> [50, 53].

Many researchers have reported that the nucleation of α’-martensite takes place at the intersections of shear bands [49-50, 54-55], while others have reported nucleation within a single shear band [56-57]. Since in many cases shear bands consist of ε-martensite, it is often considered an intermediate phase in the γ_{fcc} \rightarrow α’_{bct} phase transformation, which can then be written as γ_{fcc} \rightarrow ε_{hcp} \rightarrow α’_{bct}. Whether the crystal structure of the α’-martensite should be considered bcc or slightly tetragonally distorted bcc (bct), depends on the carbon content.

2.1.4 Deformation twins

The classical definition of twinning requires that the lattices of the twin and the matrix are related either by a reflection in some plane, called the twin plane, or by rotation of 180° about some axis, called the twin axis [58]. In crystals of high symmetry, such as fcc, these orientations are equivalent. Deformation twins in fcc crystals form, at least in principle, by a homogeneous shear of the parent lattice along the {111} plane in the <112> direction. Frank [59] proposed a way of creating a twin by stacking or overlapping of intrinsic stacking faults, i.e., by the glide of Shockley partials with identical Burger’s vectors on successive {111} planes. The case of two intrinsic stacking faults overlapping on successive {111} planes is referred to as an extrinsic stacking fault (see section 2.1.1), which can also be considered as the special case of a twin, two atom-layers in thickness. If the overlapping of intrinsic stacking faults proceeds on successive {111} planes, the twin grows in thickness. The thickness of single twins is therefore only of nanometer scale, but stacks of nano-twins can add up to twin bands micrometers in thickness. The
actual substructure in twin bands is lamellae like, where arrays of twins and matrix alternate [88-89].

As opposed to dislocation glide, mechanical twinning is polarized, i.e., the shear displacement during twin formation can take place only in a definite direction but cannot be reversed. Different twin systems, i.e., combinations of twin plane and twin direction are thus active depending on whether the material is loaded in tension or compression. Since in cubic crystal structures there are 12 crystallographically equivalent twin systems, of which at least one is always oriented favorably to support the imposed change in shape, the directional character of twins does not manifest itself in isotropic materials [62, 96]. However, in the presence of texture the twinning stresses in compression and tension can be different.

In most cases the critical event in the twinning process is nucleation, while growth can occur at stresses much lower than the nucleation stress [60-62]. The critical twinning stress can, at least in theory, be formulated in analogy to the critical resolved shear stress \( \tau_{\text{CRSS}} \) for dislocation glide. However, the experimental evidence supporting this theory suffers from large scatter in measured twinning stresses [58]. For instance Thompson and Millard [118] reported a critical resolved shear stress \( \tau_{\text{CRSS}} \) of 13.5 MPa for twinning in cadmium, whereas Bell and Cahn [119] reported values of 41 and 4.9 MPa for the same material. Several critical factors affecting the experimental results can be the reason for this large scatter. On the one hand, the incidence of twinning as well as the twinning-slip competition are very sensitive to orientation and therefore require that very accurate and reproducible orientations in the test specimens are guaranteed. On the other hand, impurity levels and defect structures giving rise to local stress concentrations can have a great effect on the twin nucleation in polycrystals, while in single crystals stress concentration sites of different nature can have a similar effect (e.g. surface notches, internal flaws, etc.). Twinning in fcc metals often nucleates at microscopic defect structures [63] and the local stress at which twinning initiates is considerably higher than the externally applied stress [62]. Twin nucleation mechanisms for fcc metals have been suggested for example by Suzuki and Barrett [65], Haasen and King [66], Bolling and Richman [67], Orowan [68], and others. Another cause for assumingly erroneous twinning stresses presented in the literature is the misinterpretation of load drops in the stress-strain curve, especially at low temperatures, falsely attributed to twinning when they may rather represent adiabatic softening accompanying a local slip avalanche [58, 120].

A relationship between the stacking fault energy \( \gamma_{\text{SFE}} \) and the critical twinning stress \( \tau_{\text{C}} \) has been proposed by Suzuki and Barrett [65] and Venables [69-70], stating that \( \tau_{\text{C}} \) increases with increasing \( \gamma_{\text{SFE}} \). This has proven to be true mostly for fcc metals. Venables [69-70] found a linear relationship for the twinning stress as a function of the square root of \( \gamma_{\text{SFE}} \). Narita et al. [125] have confirmed that \( \tau_{\text{Cu}}^{\text{Cu}} > \tau_{\text{Ag}}^{\text{Ag}} > \tau_{\text{Au}}^{\text{Au}} \) in accordance with \( \gamma_{\text{SFE}}^{\text{Cu}} > \gamma_{\text{SFE}}^{\text{Ag}} > \gamma_{\text{SFE}}^{\text{Au}} \), and Narita and Takamura [126] found a proportional relationship for Ni-Ge alloys, Cu, Au, and Ag according to \( \gamma_{\text{SFE}} = 2b_s \tau_{\text{C}} \), where \( b_s \) is the Burger’s vector.
of a Shockley partial. A more comprehensive model to describe the relationship between $\tau_c$ and $\gamma_{SFE}$ has been proposed for example by Meyers et al. [62].

The effect of grain size on the twinning stress obeys in most cases the Hall-Petch relationship,

$$\sigma_T = \sigma_{T0} + k_T d^{-1/2}$$

where the slope $k_T$, however, is larger than the slope $k_S$ for slip, i.e., the twinning stress is more sensitive to grain size than the slip stress [62, 64]. The reason for the difference is still not fully understood, but Armstrong and Worthington [64] attributed the higher grain size sensitivity of the twinning stress to microplasticity, i.e., to localized dislocation activity at stress levels where the metal is globally in the elastic domain, whereas the yield stress is associated with the onset of global plastic deformation.

2.2 Advanced high strength steels (AHSS): DP, TRIP, and TWIP

The superior mechanical properties of advanced high strength steels (AHSS) over conventional high strength steels (HSS) basically arise from their different microstructure. The microstructure of HSS’s comprises mainly ferrite phase, while AHSS’s are primarily multi-phase steels containing ferrite, martensite, bainite, and/or retained austenite. Being composed of only one phase, the more recently developed high manganese austenitic AHSS (either TRIP or TWIP) represent an exception to the multiphase concept. While in the multi-phase steels the initial quantities of phases other than ferrite are sufficient to produce the desired mechanical properties, the austenitic steels take advantage of mechanical twinning and/or phase transformations during plastic deformation of the material.

The aforementioned beneficial properties of AHSS’s can be of different nature, such as higher strain hardening capacity resulting in improved strength-ductility balance, ultra-high yield and tensile strengths, or the occurrence of a bake hardening effect. The general design criteria for the steels therefore depend on the particular application. In most cases, however, the tradeoff between strength and ductility is the most topical issue. Especially in the automotive industry, complex forming procedures call for materials with improved ductility, which delays the onset of plastic instability during forming and, in addition, ensures crashworthiness and stiffness of the final car body structure.

2.2.1 Dual phase (DP) steel

DP steels possess a two-phase microstructure that in the case of hot-rolled products is achieved by controlled cooling from the austenite phase, and in the case of cold-rolled and hot-dip coated products by intercritical annealing of the cold rolled steel strip followed by quenching to room temperature. During quenching, the austenite in the ferritic-austenitic microstructure transforms to martensite, leading to a ferritic-martensitic microstructure. After quenching, the mobile dislocation density in the soft ferrite phase is very high, resulting in the continuous yielding and strong strain hardening typical of DP
steels [77-78]. Hard martensite islands dispersed in the ferrite matrix are, in turn, responsible for the high strength. The amount of martensite in the final microstructure, and thus the strength of the steel, is mainly determined by the annealing temperature, quenching rate, and chemical composition of the steel. However, the various forming processes during and the possible bake hardening treatments after manufacturing affect the mechanical performance of the final parts and structures, normally increasing their strength but reducing their further deformability.

2.2.2 Transformation induced plasticity (TRIP) steel

TRIP steels’ microstructure usually consists of bainite, martensite, and retained austenite embedded in a continuous soft ferrite matrix. The microstructure is achieved through an isothermal hold below the bainite start temperature during cooling from the intercritical annealing temperature. Furthermore, higher quantities of carbon, silicon and/or aluminum are usually used in comparison to DP steels. Similarly to DP steels, the ferritic matrix is behind the relatively low yield strength of TRIP steels. The initial amount of hard phases, i.e., bainite and martensite, however, is lower than in DP steels, but the presence of retained fcc γ-austenite phase results in an austenite-to-martensite phase transformation during mechanical deformation, and in interactions of deformation and phase transformation mechanisms, which in general explain the high elongations of the TRIP steels as follows [80-81]: At the onset of microscopic instability, the increased stress in the area of localized deformation ‘triggers’ the austenite to martensite transformation, which locally strengthens the material and compensates for the reduction of the cross sectional area. Local necking does not proceed further and deformation shifts to a weaker region, which again locally deforms until the martensitic transformation occurs, and so forth. Macroscopic plastic instability does not occur until the transformation rate of austenite to martensite \( \gamma_{\text{fcc}} \rightarrow \alpha_{\text{bcc}}^{\text{M}} \) is too low to compensate for the reduction of the cross sectional area, or when basically all austenite has transformed to martensite. The strain level at which the retained austenite in the TRIP steel begins to transform to martensite is controlled by the carbon content and the size of the retained austenite grains [82, 97-98].

**Thermodynamics of the martensitic phase transformation and the TRIP effect**

The thermodynamic stability of martensite and austenite is illustrated in Figure 1a [38-40], where the Gibbs energies of martensite and austenite are schematically plotted as a function of temperature. \( M_s \) is the temperature at which martensite starts to form on pre-existing nucleation sites upon appropriate under-cooling, driven by the difference in the chemical free energy of the two phases \( \Delta G_{\gamma \rightarrow \alpha}^{\text{M}} \) [41]. The reaction is essentially complete at a lower temperature \( M_f \), the martensite finish temperature. At temperature \( T_0 \), both phases have exactly the same Gibbs energy \( G_\gamma = G_{\alpha'} \). At temperatures below \( T_0 \) but above \( M_s \), martensite does not form spontaneously as the value of the chemical free energy \( \Delta G_{\gamma \rightarrow \alpha'}^{\text{M}} \) is still insufficient for the transformation. The transformation can, however, occur also under the assistance of a mechanical driving force \( U' \), if this
additional driving force added to the chemical driving force $\Delta G_{M_s}^{\gamma \rightarrow \alpha'}$ equals or exceeds the total free energy difference $\Delta G_{M_s}^{\gamma \rightarrow \alpha'}$, i.e.,

$$\Delta G_{M_s}^{\gamma \rightarrow \alpha'} = \Delta G_{M_s}^{\gamma \rightarrow \alpha'} + U'$$

(6)

assuming that the critical driving force needed to trigger martensitic transformation remains constant over the temperature range of interest, i.e., $\Delta G_{M_s}^{\gamma \rightarrow \alpha'} = \Delta G_{M_s}^{\gamma \rightarrow \alpha'}$ [38-39, 42, 46-47]. As depicted in Figure 1a, the chemical driving force $\Delta G_{M_s}^{\gamma \rightarrow \alpha'}$ for the martensitic phase transformation decreases with increasing temperature, and consequently the required additional mechanical driving force $U'$ increases.

At temperatures higher than $M_S^\sigma$, the stress needed to initiate the martensite transformation exceeds the yield strength of the parent austenite phase, and the phase transformation can only take place after some amount of plastic deformation in the austenite phase [41]. The temperature region $M_S^\sigma \leq T \leq M_d$ is therefore usually referred to as the strain-induced transformation regime, while $M_S \leq T \leq M_S^\sigma$ is called the stress-assisted transformation regime [40]. In addition to the increasing flow stress due to strain hardening in the strain-induced transformation regime, deformation induced defects may also have an influence on the progress of the martensite transformation, e.g., by providing additional nucleation sites [38, 43-45].

At temperatures exceeding $T_0$, the chemical driving force opposes the $\gamma_{fcc} \rightarrow \alpha'$ transformation, and as a consequence, the mechanical component $U'$ has to be larger than the chemical driving force $\Delta G_{M_s}^{\gamma \rightarrow \alpha'}$. This, in turn, requires increasingly higher flow stresses, which can be achieved, in principle, only by substantial permanent strain hardening and/or transient strain rate hardening. With increasing temperature, however, the flow stress of the austenite phase tends to decrease at any strain rate, and above $M_{h}$ the required additional driving force can no longer be provided by the mechanical stress, and the deformation occurs by ordinary dislocation slip and, possibly, by twinning. Figure 1b illustrates the stress levels needed for the martensitic phase transformation to initiate, as well as the governing phase transformation modes at different temperatures [40].
Figure 1. Gibbs energy of martensite and austenite [after 38-40] (a) and the critical stress for the initiation of the martensite transformation (b) as a function of temperature [40].

2.2.3 High manganese twinning induced plasticity (TWIP) steel

In TWIP steels, the fully austenitic microstructure can be retained by means of high level alloying with elements such as manganese, aluminum, and silicon. Manganese is an austenite stabilizer, which in high concentrations drastically lowers the transformation temperature of austenite and increases the carbon solubility in austenite. Al and Si are mainly used to adjust the magnitude of the stacking fault energy $\gamma_{SFE}$ of austenite, but they also strengthen the steel by solid solution hardening and stabilize austenite due to their ability of slowing down the precipitation of carbides, especially cementite, leaving more carbon available for the enrichment of austenite [83]. Additions of Al generally increase $\gamma_{SFE}$ [84], while Si is reported to lower it at concentrations approx. $c_{Si} \geq 4$ wt% [85] but to increase it at low concentrations [34].

The explanation for the achieved enhanced elongations with twin formation is in principle the same as for the TRIP effect, i.e., as strain increases the formation of mechanical twins reduces the mean free path of dislocations. This was confirmed already in 1960 by Babyak’s [86] studies on low stacking fault energy brass, where twin interfaces were shown to act as strong barriers to dislocation motion with an effect of the same order of magnitude as that of grain boundaries. Similar observations in thin layered stainless steels as well as in Inconel 600 alloys were reported by Zhang et al. [87] and Tao et al. [88], respectively. During ongoing deformation, dislocations pile up at twin boundaries just like at grain boundaries. Twinning thus refines the microstructure by forming sub-boundaries that are associated with Hall–Petch type of strengthening under progressing strain [90-91, 95].
2.3 Dislocation dynamics

According to Isaac and Granato [99], the plastic deformation behavior of a material can generally be divided into three strain rate regimes depending on the limiting factor in dislocation kinetics. At low and intermediate strain rates, dislocation motion is controlled by thermally activated processes. For most metals, the flow stress increases rather moderately with increasing strain rate in this regime. Increasing the strain rate above roughly $\dot{\varepsilon} = 10^3 \text{ s}^{-1}$, where the exact value depends on the material, a transition to dislocation drag controlled deformation mechanisms takes place and limits the dislocation velocity, which causes a steep upturn in the strength values. The two regions are illustrated in Figure 2, where flow stress values of OFHC copper are plotted as a function of strain rate [111]. At even higher strain rates the dislocation velocity approaches the velocity of transverse sound waves and relativistic effects come into play [99-100].

![Figure 2. The flow stress of OFHC copper at 15 % plastic strain as a function of strain rate [111].](image)

2.3.1 Thermally activated dislocation motion

At temperatures clearly below the melting point, the plasticity of a metal is determined by the mobility of its dislocations. Obstacles getting in the way of moving dislocations disturb the mobility and thereby increase the strength of the material [100-101].

The periodic potential of the crystal lattice acts as an inherent barrier that any dislocation encounters with the periodicity of the lattice. The force opposing the dislocation movement in this case is the Peierls-Nabarro force, which acts on the atomic level. Most real crystal lattices, however, contain a number of different lattice defects, which form further obstacles to gliding dislocations. Depending on the width of the energy barrier, a distinction between short-range and long-range obstacles is made.
Obstacles can be overcome if the energy that drives the dislocation is high enough to lift the dislocation over the energy barrier. In general, the necessary energy can be provided by two possible sources: externally applied stress and/or thermal vibration energy [100-103]. While externally applied stress can provide the necessary energy to surmount both short- and long-range obstacles, thermal vibration energy can only assist in overcoming short-range obstacles. Obstacles of long-range nature are therefore also referred to as athermal obstacles and short-range obstacles as thermal obstacles. In other words, a thermal obstacle can be overcome by the assistance of thermal energy if the sum of the thermal energy $\Delta G_T$ and the mechanical energy $\Delta G_M$ provided by the external stress equals, or is bigger, than the magnitude of the energy barrier $\Delta G$ of the thermal obstacle, i.e.,

$$\Delta G = \Delta G_M + \Delta G_T$$

An increase in the material’s thermal energy is equivalent to an increase in the amplitude of the vibration of the atoms in the lattice, and above a certain temperature, usually denoted by $T_C$ (c.f. Figure 3), is alone sufficient to overcome all thermal obstacles. The probability for the occurrence of a fluctuation $\Delta G_T$ can be derived from statistical mechanics and is described by an Arrhenius type equation as [100]

$$p_B = \exp\left(-\frac{\Delta G_T}{k_B T}\right)$$

where $k_B$ is the Boltzmann’s constant. Equation 8 is basically the key in describing the temperature dependence of any crystalline material’s plastic deformation behavior; with increasing thermal energy $E_T = k_B \cdot T$, the probability for a dislocation to overcome a thermal obstacle increases.

### 2.3.2 Strain rate dependent mechanical behavior due to thermally activated dislocation motion

From further considerations regarding the theory of thermally activated dislocation motion, also the strain rate dependent deformation behavior of metals can be explained. The thermal energy term $\Delta G_T$ in Equation 7 is only provided with certainty at a given temperature if an infinite period of time is available. In practice, however, dislocations are vibrating with a frequency $\nu_0$, which according to Kocks et al. [104] is by about the factor $b/4l$ (Burger’s vector $b$ and length of a dislocation segment $l$) or two orders of magnitude smaller than the vibrational frequency of atoms $\nu_A$, i.e., $10^{11}$ s$^{-1}$. A dislocation can therefore only once within the period of an oscillation ‘make an attempt’ to ‘jump’ over the barrier ahead with the probability $p_B$ of success, such that a frequency with which a dislocation successfully overcomes an obstacle can be defined as [100]

$$\nu_1 = \nu_0 p_B = \nu_0 \exp\left(-\frac{\Delta G_T}{k_B T}\right)$$
The time $\Delta t$ that it takes for a dislocation to travel a certain distance $\Delta l$ can be divided into the running time $t_r$ and the waiting time $t_w$ in front of the obstacles that the dislocation encounters as

$$\Delta t = t_r + t_w$$  \hspace{1cm} (10)$$

The waiting time $t_w$ is the reciprocal of the frequency $v_i$, with which the dislocation overcomes an obstacle, i.e.,

$$t_w = v_i^{-1} = v_0^{-1} \exp \left( \frac{\Delta G_T}{k_B T} \right)$$  \hspace{1cm} (11)$$

The macroscopic strain rate for the theoretical case where only one slip system is active$^1$ in a material can be expressed as a function of microscopic quantities by the so-called Orowan equation

$$\frac{\partial \dot{\varepsilon}}{\partial t} = \dot{\varepsilon} = \frac{1}{M} \rho b \bar{v}$$  \hspace{1cm} (12)$$

where $M = (\cos \lambda \cdot \cos \kappa)^{-1}$ is the orientation factor with $\lambda$ and $\kappa$ the angles between the direction of applied stress and the glide direction and the glide plane normal, respectively, $\rho$ the dislocation density, $b$ the Burger’s vector, and $\bar{v}$ the average dislocation velocity. From Equation 12 and $\bar{v} = \Delta \bar{v} / \Delta t$ it follows that

$$\dot{\varepsilon} \propto \frac{\Delta \bar{v}}{\Delta t}$$  \hspace{1cm} (13)$$

and hence

$$\Delta t = \Delta t_r + \Delta t_w \propto \frac{\Delta \bar{v}}{\dot{\varepsilon}}$$  \hspace{1cm} (14)$$

Thus, assuming that a steady state between running and waiting dislocations has been reached and that the average waiting time of a dislocation is much longer than the running time between subsequent obstacles, it is obvious that the higher the strain rate, the less time is available for a dislocation waiting in front of an obstacle to receive the extra amount of thermal energy $\Delta G_T$. The consequence of this is the commonly observed increase in strength with increasing strain rate.

$^1$ Since in practice many slip systems are simultaneously active in a crystal, Equation 12 should actually be expressed in a tensorial form as $\dot{\varepsilon}^p = \sum_{k=1} \nu_k \hat{m}^k \otimes \hat{n}^k$, [187].
Taking into account the temperature and strain rate dependence of the thermally activated dislocation motion, the flow stress of a material can be expressed as [100]

\[ \tau_C = \tau_A(\text{structure}) + \tau^*(T, \dot{\varepsilon}, \text{structure}) \]  

(15)

where \( \tau_C \) is the critical shear stress, \( \tau_A \) is the athermal component of the shear stress due to long range obstacles, i.e., the part that can only be overcome by pure mechanical stress, and \( \tau^* \) is the amount of externally applied stress that is needed to overcome the thermal obstacles. In other words, \( \tau_T = \tau_C(0K) - (\tau_A(T) + \tau^*(T)) \) is the thermal stress equivalent, by the amount of which the critical shear stress is lowered due to the assistance of thermal energy.

The combined effect of temperature and strain rate on the flow stress of a metal is illustrated in Figure 3 [100]. Above a critical temperature \( T_C \) the thermal energy is sufficient to overcome all thermal obstacles. Additional thermal energy cannot contribute to the process of passing obstacles since the character of all remaining obstacles is of athermal nature. The flow stress is controlled only by the athermal component of the shear stress \( \tau_A \). The athermal component of the shear stress \( \tau_A \) at 0K is obtained by linear extrapolation of the flow stress curve from above \( T_C \) to 0K. The slight temperature dependence of \( \tau_A \) is due to the temperature dependence of the shear modulus \( G \), which affects the shear stress acting on a bowing dislocation segment of length \( l \) (average distance between obstacles) via the relationship expressed by the Orowan equation, i.e.,

\[ \tau = \frac{Gb}{l} \]  

(16)

where \( b \) is the Burger’s vector of the dislocation. The value of \( l \) depends in each case on the characteristics of the obstacle structure.

At 0K, no thermal energy is available and the entire flow stress described by the sum of the athermal component \( \tau_A \) and the thermal component \( \tau^* \) has to be provided by mechanical energy. For \( 0K < T \leq T_C \), the flow stress decreases distinctly with increasing temperature due to the growing thermal contribution in overcoming short-range obstacles. In this temperature regime also the strain rate sensitivity takes effect.
While the Peierls-Nabarro stress is the principal short-range obstacle in bcc metals, forest dislocations are the rate controlling obstacles in fcc metals. The different nature of these obstacles is mainly responsible for the difference in the strain rate sensitivity between these two crystal structures.

2.3.3 Dislocation drag

The theory of thermally activated dislocation motion described in the previous subsections is valid only up to a certain strain rate, the exact value of which depends on the particular material. At higher strain rates, a steep upturn in the stress levels as seen in Figure 2 can usually be observed. For crystalline materials, this effect is generally attributed to so-called viscous dislocation drag mechanisms [100], which start to act in addition to the thermal obstacles discussed in the previous subsection, and finally dominate the deformation behavior at very high strain rates. The interactions between the matrix and moving dislocations causing these drag effects can originate from different mechanisms. The most important mechanisms are summarized under the terms of phonon viscosity and electron viscosity. At temperatures around room temperature, phonon viscosity mechanisms are the dominating viscous drag effects. Moving dislocations experience frictional forces due to the quantized modes of vibration occurring in the crystal lattice, causing the dissipation of the dislocations’ kinetic energy. At decreasing temperatures, the mechanism of viscous electron drag gains importance and dominates at very low temperatures. Electron drag is due to the interaction of propagating elastic waves with conduction electrons. The magnitude of the drag effect is highly dislocation velocity dependent.

No actual consensus on the role and details of the drag effects, however, has been reached among researchers, and the exact reasons for the sudden increase in the flow stress levels with increasing strain rate are also still under some debate [100, 105-110]. Studies on the strain rate sensitivity of copper, copper-aluminum, and austenitic stainless steels by measurements of the mechanical threshold stress, coupled with analysis of the kinetics of dislocation-obstacle interactions, led Follansbee [110] to the conclusion that the increase in flow stress values at strain rates less than $10^4 \text{ s}^{-1}$ arise from the strain rate
sensitivity of structure evolution rather than from a change in the transient strain hardening mechanism. Either way, the phenomenological relation between (shear) stress $\tau$ and (shear) strain rate $\dot{\gamma}$ in this regime can basically be described by the equations of a Newtonian viscous fluid such that a linear relationship between stress and strain rate arises as

$$\tau = \frac{B \dot{\gamma}}{\rho_m b^2}$$  \hspace{1cm} (17)

where the linear proportionality factor $B$ is called the drag or damping coefficient, and $\rho_m$ is the dislocation density. The drag coefficient $B$ appears to be not exactly constant up to very high dislocation velocities but rather to increase with increasing velocity due to the growing relativistic contribution to the damping term. Gillis et al. [112] proposed to take account for this deviation from Newtonian behavior by the empirical expression

$$B = B_0 \frac{1}{1 - v_s^2 / v_S^2}$$  \hspace{1cm} (18)

where $v_s$ is the shear wave velocity and $B_0$ the viscosity at rest. Later theoretical calculations by Roos [170] are in good agreement with the empirically determined drag coefficient.

### 2.3.4 Effects of strain rate and temperature on phase transformations and twinning

The stability of the austenite and martensite phases in terms of thermodynamics was already described in section 2.2.2. The change in the phase fractions of austenite and martensite as a function of strain due to the $\gamma_{fcc} \rightarrow \alpha'$ reaction, i.e., $\partial V_{\gamma \rightarrow \alpha'} / \partial \varepsilon$, in turn, depends on both temperature and strain rate, and therefore the contribution of the phase transformation to the overall mechanical performance of the material also strongly depends on these variables. Berrahmoune et al. [113] reported an increase in the phase transformation rate in multiphase TRIP steels with decreasing temperature, which is in agreement with the increase in the driving force upon cooling. Van Slycken et al. [114] and Wei et al. [115] found that the $\gamma_{fcc} \rightarrow \alpha'$ phase transformation in multiphase TRIP steels was partly suppressed by an increase in strain rate which, however, can be attributed to the secondary effect of adiabatic heating of the specimen during high strain rate testing. No conclusions on the isolated effect of strain rate can therefore be drawn from these studies. Studies on TRIP steels by Oliver et al. [116] revealed that more austenite transformed to martensite when the strain rate was increased from quasi-static to intermediate strain rates. However, reliable studies on the effect of the strain rate on the $\gamma_{fcc} \rightarrow \alpha'$ transformation kinetics are complicated due the fact that strain is not a true state variable (see also section 2.3.5).

A further challenge is the exact determination of phase fractions. Techniques by which the volume fractions of phases present in the microstructure can be determined are, for
example, etching techniques combined with optical microscopy, X-ray and neutron diffraction, electron backscatter diffraction (EBSD), and magnetic measurements. However, each of these techniques underlies individual limitations in accuracy, and due to the different nature of these techniques, no comparison between the absolute values obtained by the different techniques is possible. XRD and EBSD, for example, are surface sensitive techniques but differ in their interaction volume, while magnetic measurements usually cover the whole bulk of the sample. Zhao et al. [185], for example, conducted a study on the comparability between XRD and magnetic measurements for austenite phase fraction determination, and found a significant difference between the results obtained by these two techniques.

As opposed to the martensitic phase transformation, mechanical twinning does not occur spontaneously upon undercooling but is a purely mechanical deformation mechanism. However, the twinning propensity is strongly temperature and strain rate dependent. The contribution of twinning to the deformation in most metals increases as the temperature is lowered. This can be explained by the fact that the twinning stress in most materials increases less rapidly with decreasing temperature than the yield or flow stress for the plastic deformation by slip. For materials with yield stresses less sensitive to temperature, such as fcc metals, the temperature effect is usually less obvious. However, the flow stress may ultimately reach the twinning stress by work hardening. In that case, when lowering the temperature, the slip-to-twinning transition is governed, at least partly, by the strong temperature dependence of the work hardening rate [58]. Based on their studies on copper deformed at 4K, Blewitt et al. [117] gave the first unambiguous evidence for twinning in a pure fcc material in 1957. In low stacking fault alloys, such as many copper alloys and austenitic steels, twinning at or above room temperature is, on the other hand, quite common.

Strain rate and temperature effects in material behavior are usually coupled by an Arrhenius type equation, which is characteristic for thermally activated processes [58]. High sensitivity to changes in temperature thus also implies high sensitivity to imposed strain rate, and vice versa. This is usually also observed for twinning, where a general equivalence of low temperatures and high strain rates exists. However, in the competition between slip and twinning a rather weak temperature sensitivity prevails for the substitution of slip by twinning, but a very distinct strain rate sensitivity is generally observed [58]. All bcc and hcp materials, as well as many fcc materials deform solely by twinning when deformed under shock-loading. Fcc metals with a high stacking fault energy, such as aluminum, do not deform by twinning under normal deformation conditions, but twinning has been also observed in shock-loaded Al-Mg alloys [121]. The critical twinning stress for most metals seems to be rather temperature insensitive. There are, however, also reports of a gradual increase in the twinning stress with increasing temperature, i.e., positive temperature sensitivity, and at the same time a decrease in the twinning stress with increasing strain rate [41, 122, 123], both changes being opposite to those for the case of thermally activated dislocation motion. Reed-Hill [124] suggested that twinning always has this intrinsic dependence but only when twinning is the main mode of deformation, this behavior will be observed. In other words, in cases where slip and twinning occur together but slip produces the major part of strain, the temperature
and strain rate dependence of twinning will be masked by that of slip. A possible
terpretation for the positive temperature and negative strain rate sensitivity of twinning
was suggested by Bolling and Richman [67]. They argued that a lenticular twin
accommodated by the matrix is always accompanied by stress concentrations near its
edge, which results in slip in this region. The applied stress needed to propagate the twin
will increase together with the slip activity. For the occurrence of slip, the theory of
thermally activated dislocation motion is valid for the temperature and strain rate
dependences as usual, but as far as twinning is concerned, the dependences are inverted.

By experiments it has been shown that twins often form with an effective interface
velocity that is only fractions lower than the velocity of sound and thus much faster than
the formation of slip bands. Yet, the reasons for the difference in the formation speeds are
not fully understood, but if twinning is accompanied by dislocation slip, also unusually
high dislocation mobility is required for twin formation [58].

2.3.5 The strain rate sensitivity factor

Various definitions of strain rate sensitivity are commonly used in the literature. In any
case, the apparent strain rate sensitivity is usually determined from the region of
thermally activated dislocation motion by fitting linear regression lines to the yield
strength or UTS data points over the strain rate, such that the strain rate sensitivity factor
is either obtained as \( m = d \log \sigma / d \log \dot{\varepsilon} \) or \( m = d \sigma / d \log \dot{\varepsilon} \) (also the reciprocal value of
the first definition is regularly used). The first definition, however, is a measure of the
relative strain rate sensitivity, as opposed to the second definition, which describes the
absolute strain rate sensitivity, as can be shown by simple mathematical manipulations
[127]

\[
m = \frac{\Delta \log \sigma}{\Delta \log \dot{\varepsilon}} = \frac{\log \sigma_2 - \log \sigma_1}{\log \dot{\varepsilon}_2 - \log \dot{\varepsilon}_1} = \frac{\log \sigma_2}{\log \dot{\varepsilon}_2} = \frac{\log \sigma_1}{\log \dot{\varepsilon}_1}
\]

and

\[
m = \frac{\Delta \sigma}{\Delta \log \dot{\varepsilon}} = \frac{\sigma_2 - \sigma_1}{\log \dot{\varepsilon}_2 - \log \dot{\varepsilon}_1} = \frac{\sigma_2 - \sigma_1}{\log \dot{\varepsilon}_2} = \frac{\sigma_1}{\log \dot{\varepsilon}_1}
\]

The denominators in both cases are the same, whereas the nominators are fundamentally
different. In the first case the strain rate sensitivity depends on the relative change of
stress, whereas in the second case it depends on the absolute change of stress. For the
assessment of the material behavior in, e.g., a car crash event, or for modeling purposes,
the presented ‘apparent’ m-values provide easily obtainable and very useful engineering
values.
However, when obtained from high strain rate experiments, where material is deformed
to high deformation degrees in a single deformation step, multiple deformation
mechanisms are incorporated in a single test and both definitions are therefore strictly
speaking violating the mechanistic formulations in terms of what strain rate sensitivity
(SRS) originally stands for [128]. SRS in its original definition is defined by the partial
derivative \( m = \frac{d \ln \sigma}{d \ln \dot{\varepsilon}} \bigg|_{\varepsilon,T,S} \), where strain \( \varepsilon \), absolute temperature \( T \), and structure \( s \),
are held constant. To obtain data suitable for studies of micro-mechanisms only, the
differential response quantity \( m = \frac{d \ln \sigma}{d \ln \dot{\varepsilon}} \) should be measured by nearly
infinitesimal changes in strain rate while holding temperature, strain, and structure
constant, avoiding changes in deformation history between samples and eliminating
strain hardening and temperature changes experienced by different samples of the test
material [128]. These considerations are basically a consequence of the dependences
described above by Equation 15 and the fact that strain is not a state variable. The
microstructure is history dependent, i.e., dependent on the deformation path by which it
was obtained (at what temperature(s), strain rate(s), etc.).

2.4 Stacking fault energy

The deformation mechanisms and mechanical properties of fcc metals are strongly
related to their stacking fault energy (SFE) \( \gamma_{\text{SFE}} \). The dimensionless parameter \( \gamma_{\text{SFE}}/Gb \),
where \( G \) is the shear modulus and \( b \) the slip distance, is a measure for the ease of cross
slip of screw dislocations and therefore determines the work hardening behavior in stage
III of deformation [24]. To fully exploit the twinning and/or phase transformation mecha-
nisms in austenitic steels, the magnitude of \( \gamma_{\text{SFE}} \) has to be properly adjusted by choosing
the right amount of chemical alloying elements. At low \( \gamma_{\text{SFE}} \), wide dissociation of disloca-
tions into Shockley partials can hinder dislocation glide and thus favor mechanical twinning
(\( \gamma \rightarrow \gamma'_T \)) or martensitic phase transformation (\( \gamma_{\text{fc}} \rightarrow a_{\text{bfc}}^{18} \) or \( \gamma_{\text{fc}} \rightarrow a_{\text{hcp}}^{18} \rightarrow a_{\text{bfc}}^{18} \)).

According to Frommeyer et al. [72], martensitic phase transformation in high manganese
TWIP steels occurs under straining if the SFE is about \( \gamma_{\text{SFE}} \leq 16 \text{ mJ/m}^2 \) while the molar
Gibbs energy value of the martensitic reaction \( \gamma_{\text{fc}} \rightarrow a_{\text{bfc}}^{18} \) has a negative value of
\( \Delta G^{\gamma\rightarrow\gamma'} \approx -220 \text{ J/mol} \) or less, depending on the composition [16]. Similar SFE ranges,
i.e., \( \gamma_{\text{SFE}} \leq 18 \text{ mJ/m}^2 \) and \( \gamma_{\text{SFE}} \leq 20 \text{ mJ/m}^2 \) for the martensitic reaction were reported by
Allain et al.[18] and Grässel et al. [71], respectively. For stacking fault energies of the
order \( \gamma_{\text{SFE}} = 25 \text{ mJ/m}^2 \) and a molar Gibbs energy positive of \( \Delta G^{\gamma\rightarrow\gamma'} \approx 110 \) to 250 J/mol,
twining becomes the dominant deformation mechanism according to Frommeyer et al.
[72] and Grässel et al. [71]. Values of the same order of magnitude have been reported by
Allain et al. [18]. The twinning stress therefore increases with increasing \( \gamma_{\text{SFE}} \) [65, 69-70].
At even higher stacking fault energies of approximately \( \gamma_{\text{SFE}} \geq 45 \text{ mJ/m}^2 \), plasticity is
con-trolled solely by dislocation glide.

Different approaches to determine the SFE of metals are presented in the literature. First
calculations of the SFE in fcc metals were based on the electron theory in metals [21-22].
Later Cotterill et al. [23] calculated the SFE of copper by a variational method using a
central-force approximation, where the pairwise interaction between discrete atoms was
represented by a Morse potential function. Olson and Cohen [26] proposed a thermody-
namic modeling approach to calculate the SFE. With the extended nodes method, Whelan [25] presented an experimental approach to measure the SFE based on electron diffraction measurements in transmission electron microscopy. The line profile analysis of XRD spectrum is another experimental technique [92-94].

2.4.1 Thermodynamic Stacking Fault Energy Calculations

In the present studies, thermodynamic SFE calculations were performed based on the modeling approach originally proposed by Olson and Cohen [26] and later adopted and modified by various authors.

**Thermodynamic model**

The SFE was calculated for the Fe-Mn-Al-Si-C system over a wide range of temperatures based on the thermodynamic modeling approach originally proposed by Olson and Cohen [26]. The model accounts for the change in the Gibbs energy \( \Delta G_{\rho}^{\gamma \rightarrow \epsilon} \) of each element \( \varphi \) upon the \( \gamma_{\text{fcc}} \rightarrow \epsilon_{\text{hcp}} \) phase transformation, as well as for the first order interactions between elements \( \varphi \) and \( \phi \) by introducing the excess free energy term \( \Omega_{\varphi \phi}^{\gamma \rightarrow \epsilon} \). The change in the Gibbs energy due to magnetic contributions to the Gibbs energy \( G \), including paramagnetic to antiferromagnetic transitions of the \( \gamma \)- and \( \epsilon \)-phases, is also included in the model.

![Figure 4](image.png)

**Figure 4.** A stacking fault in an fcc crystal described by stacking operators, where B layer over A, C over B, and A over C is denoted by ▶, and C over A, B over C, and A over B by ◀. In this classical approach, two matrix areas are rigidly displaced from each other. An infinite stacking fault is illustrated in (a), while (b) illustrates a terminated stacking fault bounded by two partial dislocations [20].
The formation of a stacking fault in terms of crystallography is described in section 2.1.1. The change in the stacking sequence produces an interface separating two regions of the matrix as illustrated in Figure 4. Due to the change in the chemical bonding, the stacking fault can be treated as a specific case of a Gibbs interface [20]. Alternatively, the stacking fault can be treated as a separate phase with a different stacking (hcp) and two interface boundaries parallel to the faulted plane, as illustrated in Figure 5. The interfaces result from a difference in the layer spacing between the layers in the matrix and in the stacking fault phase (SFP) [27].

When introducing a stacking fault into the undisturbed stacking sequence of a perfect crystal, a change in the Gibbs energy $\Delta G$ will occur, the magnitude of which must be the same regardless of the approach, i.e., interface approach or volume approach:

$$\Delta G_{if} = \Delta G_{vol}$$  \hspace{1cm} (21)

The change in the Gibbs energy in the case of the interface approach can be expressed as

$$\Delta G_{if} = A \gamma^I$$  \hspace{1cm} (22)

where $A$ is the interfacial area and $\gamma^I$ the interfacial energy per unit area of interface. The change in the Gibbs energy based on the volume approach, on the other hand, can be expressed as

$$\Delta G_{vol} = V^{SFP} \left( G_{v}^{SFP} - G_{v}^{M} \right) + 2A \sigma$$  \hspace{1cm} (23)

where $V^{SFP}$ is the volume of the SFP, $G_{v}^{SFP}$ the Gibbs energy per unit volume of SFP, $G_{v}^{M}$ the Gibbs energy per unit volume of matrix, and $\sigma$ the interfacial energy per unit area of phase boundary. By substituting Equation 22 and 23 into Equation 21 and differentiating Equation 21 with respect to $A$,

---

**Figure 5.** A stacking fault in an fcc crystal described by stacking operators. In this alternative model approach the stacking fault is considered as a separate phase consisting of two layers with a different stacking sequence. An infinite stacking fault is illustrated in (a), while (b) illustrates a terminated stacking fault bounded by two partial dislocations.[20]
one gets an expression for the ideal stacking fault energy

$$\gamma^\infty = 2s(G_{s}^{\text{SFP}} - G_{v}^{M}) + 2\sigma$$  \hspace{1cm} (25)$$

where \(s\) is the interplanar spacing between close-packed planes parallel to the fault plane.

In most cases it is more convenient to work with molar Gibbs energies. The ideal stacking fault energy is then expressed as

$$\gamma_{\text{SFE}} = 2\rho \Delta G_{\gamma^\infty} + 2\sigma$$  \hspace{1cm} (26)$$

with \(\rho\), the molar surface density along \{111\} planes and \(\Delta G_{\gamma^\infty}\), the molar Gibbs energy of the \(\gamma\rightarrow\varepsilon\) phase transformation. The molar surface density \(\rho\) can be calculated by using the lattice parameter \(a\) and the Avogadro’s constant \(N\) as

$$\rho = \frac{4}{\sqrt{3}} \frac{1}{a^2 N}$$  \hspace{1cm} (27)$$

Strictly speaking, one has to distinguish between infinite stacking faults and terminated stacking faults, which are associated with the ideal SFE \(\gamma^\infty\) and the effective SFE \(\gamma^*\), respectively [19-20]. The previous derivations used the ideal SFE \(\gamma^\infty\), which is defined as the energy per unit area of an infinitely extended stacking fault in an infinite ideal crystal. In the following, when using the term \(\gamma_{\text{SFE}}\), it refers to the ideal SFE \(\gamma^\infty\). The effective SFE \(\gamma^*\), on the other hand, is an attempt of modeling a situation closer to the real case where the fault is bounded by partial dislocations (as illustrated in Figures 4b and 5b), which introduces an additional strain field. Due to the additional energy associated with the strain field, the effective SFE \(\gamma^*\) is by a term \(2s\Delta G_S\) higher than the ideal SFE \(\gamma^\infty\), i.e.,

$$\gamma^* = 2\rho \Delta G_{\gamma^\infty} + 2\sigma + 2s\Delta G_S$$  \hspace{1cm} (28)$$

where \(\Delta G_S\) is the strain energy per unit volume of the stacking fault. The strain energy term \(\Delta G_S\) is shape dependent, i.e., sensitive to dislocation configurations, presence of nearby defects, grain size, etc. [28]. Therefore \(\gamma^*\) is not an intrinsic material property as opposed to \(\gamma^\infty\).

Müllner et al. [19] have estimated \(\Delta G_S\) for the particular case of two parallel partial dislocations as

$$\Delta G_S = \frac{\pi D \varepsilon^2}{2} = \frac{\mu \varepsilon^2}{4(1-\nu)}$$  \hspace{1cm} (29)$$
where \( D = \mu / 2\pi(1 - v) \), \( \mu \) the shear modulus, \( v \) Poisson’s ratio, and \( \varepsilon \) the strain normal to the fault plane.

**Calculation of the change in Gibbs energy**

The change in the molar Gibbs energy \( \Delta G^{\gamma \rightarrow \varepsilon} \) upon the \( \gamma_{fc} \rightarrow \varepsilon_{hcp} \) phase transformation can be written as a sum of three terms

\[
\Delta G^{\gamma \rightarrow \varepsilon} = \sum_i \chi_i \Delta G_i^{\gamma \rightarrow \varepsilon} + \sum_{ij} \chi_i \chi_j \Omega_{ij}^{\gamma \rightarrow \varepsilon} + \Delta G_{mg}^{\gamma \rightarrow \varepsilon} \tag{30}
\]

with \( \chi \) the molar fractions of the pure alloying elements. The first term sums over the chemical contributions to the change in Gibbs energy due to the pure element fractions \( \chi_{\nu} \), where

\[
\Delta G_{\nu}^{\gamma \rightarrow \varepsilon} = (G_{\nu}^{\varepsilon} - G_{\nu}^{\gamma}) \tag{31}
\]

with \( G_{\nu}^{\varepsilon} \) and \( G_{\nu}^{\gamma} \) the molar Gibbs energy of the pure elements in the \( \varepsilon \)-phase state and the \( \gamma \)-phase state, respectively. The Gibbs energy is represented as a power series in terms of temperature \( T \) as

\[
G_{\nu}^{\Phi} = a + bT + cT \ln(T) + \sum dT^n \tag{32}
\]

where \( a, b, c, \) and \( d \) are coefficients, and \( n \) represents a set of integers, typically taking the values 2, 3, and -1 [33]. The second term in Equation 30 sums over the excess free energies, which according to a subregular modeling approximation can be expressed as

\[
\Omega_{\nu\nu}^{\gamma \rightarrow \varepsilon} = (0L_{\varepsilon}^{e} - 0L_{\gamma}^{e}) + (1L_{\varepsilon}^{e} - 1L_{\gamma}^{e})(\chi_{\nu} - \chi_{\phi}) \tag{33}
\]

with \( 0L_{\nu}^{e} \) a linear function of the temperature, and \( 1L_{\nu}^{e} \) a constant [29-30]. The last term in Equation 30, \( \Delta G_{mg}^{\gamma \rightarrow \varepsilon} \), accounts for the molar Gibbs energy due the magnetic contribution of each phase \( \Phi \). The magnetic contribution \( G_{mg}^{\Phi} \) to the Gibbs energy \( G \) of phase \( \Phi \) can be described by Hillert and Jarl’s [31] modification of the model proposed by Inden [32] as

\[
G_{mg}^{\Phi} = RT \ln(\beta^{\Phi} + 1)f^{\Phi}(\tau^{\Phi}) \tag{34}
\]

where \( R \) is the gas constant, \( T \) temperature, and \( \beta^{\Phi} \) the magnetic moment of phase \( \Phi \) divided by the Bohr magneton \( \mu_B \). \( f^{\Phi}(\tau^{\Phi}) \) is a polynomial function of the scaled Néel temperature \( \tau^{\Phi} = T / T_{Neel}^{\Phi} \), which was proposed by Hillert and Jarl [31]. Since for decreasing temperatures both the \( \varepsilon \)-phase and the \( \gamma \)-phase undergo a paramagnetic to anti-
ferromagnetic state change upon passing the Néel temperature, the function $f^\phi(\tau^\phi)$ contains a conditional statement which accounts for the change in the magnetic state. For $\tau^\phi \leq 1$, $f^\phi(\tau^\phi)$ is given by

$$f^\phi(\tau^\phi) = 1 - \frac{79\tau^{-1}}{140p} + \frac{474(1/p - 1)}{497} \left( \frac{\tau^5}{6} + \frac{\tau^9}{135} + \frac{\tau^{11}}{600} \right) / D$$  \hspace{1cm} (35)$$

and for $\tau^\phi > 1$ by

$$f^\phi(\tau^\phi) = \left[ \frac{\tau^{-5}}{10} + \frac{\tau^{-15}}{315} + \frac{\tau^{-25}}{1500} \right] / D$$  \hspace{1cm} (36)$$

where $D = 518/1125 + 11692/15975(1/p-1)$. The value for $p$ is structure dependent and describes basically the fraction of the magnetic enthalpy that is being absorbed above the critical (Néel) temperature. For most common phases the value for $p = 0.28$. The thermodynamic stacking fault energy calculations are summarized in the flow chart presented in Figure 6.
**Figure 6.** Flow chart of the thermodynamic stacking fault energy calculations.
3 MECHANICAL MATERIALS TESTING

The basic principle common to almost all mechanical materials testing techniques is that a load is applied to a material and simultaneously the material’s resistance to deformation is measured. The details in the testing techniques, however, can differ fundamentally depending on the objective of the test. Performing mechanical tests at varying strain rates requires special considerations due to changing testing conditions and becomes usually more challenging with increasing strain rates. The most common testing techniques with the corresponding strain rate ranges and prevailing test conditions are presented in Figure 7.

**Figure 7. Common mechanical materials testing techniques, their corresponding strain rate range, and the prevailing test conditions [15].**

At quasistatic strain rates, either servohydraulic or screw driven materials testing machines can be utilized. Although screw driven machines usually provide somewhat more stable crosshead displacement rates, nowadays servohydraulic universal testing machines are the most commonly used machines due to their superior versatility, which enables to test materials in tension, compression and bending in a cyclic manner or with a single continuous movement at a constant rate of strain, displacement, or load. With a conventional setup strain rates up to 1 s\(^{-1}\), depending on the material and the specimen dimensions, can be achieved, while tests up to about 500 s\(^{-1}\) are possible with specially designed servohydraulic machines. The crucial factor in these tests is the acceleration of the load-frame/piston since achieving a certain velocity requires time, and a load pulse is associated with the acceleration. So-called acceleration grips are often used, which grip the specimen only once the machine’s piston has already reached a high velocity. Severe oscillations in the measured force signal, however, are unavoidable mostly due to the elastic waves caused by the sudden engagement of the acceleration grips. Alternative measuring techniques in the intermediate strain rate range are, for example, the drop tower test and the cam plastometer. At even higher strain rates, roughly from 10\(^2\)-10\(^4\) s\(^{-1}\), the most commonly utilized measuring technique is the Hopkinson split bar technique. For strain rates in excess of that, explosively driven techniques or impact tests are usually used.
3.1 Inertia and wave propagation effects

A fundamental difference between high strain rate tests and tests performed at quasi-static strain rates is that inertia and wave propagation effects become increasingly important at high strain rates [15]. Therefore stress wave propagation along the length of the specimen, and eventually along parts of the testing machine, have to be taken into consideration during the analysis and validation of high strain rate test results.

The strain in a specimen deforming at high loading rates may not be uniform due to inertia opposing uniform deformation. This is illustrated for the case of a uni-axial tensile test in Figure 8. The specimen can be divided into incremental elements of length $dx_0$ and cross sectional area $A_0$. The initial position of the element is described by the position vector $\vec{x}$. No forces along the tensile axes are acting on the element in its initial condition. Upon loading of the specimen, the element will be displaced by a distance $u$ and deformed to the new dimensions $dx$ and $A$. The forces acting on the element are $\vec{F}_{-x}$ on the back-side and $\vec{F}_{+x} + d\vec{F}_x$ on the front-side. The displacement $\vec{u}$ of the element is due to the difference $d\vec{F}_x$ between the end-face forces causing the motion of the element, which is governed by Newton's second law, i.e., force $F$ equals mass $m$ times acceleration $\ddot{a}$:

$$d\vec{F}_x = m \cdot \ddot{a} = \rho_0 A_0 dx_0 \left( \frac{d^2 \vec{u}}{dt^2} \right)$$ (37)

where $\rho_0$ is the specific density of the material. From Equation 37 it is obvious that if tests are conducted at very low speeds where very small accelerations $\ddot{a}$ are involved, the variation in the force $d\vec{F}_x$ between the end-faces is negligibly small. The specimen can then be considered as being in force equilibrium.

However, at increasingly higher deformation speeds, the acceleration term in Equation 37 gains significance, and the variation in the force $d\vec{F}_x$ between the end-faces increases. Since the specimen consists of a large number of incremental elements chained up in a row along the x-axis, a variation of the axial force along the entire length of the specimen is the consequence. As the force becomes more non-uniform, so must the stress, and due to the interrelation between stress and strain also the strain and the strain rate will vary with the axial position in the specimen. The concept of stress-equilibrium, based on which mechanical tests at quasi-static strain rates are conducted, is no longer valid and the deformation of the specimen must be described by wave propagation theory. In practice, wave propagation phenomena start taking effect at strain rates around $10^2 \text{s}^{-1}$ [15].
Figure 8. The deformation of a tensile test specimen by a stress wave. The specimen contains incremental elements of length $dx_0$ and initial cross-sectional area $A_0$. The displacement of the element is $u$, the differential length of the element as a function of time is $dx$, and the forces acting on the faces of the element are $F$ and $F + dF$. [after 15]

A stress wave is transmitted and reflected within the test piece, where the stress wave propagates at the speed of sound. In the case of purely elastic deformation, the longitudinal sound velocity in an infinite slender bar is $c_{EL} = \sqrt{\rho/E}$, where $E$ is Young’s modulus and $\rho$ is the density of the material. If the material has deformed into the plastic region, the wave propagation is better described by the plastic-wave velocity. According to wave theory, the one-dimensional plastic wave propagates at a velocity

$$c_{PL} = \sqrt{\frac{d\sigma}{d\varepsilon} / \rho}$$

(38)

where $d\sigma/d\varepsilon$ is the slope of the true stress vs. true strain curve. For many materials, the initial work-hardening rate $d\sigma/d\varepsilon$ is approximately 1% of $E$, and thus the plastic wave velocity is approximately an order of magnitude smaller than the elastic wave velocity. Experimental data on a variety of metals, however, shows that plastic waves travel, at least initially, at velocities closer to the elastic wave velocity rather than the plastic wave velocity predicted by Equation 38. The estimate in Equation 38 is therefore obviously too conservative and the plastic wave propagation more complicated.

The propagation of stress in terms of waves with a certain wave velocity sets also some requirements to the specimen geometry used in high strain rate tests, which will be discussed in more detail in section 3.3.4. A sufficient number of wave transits has to
occur during a test so that the use of average stresses and strains is justified. Davies and Hunter [186] have estimated as a first approximation that three reverberations are needed to achieve stress equilibration in compression testing. The time until stress equilibrium is achieved is called the ringup time.

### 3.2 Adiabatic heating

Another effect that comes into play at high strain rates is the adiabatic deformation heating. At low strain rates, the test conditions can be considered essentially isothermal, i.e., there is sufficient time available for the heat generated by plastic deformation to be dissipated to the surroundings and the specimen temperature therefore remains constant. With increasing deformation rates, the test conditions first change to quasi-adiabatic and at very rapidly occurring deformations, to adiabatic. Under essentially adiabatic deformation conditions the generated heat has a strong effect on the constitutive material behavior due to thermal softening of the material. At intermediate strain rates, both heat generation and heat conduction, i.e., thermal diffusivity have to be considered.

Depending on the strain and strain rate, the temperature increase due to adiabatic heating can easily be as much as 50-100 °C [139-141]. Larour et al. [141] reported an increase in temperature of 95 °C during dynamic testing of TRIP steels, resulting in a decrease of the material’s strength by 80 MPa due to thermal softening. The temperature rise due to adiabatic heat generation can be expressed based on the specific heat capacity $C_p$

$$ C_p = \frac{\Delta Q}{m \Delta T} $$

(39)

as

$$ \Delta T = \frac{\Delta Q}{\rho C_p} = \frac{\beta}{\rho C_p \varepsilon_0} \int \sigma \varepsilon $$

(40)

where $\rho$ is the density and $m$ the mass of the material. The integral term is the total mechanical energy generated during deformation, where $\sigma$ and $\varepsilon$ are stress and strain, respectively, and $\beta$ describes the fraction of mechanical energy that is converted to heat energy $\Delta Q$. $\Delta Q$ can be determined from calorimetric measurements, from where $\beta$ can then be calculated easily. Manson et al. [129] obtained values around 0.8-0.9 for $\beta$ from high strain rate experiments on 2024 Al and 4340 steel. The experiments were conducted with the Hopkinson split bar technique coupled with infrared detectors. Experiments on Ti-6Al-4V revealed a drop in the $\beta$-values from almost 1 to 0.6 as the strain increased from 0.05 to 0.2. Meyers [100] suggested that the anomalous conversion fractions might be related to deformation mechanisms other than dislocation motion, i.e., twinning and martensitic phase transformations. For metals the value of $\beta$ is usually around 0.9.

Also the formation of shear bands, i.e., regions of highly concentrated plastic deformation can be attributed to adiabatic heating. If a local region in a material undergoes an incre-
ment of plastic strain \( \frac{d\varepsilon}{dt} \) adiabatically, strain hardening causes an increase in stress, while thermal softening opposes the strain hardening \[142\]. If the localized thermal softening exceeds the strain hardening, adiabatic shear bands can be formed.

### 3.3 Hopkinson split bar technique

The Hopkinson split bar technique is the most commonly used technique to study the mechanical behavior of materials at strain rates of about \(10^2 - 10^4\) s\(^{-1}\). The technique is based on the propagation of elastic waves in slender cylindrical bars.

#### 3.3.1 Elastic waves in cylindrical bars

For the case that the length of the bar is long compared to its diameter, the description of the wave propagation simplifies to a one dimensional, i.e., uni-axial problem. If strains and inertia in the direction transverse to the bar are excluded, the description becomes linear and can be expressed by a second-order partial differential equation

\[
\frac{d^2u}{dt^2} = C_0^2 \frac{d^2u}{dx^2} \tag{41}
\]

where \(u\) is the particle (atom) displacement along the x-axis (bar-axis), \(t\) the time, \(C_0\) the speed of sound, and \(x\) the x-axis coordinate \[100, 150\]. The solution of the equation describes the shape of stress pulses propagating along the x-axis of the bar with the velocity \(C_0\). The shape of the pulse remains unchanged with time.

Impacting a cylindrical bar with a cylindrical projectile of length \(L\), while both bars are made from the exact same material and have the same cross-sectional area \(A\) with flat and perfectly aligned ends, a rectangular pulse of length \(2L\) is generated, at least according to idealized theory (see Figure 9). The stress in the bar \(\sigma_b\) generated by the impact of the striker at velocity \(v_s\) is

\[
\sigma_b = \frac{1}{2} v_s \rho_s C_0 \tag{42}
\]

where \(\rho_s\) is the density of the striker bar material. However, in case that the bar materials are not identical and/or differ in their cross-sectional area, an acoustic impedance mismatch has to be taken into consideration, causing an alteration of Equation 42. The acoustic impedance of a bar is given by \(Z = A \cdot C_0 \cdot \rho\), and for the case of a long rod where transversal strain can be neglected \(C_0 = \sqrt{E/\rho}\), and hence

\[
Z = A \sqrt{\rho E} \tag{43}
\]

The stress caused by an incident stress wave \(\sigma_i\) transmitted from bar 1 into bar 2 follows the relationship
\[
\sigma_r = \frac{2(Z_2/Z_1)(A_i/A_2)}{1 + (Z_2/Z_1)} \sigma_i
\]

while the stress of a wave reflected back into bar 1 from the bar 1/bar 2 interface is described by

\[
\sigma_r = \frac{(Z_2/Z_1) - 1}{1 + (Z_2/Z_1)} \sigma_i
\]

Impedance changes are of importance for the evaluation of Hopkinson split bar measurement data, where both intrinsic and extrinsic impedance changes can occur, e.g., due to a material or bar diameter mismatch, or heating of sections of a bar, etc..

**Figure 9.** Cylindrical projectile impacting cylindrical bar and the corresponding idealized stress pulse. [after 100].

### 3.3.2 Compression and tensile Hopkinson split bar

The compression Hopkinson split pressure bar (HSB) is schematically illustrated in Figure 10b. The setup consists of two elastic bars, between which a specimen is sandwiched. An elastic compressive pulse in the incident bar is generated through the direct impact of a third bar, called the striker bar or the impactor, with the free end of the incident bar. The compressive pulse propagates towards the specimen, where part of the pulse continues through the specimen into the transmitted bar as a compressive wave while the rest of the pulse is reflected back into the incident bar as a wave of tension. Upon penetration by the pressure wave, the specimen is being deformed elasto-plastically. All three stress pulses are measured by strain gages, which are usually attached to the mid-span of the bars in pairs on opposite sides. The signals are amplified and recorded, for example, by a digital oscilloscope. In data post processing on a PC, stress, strain, and strain rate in the specimen can be calculated.
The tensile Hopkinson split bar (T-HSB) technique, in principle, is identical with the compressive HSB, the elastic waves only being of opposite sign, i.e., compressive waves are replaced by tensile waves and vice versa. However, the mechanical construction of the T-HSB device is somewhat more complicated due to the fact that a tensile pulse needs to be generated and the fixation of the specimen requires special engineering considerations. Additional challenges are associated with the tensile specimen geometry, which will be discussed in section 3.3.4. Different approaches to create a tensile pulse have been proposed in the literature [143-149], e.g., by direct impact of a tube-like striker onto a flange at the end of the incident bar, where the compressive pulse is immediately reflected as a wave of tension (Figure 10a). Other techniques involve controlled pre-straining of a part of the incident bar, which creates a tensile wave upon releasing a clamp. Lindholm and Yeakley [143] proposed a technique which is only a slight modification of the compressive bar. The transmitted bar in this technique is a hollow tube and a hat shaped specimen undergoes tensile deformation. Also different approaches for the fixation of the specimen between the incident and transmitted bar have been proposed, varying from mechanical clamping by threads, fixation by bolts or pins through a hole in both the bar and specimens ends, to gluing of the specimen into slits machined at the ends of the bars. The crucial aspect concerning the fixation method is that the stress waves should be left as undistorted as possible by the fixation and at the same time guarantee firm hold of the specimen.

Figure 10. Schematic presentation of the Hopkinson split pressure bar setup in tension.
Figure 11. a) Typical strain gage data after signal conditioning and amplification from a tensile split Hopkinson pressure bar test on a stainless 304 steel sample. The sum of the reflected signal and transmitted signals has to be equal to the incident signal. The transmitted bar signal was in this case amplified more than the incident bar signal. b) Sketch of an actual stress pulse (oscillations exaggerated) [after 100].

Figure 11 shows typical pulses acquired from a test on a steel sample. Knowing the cross-sectional area of the pressure bars $A_b$ and the Young’s modulus of the bar material $E$, the forces on the front face and the back face of the specimen can be calculated as

$$F_{\text{front}}(t) = A_b E \left[ \varepsilon_i(t) + \varepsilon_r(t) \right]$$  \hspace{1cm} (46)

and

$$F_{\text{back}}(t) = A_b E \varepsilon_r(t)$$  \hspace{1cm} (47)

where $\varepsilon_i(t)$, $\varepsilon_r(t)$, and $\varepsilon_t(t)$ are the incident, reflected, and transmitted wave signals measured from the strain gages after conversion into axial strains. The average stress in the specimen can then be calculated as

$$\sigma(t) = \frac{F_{\text{front}}(t) + F_{\text{back}}(t)}{2 A_s} = \frac{A_b E \left[ \varepsilon_i(t) + \varepsilon_r(t) + \varepsilon_t(t) \right]}{2 A_s}$$  \hspace{1cm} (48a)

where $A_s$ is the cross-sectional area of the specimen. Since the equation involves all three waves, it is usually referred to as three-wave equation. The strain rate at which the specimen is being deformed can be calculated from the three waves as

$$\dot{\varepsilon} = \frac{C_o \left[ \varepsilon_i(t) + \varepsilon_r(t) + \varepsilon_t(t) \right]}{L_s}$$  \hspace{1cm} (49a)
which, when integrated with respect to time over the whole duration of the deformation process gives the strain as

$$\varepsilon = \int_0^t \dot{\varepsilon}(t) dt = \frac{C_0}{L_S} \int_0^t (\varepsilon_i(t) + \varepsilon_s(t) + \varepsilon_r(t)) dt$$  \hspace{1cm} (50a)

where $C_0$ is the elastic wave speed in the bar material. If the specimen has achieved stress equilibrium, the forces on the front face and the back face of the specimen are equal, i.e., $F_{\text{front}}(t) = F_{\text{back}}(t)$. From Equations 46 and 47 then follows $\varepsilon_i(t) + \varepsilon_s(t) = \varepsilon(t)$. Substituting into Equations 48-50, one obtains the so-called one-wave equations, i.e., equations for stress, strain, and strain rate, which are only dependent on one elastic wave signal each:

$$\sigma(t) = E \frac{A_h}{A_s} \varepsilon_r(t)$$  \hspace{1cm} (48b)

$$\dot{\varepsilon}(t) = 2 \frac{C_0}{L_S} \varepsilon_r(t)$$  \hspace{1cm} (49b)

$$\varepsilon(t) = 2 \frac{C_0}{L_S} \int_0^t (\varepsilon_r) dt$$  \hspace{1cm} (50b)

From a practical point of view, Equation 48b normally results in smoother stress curves since the incident bar signals often contain a lot of high-frequency oscillations, which are largely filtered out by the deforming specimen.

### 3.3.3 Elastic wave dispersion

An exaggerated sketch of an actual stress pulse is shown in Figure 11b, which differs from the idealized stress pulse sketched in Figure 9. There are significant fluctuations in the pulse caused by wave dispersion effects as well as an increase in the rise time of the pulse. Wave dispersion results from the fact that the elastic wave propagation in reality is not entirely uni-axial. There are radial stress components due to the Poisson’s effect facing radial inertia, and wave interactions with the external bar surfaces. Already slightest misalignment of the bar faces will introduce additional transversal waves. Furthermore, the load pulse is composed of different frequency components traveling at different velocities. The load pulse thus undergoes phase shifts between the different frequency components and ‘smears out’ with time. Since in practice the pulses are recorded at some distance from the bar-specimen interface, dispersion effects have to be accounted for in order to obtain correct results. Numerical dispersion correction methods have been proposed for example by Bacon [151], Follansbee et al. [152], and Gorman et al. [153].

Increased pulse rise times can be a desired effect, because the deforming specimen will reach stress-state equilibrium at lower strains. If the rise time of the incident wave is of
the order of the ring-up time of the specimen, data will be valid also at lower strains. An increase in the rise time can be achieved by the use of a so-called pulse shaper, which is a little shim placed between the striker bar and the incident bar. The choice of the exact measures and material for the shim is usually an empirical matter of trial and error, although some numerical solutions have been proposed as well [171].

### 3.3.4 Effects of specimen geometry

High strain rate testing in both compression and tension sets certain requirements to the specimen geometry and dimensions. In order to obtain an essentially uni-axial and uniform stress state in compression test specimens, friction effects and effects of longitudinal and radial inertia have to be reduced to a minimum. According to Hunter and Davies [156], the optimum specimen length-to-diameter ratio $L/D$ is around 0.5 for steels. At ratios significantly lower, frictional effects at the specimen/bar interface become pronounced and at the same time radial inertia effects gain significance. At $L/D$ ratios much higher, the stress-equilibrium condition gets eventually violated.

In the tensile test the gage length of the specimens is typically somewhat longer than in compression tests, and the specimen cross-section changes between the gage and grip sections. Thus the deformation within the specimen may not be completely homogeneous and uni-axial. Specimen geometry and dimension dependence in tensile tests has indeed been reported from the very beginning of tensile Hopkinson Split Bar testing [5].

In many cases, the material of interest is in the form of a thin metal sheet, which sets some special requirements for the specimen preparation and mounting in the tensile HSB device. Many different specimen geometries and dimensions are used for T-HSB testing of sheet steels in laboratories all over the world. Basically all research groups use dog-bone shaped specimens, even though the geometrical details may vary. In addition to the fact that normally only samples with a rectangular cross section can be obtained from sheet products, there are also a few further requirements that determine the specimen geometry. First, the specimen gage section has to be short enough to achieve high strain rates. Secondly, the gage length should also be short enough for a given strain rate to assure homogeneous deformation of the specimen. In order to reduce the contribution of the deformation around the shoulder area to the calculated strain, the radius of the shoulder (or fillet) of the specimen is recommended to be smaller than that used in conventional quasi-static tests [157]. On the other hand, the grip section needs to be large enough in its cross section to ensure that no, or as little as possible, plastic deformation takes place there because it would also cause an error in the strain calculation. Verleysen and Degrieck [158] studied the effect of specimen geometry in a T-HSB test using a rotating drum camera, which allowed them to measure the axial strain distribution during the tensile test. Clear specimen geometry dependence in the test results was observed, and it was also found that the strain determined from the strain gage signals tended to overestimate the strain observed by the optical method. Rusinek et al. [159] studied the neck formation behavior by numerical analysis and found that strain rate and gage length correlate with the gage location where necking sets in, and thus whether stress equilibrium was achieved or not.
4 EXPERIMENTAL PROCEDURES

The mechanical properties of a DP 600 and a TRIP 700 steel were characterized by tensile tests over a temperature range of \(-100 \leq T \leq 235\) °C at strain rates ranging from \(10^{-3}\) to \(1250\) s\(^{-1}\). The low and intermediate strain rate tests \((10^{-3} \ldots 1\) s\(^{-1}\)) were carried out with an Instron 8800 servohydraulic materials testing machine using a commercial vertical tube furnace and a specially designed cooling system for the high and low temperature tests. To examine the effect of high strain rates \((750 \ldots 1250\) s\(^{-1}\)), the steels were dynamically tested with a tensile Hopkinson Split Bar (T-HSB) device. Heating of the specimens in the T-HSB tests was achieved with resistive heating plates, and cooling with a similar cooling system as used in the low strain rate tests, where cryogenic nitrogen gas flow is used to reach the temperatures below RT.

Five experimental high-alloyed manganese TWIP steels were produced for this study. Their stacking fault energy was adjusted by chemical alloying based on thermodynamic stacking fault energy calculations. The dependence of the mechanical behavior of TWIP steels on composition (and thereby SFE), temperature, and strain rate, was studied utilizing basically the same techniques as used for the studies of the DP and TRIP steels. In order to get better insight into the active deformation mechanisms during straining of the high manganese TWIP steels, electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM) were used. Special attention was paid to the effects of temperature, strain rate and grain size on twinning and the microstructure/texture evolution in general. A fine grained TWIP steel was produced specially for the studies of the grain size effect.

4.1 Materials

Both the TRIP and DP steels studied were commercial grades produced by Rautaruukki Oyj, while the TWIP steels were experimental grades prepared at the Department of Materials Science of TUT.

4.1.1 TRIP and DP steels

The basic chemical compositions of the TRIP and DP steels under investigation are presented in Table 1. After initial cold rolling, both steels were annealed in the austenite-ferrite region at 800 °C for 60 s, followed by rapid cooling to 450 °C below the bainite start temperature, and an isothermal hold at this temperature for 90 s before final cooling to room temperature. The nominal thickness of the DP steel was 1.2 mm and that of the TRIP steel 1.2 and 1.6 mm. X-ray diffraction measurements with Ni-filtered Cu K\(_\alpha\) radiation were used to determine the initial volume fraction of retained austenite in the TRIP steel microstructure. By applying the direct comparison method as described in [160], a value of 11.8 vol.% was obtained, which is consistent with the optical and scanning electron micrographs of the strips in the as-delivered condition.
Table 1. Chemical compositions of the studied TRIP 700 and DP 600 steels in weight percentages.

<table>
<thead>
<tr>
<th>Material</th>
<th>C [%]</th>
<th>Si + Cr + Mo [%]</th>
<th>Al + Mn [%]</th>
<th>Fe [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>TRIP 700</td>
<td>0.2</td>
<td>0.4</td>
<td>2.9</td>
<td>Bal.</td>
</tr>
<tr>
<td>DP 600</td>
<td>0.12</td>
<td>0.75</td>
<td>1.5</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

Table 2. Chemical compositions of the tested TWIP steels. The values were obtained using a Baird Dv-6 optical emission spectrometer.

<table>
<thead>
<tr>
<th>Material</th>
<th>Mn [%]</th>
<th>Al [%]</th>
<th>Si [%]</th>
<th>C [%]</th>
<th>Cr + Mo [%]</th>
<th>Nb [%]</th>
<th>Fe [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>TWIP 1</td>
<td>28</td>
<td>1.6</td>
<td>0.28</td>
<td>0.08</td>
<td>&lt;0.01</td>
<td>&lt;0.001</td>
<td>Bal.</td>
</tr>
<tr>
<td>TWIP 2</td>
<td>25</td>
<td>1.6</td>
<td>0.24</td>
<td>0.08</td>
<td>&lt;0.01</td>
<td>0.05</td>
<td>Bal.</td>
</tr>
<tr>
<td>TWIP 3</td>
<td>27</td>
<td>4.1</td>
<td>0.52</td>
<td>0.08</td>
<td>&lt;0.01</td>
<td>0.05</td>
<td>Bal.</td>
</tr>
<tr>
<td>TWIP 4</td>
<td>28.1</td>
<td>0.9</td>
<td>0.54</td>
<td>0.17</td>
<td>&lt;0.01</td>
<td>&lt;0.001</td>
<td>Bal.</td>
</tr>
<tr>
<td>TWIP 5</td>
<td>23.9</td>
<td>3.5</td>
<td>0.448</td>
<td>0.11</td>
<td>1.0</td>
<td>&lt;0.001</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

4.1.2 TWIP steels

Five TWIP steel grades were prepared by induction melting in a vacuum furnace and cast in 0.2 bar argon atmosphere to bars of 60 * 90 * 400 mm. The bars were reheated at 1200 °C and hot rolled to sheets of 2 mm in thickness. The finish rolling temperature was 900 °C and the ‘coiling’ temperature 600 °C. TWIP 4 and 5 were annealed at 1150 °C subsequent to hot rolling and finally quenched with nitrogen gas. The chemical compositions of the studied TWIP steels are presented in Table 2. An additional fine grained TWIP steel grade was produced after cold rolling to 30 % reduction.

4.2.1 Compression test specimens

The compression test specimens were cut from the sheets either by wire-cutting or by water jet cutting to discs of 8 mm in diameter, where the cutting direction was along the sheet normal direction. The disks were ground and lapped to produce parallel end face surfaces with good surface quality. If necessary, the required specimen length was acquired by stacking several discs to lengths between about 4-6 mm. The disc stacking technique has been analyzed and discussed in Ref. [172].

4.2.2 Tensile test specimens

The specimens were glued into the incident and transmitted bars with a cyanoacrylate adhesive Loctite ® 480, which had sufficient shear strength to hold the specimen during the test but could be removed relatively easily with acetone and mild heating after the test. For gluing, a 2 mm wide and 24 mm deep slit was machined at the specimen end of both pressure bars. A testing arrangement where the specimen was fixed to the transmitted bar with a bolt was also tried, but this method led to a very poor transmitted pulse quality and was therefore abandoned.
The choice of the specimen geometry for the tensile tests involved extensive experimental testing as well as elasto-plastic finite element simulations with the main objective of determining the effective gage length of specimens having different geometries and dimensions. A good agreement between the observations made directly from the specimens and the results of the FE simulations was obtained, and the most relevant conclusions are reported in this section. The detailed experimental procedures and the results of this study are presented in the attached Publication I.

For the preparation of the specimens, specimen blanks were first sawed out from the steel sheets, after which the gage section with desired dimensions and geometry was machined with a milling cutter. Different cutter discs were used to achieve different gage lengths and fillet radii. The milling cutter was driven along the axis perpendicular to the surface to be milled. Figure 12 illustrates the specimen geometry that was used after the specimen geometry studies. The gage length \( L \) of the specimens was 6 mm, while the width \( W \) and thickness \( T \) were about 4 mm and 2 mm, respectively. The radius \( R \) of the transition zone was either 1 or 2 mm. For better comparability of the test results, the same specimen geometry was also used in the low and intermediate strain rate tests. Further gage lengths and fillet radii had been tested in the framework of the specimen geometry studies.

![Figure 12. Specimen geometry used in the tensile test, \( L = 6 \) mm, \( W = 4 \) mm, and \( T = 2 \) mm. \( R \) was either 1 or 2 mm.](image)

In order to verify the observed deformation characteristics and to allow better judgment of certain specimen geometries, numerical simulations were performed using ABAQUS/Explicit. The specimens were modeled using a 3D calculation with 8-node solid elements with reduced integration and hourglass control. Element size of approximately 0.3 mm in element edge length was used in the region of interest, i.e., in the gage section, and about double sized edge length in the grip sections. A region of 2 mm in length behind the shoulders was also meshed with the finer mesh in order to get precise results for the deformation behind the shoulders. The meshing resulted in about 15,000-25,000 elements depending on the used specimen geometry. The following boundary conditions were used (Figure 12): the tensile specimen was fixed at the left end so that all degrees of freedom were constrained, and the right grip zone was constrained in \( y-z \) displacement. A dynamic load pulse of 5 kN, lasting 300 \( \mu \)s, was applied to the
right end of the specimen. This is equivalent to the conditions of shooting the 800 mm striker tube of the present T-HSB device with a speed of about 11 m/s against the flange of the incident bar. Strain rate and temperature effects, including adiabatic heating were not included in the simulation.

**Figure 13.** a) Comparison between true stress-strain curves of an AISI 304 stainless steel specimen with $L = 6$ and $R = 2$ mm obtained from a T-HSB experiment and the FE simulation. The strain rate was about 1600 s$^{-1}$. b) Line and nodal displacements obtained from the measurements and the simulation, respectively. The gage section and the transition zones are marked with double-headed arrows.

Figure 13a shows a comparison of the stress-strain curves obtained from experimental tests and FE simulations using the iteration procedure described in Publication I. Figure 13b shows the local displacement along the central x-axis of the fractured specimen for both a T-HSB test and an FE simulation.

Figure 14 presents the results of a quasi-static FE calculation showing the stress distribution in the specimen during a tensile test as it appears in the beginning of the test before yielding, i.e., in the elastic regime. The modeled stress state is based on the multi-axial von Mises stress. Stress concentrations in the corners or curvatures of the transition regions promote initial yielding of the specimen in these areas, from where the stress concentration then proceeds towards the center of the specimen. Especially for the lower ductility materials, fracture initiating at the corner regions could be expected. However, none of the tested specimens showed necking or final fracture in the transition region, and all specimens fractured in the mid-length in both low rate servo-hydraulic tests and in high rate HSB tests. This is obviously because strain hardening of the materials compensates for the effect of stress concentration after the beginning of yielding.
For ductile materials the specimen geometry has to be chosen carefully, since the strain distribution along the gage section is strongly geometry dependent. Considerable deformation can take place even behind the shoulders, as can be clearly seen in Figure 15, where the transition regions of AISI 304 specimens with fillet radii of 0, 1 and 2 mm are shown. Comparison of the curvatures of deformed grid lines shows that a sharp corner or small fillet radius leads to significant deformation behind the shoulder. The contribution of deformation from this region to the total deformation, however, decreases with increasing fillet radius and is only very limited for a fillet radius of 2 mm. Thus, for ductile and strongly strain hardening materials, a larger transition radius should be chosen. The extra amount of strain in the transition region, however, has still to be taken into consideration. If the extra amount of elongation is known, this can be done, e.g., by calculating the stress-strain curve with a ‘virtually’ extended gage length. However, also in that case the exact stress-strain correlation cannot be evaluated since the contribution of the deformation in the transition region to the total deformation depends on the stage and progress of deformation.
Figure 15. Transition regions after high strain rate tensile testing of AISI 304 stainless steel specimens with fillet radius of a) 0 mm, b) 1 mm, and c) 2 mm.

Figure 16. A dual phase steel DP800 tensile test specimen before and after tensile testing. The distance between the grid lines was measured before and after the tests to determine the amount of local strain in the specimen.

For materials with lower ductility the aforementioned considerations are of less significance. For example in DP800 specimens, no significant deformation took place in the grip area behind the shoulder as the straight grid lines in the deformed specimen in Figure 18 indicate. The deformation in the transition zone caused an additional strain of less than 2 % for transition radii of both \( R = 1 \text{ mm} \) and \( R = 2 \text{ mm} \) geometries, which can be almost neglected in the calculations. Indeed, the stress-strain curves obtained from these two specimen geometries were almost identical. For most materials included in the present study, the fillet radius of 1 mm was found to be optimal.

Stress-equilibrium during the tests was confirmed by both FE simulations and experiments, where in general the force on the front-face equaled the force on the back-face of the specimen. The specimens deformed almost homogeneously during the entire duration of the stress pulses, and specimens failed in the middle of the gage section. Time resolved strain measurements from three strain gages attached to different parts of a
specimen’s gage section confirmed the findings for the initial phase of deformation, as can be seen from Figure 17a. All the observations described above are also in good agreement with the observations made by Gilat [188], who studied the deformation behavior in T-HSB experiments with the digital image correlation (DIC) technique.

With increasing length of the gage section the deformation seems to become non-homogeneous to some extent. From the graphs in Figure 17b, it can be seen that the local strains are slightly higher on the incident bar side of the sample and the stress concentration in the gage zone shifts to the incident bar end with increasing gage length and strain rate (Figure 18a). This is caused by the propagation of the stress wave through the specimen, causing straining first on that side of the specimen which the wave reaches first. This observation is in agreement with the observations of Huh et al. [161], who studied the dependence of the neck location on the initial gage length. The longer the gage length was, the more the neck moved towards the incident bar side, indicating that the deformation was not perfectly homogeneous along the specimen. This effect, however, also depends on the material’s strain hardening behavior. At lower striker impact speeds, the phenomenon of double-necking appears. This has been experimentally observed on an AISI 304 specimen with 8 mm gage length and strain rate of about 400-500 s\(^{-1}\). FEM simulations confirmed the findings for a 6 mm gage length and strain rate of about 250-300 s\(^{-1}\), as shown in Figure 18b. Wood [163] attributed this phenomenon to the transition from strain rates where inertia dominates the deformation behavior to wave propagation controlled deformation. The observations are also in agreement with those made by Rusinek et al. [159].

![Figure 17. Time resolved local deformation from strain gages attached to the surface of a specimen’s gage section in the middle as well as close to each transition zone. L was 8 mm and R = 2 mm. The maximum strain obtainable by the strain gages was 5 %. The gluing of the strain gage closer to the incident side failed after about 1.5 % strain (a). An example of the local strain distribution in a DP800 sample with L = 4 mm and R = 1 mm after fracture (b).](image-url)
Figure 18. Localized necking in a specimen with high ductility (a), and double necking phenomenon at lower strain rates.

4.3 Hopkinson split bar equipment

While the majority of tests in this study were performed in tension, additional HSB compression tests were performed on the TWIP steels.

4.3.1 Compression Hopkinson Split bar

The compression HSB setup used in the present study consists of high strength steel (AISI 4030) bars, each 22 mm in diameter and 1200 mm or 1800 mm in length. The striker bars are made from the same material and have the same diameter. Their length varies between 200 and 800 mm. A momentum trap bar was used in addition to the incident and transmitted bars in order to trap the residual transmitted pulse and thereby prevent it from being reflected back to the transmitted bar. Alignment of the bars is achieved via four bearing stanchions, which are adjustable in height (z-axis) and in the horizontal direction (y-axes) perpendicular to the length axis of the bars. The striker bars are accelerated by pressurized air, where the striker velocity can be controlled by adjusting the launch pressure in a gas cylinder. More detailed descriptions of the test system can be found, for example, in Refs [164, 165].

Dispersion of the pressure waves during propagation between the strain gages and the bar/specimen interfaces was accounted for by the use of the dispersion correction procedure originally proposed by Gorman [154], which transforms the measured signal from the time domain to the frequency domain by a fast Fourier transform (FFT), performs the appropriate phase shifts to each frequency component according to the Pochhammer-Chree solution, and transforms the signal back to the time domain using an inverse FFT.

4.3.2 Tensile Hopkinson Split bar

In the T-HSB setup at Tampere University of Technology (DMS /TUT), the specimen is mechanically fixed by an adhesive between two slender elastic bars and the elastic tensile pulse is generated by direct impact of a tube-like striker, accelerated to velocities typically between 1.5 - 20 m/s, with a flange machined to the end of the incident bar. The uniaxial tensile pulse propagates in the incident bar towards the bar/specimen interface, where part of the pulse continues through the specimen into the transmitted bar as a wave.
of tension, while the remaining part of the pulse is reflected back into the incident bar and propagates in the opposite direction as a wave of compression. The three pulses (incident, reflected, and transmitted) are measured by pairs of 5mm strain gages attached on the bars. The signals are recorded by a digital oscilloscope to be processed on a PC after the test. Both the striker and the incident bar are made of AISI 4340 steel and the transmitted bar of an AA 2007 aluminum alloy, which increases the transmitted bar’s signal to noise ratio due to its lower Young’s modulus. The length of the incident and transmitted bars are 6000 mm and 3000 mm, respectively, while striker tubes 800 and 1670 mm in length were used. The bar-bearing stanchions are basically identical to those used in the compressive system.

For some of the tests performed on the TWIP steels, another slightly different tensile Hopkinson Split Bar setup at the Ghent University (UGent) was used. The basic working principle is identical to the TUT setup described above, although some slight differences in the actual designs exist. At Ghent University, the striker is accelerated by a fork located above the incident bar and accelerated by compressed air. Both bars at Ghent University are made from an Al 5083 aluminum alloy, while the striker is made from Ertalon technical Nylon. Despite the differences in the setups, experiments on both setups led to consistent results.

The use of different bar materials for the incident and transmitted bars, as used at DMS /TUT, requires some modifications in the Equations 48-50 for the calculation of stress, strain, and strain rate. The three wave equations are then written as

\[
\dot{\varepsilon} = \frac{v_{incl} - v_{trans}}{L_s} = \frac{C_{incl}(\varepsilon_{incl} - \varepsilon_{ref}) - C_{trans}\varepsilon_{trans}}{L_s} \quad (51a)
\]

\[
\varepsilon = \frac{u_{incl} - u_{trans}}{L_s} = \frac{C_{incl} \int (\varepsilon_{incl} - \varepsilon_{ref}) - C_{trans} \int (\varepsilon_{trans})}{L_s} \quad (52a)
\]

\[
\sigma = \frac{E_{incl} A_{incl} (\varepsilon_{incl} + \varepsilon_{ref}) + E_{trans} A_{trans} (\varepsilon_{trans})}{2A_s} \quad (53a)
\]

where \(u\) and \(v\) are the incident and transmitted bar end displacements and velocities, respectively. All other variables have the same meaning as described in section 3.3.2. Under the assumption of stress equilibrium the use of one-wave equations is justified. For the strain and strain rate calculations, the equation is actually a two-wave equation since it includes both incident and reflected wave signals, such that

\[
\dot{\varepsilon} = \frac{C_{incl}(\varepsilon_{incl} - \varepsilon_{ref}) - C_{trans} \frac{E_{incl} A_{incl}}{E_{trans} A_{trans}} (\varepsilon_{incl} + \varepsilon_{ref})}{L_s} \quad (51b)
\]
4.3.3 Specimen heating/cooling techniques in mechanical testing

By means of various high/low temperature setups, the mechanical properties of the studied materials were determined in a temperature range from -100 up to 235 °C. For the high and low temperature tests with the servohydraulic materials testing machine, a commercial vertical tube furnace and a specially designed cooling system using cryogenic nitrogen gas flow to reach temperatures below RT were utilized.

High and low temperature tests with the HSB technique set some additional requirements for the technical setup. On the one hand, the bar material is usually restricted to be used in a rather narrow temperature range, either directly due to limits in the strength of the material or due to severe changes in the acoustic impedance of the bar material, invalidating the use of the common equations for the stress and strain calculations. In compression testing the problem can be overcome by heating/cooling of the specimen in a separate furnace and bringing it then rapidly into contact with the bars for testing. The critical factor in this technique is the transition time between the furnace and the bars, which creates some engineering challenges to the system design. The high/low temperature setup of the compression HSB at DMS/TUT is based on specimen and bar manipulator systems that retract the specimen out of the furnace after achieving the required temperature and closes the bars to minimize the contact time of the specimen with the pressure bars to less than 50 ms. A detailed description of the HT/LT system is presented in Refs [164-167].

In tensile HSB testing, heating or cooling of the specimen has to take place while the specimen is already fixed to the bars, and most techniques are therefore limited to rather narrow temperature ranges unless the supply of thermal energy happens very fast, i.e., by high energy influx to the specimen. Techniques based on this principle were used for example by McDougall [168] and Clausen et al. [169]. High rate cooling, however, can not be realized in a comparable way since heat dissipation by both thermal conduction and radiation take significantly more time. In the present T-HSB system the fact that the specimens are glued into the bars by the aid of an organic adhesive with limited thermal persistence sets further restrictions.

A sketch of the experimental setup used for low temperature testing at TUT is illustrated in Figure 19, where the same principle applies for both low and high strain rate testing. Nitrogen gas is flowed through a heat exchanger, which is immersed in liquid nitrogen, into a cooling chamber. The temperature in the chamber can be adjusted by controlling
the gas flow with the aid of a PID (proportional–integral–derivative) controlled proportional valve and a thermocouple.

![Figure 19. Schematic of the cooling device used in T-HSB testing at TUT.](image)

A schematic of the heating device used for elevated temperature T-HSB testing at DMS/TUT is illustrated in Figure 20. Two preheated heating plates consisting of cylindrical heating cartridges embedded in rectangular copper blocks of 10*10*35 mm are brought into contact with the specimen surfaces in an area 2mm wider on both sides than the actual gage section, which results in heating of the gage section. Finite element calculations performed with Abaqus/Standard as well as experiments with thermocouples embedded into drilled holes showed that no essential temperature change occurs after about 10-15 seconds contact time and no considerable temperature gradient exists along the gage section. Outside the heating section, however, the temperature gradient is very steep, and the grip sections remain at temperatures close to RT even after 30 seconds of
heating. Once the desired test temperature is reached, the T-HSB test is conducted as usual.

For a series of experiments a hot water heating system designed at Ghent University was used to control the specimen temperature in T-HSB tests. This setup is illustrated in Figure 20b. The water temperature in the water reservoir is controlled with a thermostat, and before and during the experiment the specimen is heated by the flow of hot water. Prior FEM simulations verified that 30 seconds is a sufficient time for the specimen gage section to reach thermal equilibrium.

4.4 Microstructure and texture characterization by EBSD and TEM

Combined microstructure and texture studies as well as identification of twins and twin orientations were performed by means of the EBSD technique with a Channel 5 EBSD system (HKL technology) operated on a Zeiss ULTRAplus field emission gun scanning electron microscope (FEGSEM). Cross-sectional cuts of the tensile specimens were prepared parallel to the plane defined by the rolling direction and the sheet plane normal. In order to collect statistically representative data, the EBSD analyses were first carried out using a beam step size between 0.3-0.8 μm in scans over large areas containing a sufficient number of grains. For the acquisition of more detailed EBSP quality images and orientation maps, the beam step size was reduced at minimum to 30 nm, which allowed, e.g., detailed local microstructure and texture analyses. For the characterization of the defect structures of the deformed specimens, transmission electron microscopy was used utilizing a Jeol JEM 2010 analytical transmission electron microscope.
5 RESULTS AND DISCUSSION

In the following sections, the results and major observations of the studies on the TRIP, DP, and TWIP steels are presented and discussed. Major parts of the results have been previously published in the attached publications.

5.1 TRIP and DP steels

The mechanical properties and deformation behavior of two advanced high strength steels, DP 600 and TRIP 700, were studied mainly with tensile tests at strain rates ranging from $10^{-3}$ to $1250 \text{ s}^{-1}$ in the temperature range of $-100 \leq T \leq 235 ^\circ \text{C}$. Both steels are commonly used in various structural parts and components of cars, which makes their stress-strain response at high strain rates particularly interesting and important.

5.1.1 Mechanical behavior

Figure 21 shows the true stress vs. logarithmic (true) strain curves and the corresponding strain hardening rates for the steels under investigation at room temperature at two different strain rates, $10^{-3}$ and $750 \text{ s}^{-1}$. The true stress $\sigma_{\text{true}}$ was determined from the engineering stress $s$ and the engineering strain $e$ as

$$\sigma_{\text{true}} = \frac{F}{A_0} (e + 1) = s(e + 1)$$  \hspace{1cm} (54)

where $F$ is the applied force and $A_0$ the initial cross sectional area, while the logarithmic strain was determined from the engineering strain $e$ as

$$\varepsilon = \ln(e + 1)$$  \hspace{1cm} (55)

**Figure 21.** Room temperature true stress vs. logarithmic (true) strain and strain hardening rate $\dot{\theta} = d\sigma/d\varepsilon$ of the studied TRIP and DP steels at strain rates $\dot{\varepsilon} = 10^{-3} \text{ s}^{-1}$ (a) and $750 \text{ s}^{-1}$ (b).
Figure 22. Yield and ultimate tensile strengths of the studied DP (a) and TRIP (b) steels as a function of strain rate (values refer to the left y-axis). Slope \( m = d\sigma / d\ln \dot{\varepsilon} \) is the strain rate sensitivity parameter. The effect of strain rate on the energy absorption up to 10% elongation at room temperature (values refer to the right y-axis).

From the figure it is obvious that the strain hardening behaviors of TRIP and DP steels have some clear differences. The DP steel shows a significantly higher strain hardening rate \( \theta = d\sigma / d\varepsilon \) at low strain levels, but at a plastic strain of \( \varepsilon \approx 7\% \), the strain hardening curves intersect, and after this the TRIP steel strain hardens at a much higher rate than the DP steel. This behavior can be directly attributed to the gradually proceeding strain-induced austenite/martensite transformation in the TRIP steel, which is also responsible for the much higher total elongations of the TRIP steel compared to the DP steel.

The slopes \( m = d\sigma / d\ln \dot{\varepsilon} \) of the linear regression lines fitted to the yield strength and UTS data points in Figure 22 were used as a measure of the strain rate sensitivity of the studied steels. Both steels show positive and quite linear strain rate sensitivity over the measured strain rate range. The TRIP steel is clearly less sensitive to strain rate with respect to both tensile strength and yield strength. For both steels a decrease in the strain rate sensitivity is observed at higher strain levels, which can be explained by the thermal softening caused by deformation induced adiabatic heating, partially compensating the direct flow-stress increasing effects of strain rate at high strain levels.

Because of the higher overall strain rate sensitivity of the DP steel, its yield strength and tensile strength exceed the values of the TRIP steel at higher strain rates, although at low strain rates the case is clearly the opposite. The reason why the yield strength of the DP steel shows somewhat higher strain rate sensitivity than that of the TRIP steel is not quite clear, but it could be related to the higher initial dislocation density of the DP steel, which basically is the only microstructural difference between these two steels from the yielding point of view. On the other hand, the tensile strength of the TRIP steel seems to be less sensitive to strain rate as well, which in turn might be explained by the effects of
deformation induced heating on the martensite transformation based hardening of the TRIP steel.

The difference in the strain rate sensitivities of DP and TRIP steels affects, for example, the relative impact strength of the parts and structures made of these two steels. In many applications, however, the energy absorption capability of the structure is even more important than its strength alone, and hence a better measure of the suitability of a particular material for a certain application is often its energy absorption potential. Therefore, the energy absorption capability of the studied steels was evaluated by integrating the area under the determined tensile stress-strain curves. It has become a common practice to use the value of energy absorbed by 10 % of tensile (or compressive) deformation as a classification value [74, 77], because in many cases, such as in automotive accidents, the parts of the vehicle are not designed to deform to their final fracture, or even close to it. It should be noted that in the present case the materials were tested in the cold rolled and subsequent thermally treated condition without the typical forming and bake hardening operations that automotive steels are subjected to in practice.

The energies absorbed by the studied TRIP and DP steels are also illustrated in Figure 22 for increasing strain rates at room temperature. The 10 % energy absorption values of the DP steel at low strain rates are slightly higher than those of the TRIP steel, but the difference is not yet too significant. At high strain rates, however, the strain rate sensitivity of the DP steel results in a much larger difference between the energy absorption values of these two steels, and on this basis the DP steel would appear to be a better choice for certain types of dynamic applications. The situation, however, changes markedly if instead of the 10 % energy absorption values also the total deformation capability of the studied steels is taken into account in the comparison. The main benefit of the TRIP steel is that it can reach very large uniform and total elongations and that its work hardening capability persists up to higher strains, resulting also in higher total energy absorption values, as shown in Figure 23. Especially in the cases where manufacturing of the structural parts by forming or stamping already involve high degrees of deformation, the excellent formability of TRIP steels would be an advantage because it leaves more energy absorption capacity to the finished part.

Figure 23. Absorbed energy as a function of strain at the strain rate of $\dot{\varepsilon} = 10^3$ s$^{-1}$ at room temperature.
**Dynamic mechanical properties of DP 600 at varying temperatures**

Fig. 24 presents the true stress-logarithmic strain curves of DP 600 at various strain rates of $\dot{\varepsilon} = 10^3, 10^1, 750, \text{ and } 1250 \text{ s}^{-1}$ and temperatures, ranging from -50 to 115 °C in the low strain rate tests and from -100 to 235 °C in the high strain rate tests. As the rather similar shape of the stress-strain curves suggests, the testing temperature does not have a dramatic effect on the overall strain hardening behavior of the DP steel, although all strength values decrease with increasing temperature, as Figures 24 and 25 clearly show. The yield strength values presented in Figure 25 were determined as the 0.2 % offset yield stress $R_{p0.2}$ from the *engineering* stress-strain curves. It should, however, be noted that the $R_{p0.2}$ values obtained from the dynamic tests have to be considered with caution, since in the very early stages of deformation the specimen has not necessarily reached stress equilibrium yet. In addition to the $R_{p0.2}$ values, also backward extrapolated yield strength values [171], which are often used in dynamic tests, were determined and compared with the $R_{p0.2}$ values. The values obtained by this method were generally slightly higher but did not differ in their temperature and strain rate dependence from the $R_{p0.2}$ values and are therefore not presented here. Also the ultimate tensile strength (UTS) and elongation values $e_{un}$ (uniform elongation) and $e_f$ (fracture elongation) were determined from *engineering* stress-strain curves due to the fact that in tension the *true* stress vs. *logarithmic* strain curves are only valid up the point of uniform elongation.\(^1\)

Fig 25a shows that at low strain rates the yield strength decreases only slightly when the temperature increases from -50 to 115 °C, but at high strain rates the drop is very distinct in the temperature range $-100 \leq T \leq 235$ °C. The UTS values, however, decrease with increasing temperature at a more or less constant rate at all strain rates over the studied temperature regions. The elongation values $e_{un}$ and $e_f$ shown in Fig 25b do not essentially depend on temperature in the low strain rate region, but at high strain rates a substantial decrease in the elongation values is observed at low temperatures. In general, the overall deformation behavior shown by the DP steel is quite typical for a material deforming by ordinary dislocation glide and where strain hardening is caused primarily by dislocation interactions. Figure 25c shows the average strain hardening rates $\theta = d\sigma/d\varepsilon$ between 1 and 15 % of plastic strain as a function of temperature. In general somewhat lower values are observed at high strain rates, what can be attributed mostly to the effects of adiabatic heating.

\(^1\) The YS and UTS values as well as the $e_{un}$ and $e_f$ values at different deformation temperatures for the strain rates $10^3 \text{ s}^{-1}$ and $1250 \text{ s}^{-1}$ are summarized in the Appendix in Tables 7 and 8, respectively.
Figure 24. True stress vs. logarithmic strain of the DP 600 steel at different temperatures and strain rates of $10^{-3}$ s$^{-1}$ (a), $10^{-1}$ s$^{-1}$ (b), 750 s$^{-1}$ (c), and 1250 s$^{-1}$ (d).
Figure 25. Yield and ultimate tensile strength of the DP 600 steel as a function of the deformation temperature (a). Uniform elongation $e_{un}$ and total elongation $e_f$ of the DP 600 steel as a function of the deformation temperature (b). The $e_{un}$ and $e_f$ values are extracted from the engineering stress-strain curves. Strain hardening rate $\theta = \frac{d\sigma}{d\varepsilon}$ of the DP 600 steel as a function of temperature at different strain rates (c).

Dynamic mechanical properties of TRIP 700 at varying temperatures

Because of the differences in the governing deformation and strain hardening mechanisms in TRIP and DP steels, also the strain rate and temperature dependence of the mechanical behavior of these steels can be expected to be, at least to some extent, different. The true stress-logarithmic strain curves of the studied TRIP steel are presented in Figure 26, showing that, in contrast to the DP steel, the strain hardening behavior of the TRIP steel depends strongly on the temperature, especially at low temperatures. With increasing temperature, the slope of the stress-strain curve in the plastic region decreases, which is a direct consequence of the temperature dependence of the martensitic phase transformation. XRD measurements on compressed TRIP steel specimens deformed to different levels of plastic strain confirmed these findings, as can be seen in Figure 27; at
lower deformation temperatures, the initial austenite phase fraction of 11.8 vol% decreases more rapidly with strain due to a higher phase transformation rate $dV^{\gamma\rightarrow\alpha'} / d\varepsilon$.

At low temperatures the strain hardening rates $\theta = d\sigma/d\varepsilon$ are in general somewhat higher for the material being deformed at low strain rates, as can be seen in Figure 28a, which compares the strain hardening rates at high and low strain rates at different temperatures and in Figure 28b, where the average strain hardening rates $\theta = d\sigma/d\varepsilon$ between 1 and 15% of plastic strain are plotted as a function of temperature. Since the difference in the strain hardening behavior at low temperatures occurs already at low stains, it cannot be explained by the effects of adiabatic deformation heating. However, since the yield strength of the material shows clearly positive strain rate sensitivity, possibly a higher fraction of austenite has already deformed under stress-assistance in the elastic regime and thus consumed part of the strain hardening potential brought about by the martensite transformation. However, no microstructural or XRD studies confirming this have yet been performed.

The difference in the average strain hardening rate disappears at temperatures around RT. At room temperature a reduced phase transformation rate $dV^{\gamma\rightarrow\alpha'} / d\varepsilon$, i.e., higher retained austenite fraction after deformation was observed at high strain rates already at low deformation levels, where no significant adiabatic heating effects due to plastic deformation should occur. The contribution of martensite transformation to the overall strain hardening can thus be expected to be lower at high strain rates also irrespective of the possible changes in the internal temperature of the material due to adiabatic heating. One possible explanation for this could be that twinning in the austenite phase starts to replace the martensitic phase transformation, showing a similar effect on the strain hardening behavior as the martensite transformation. The general conditions for twinning, which is assisted by large stress concentrations rather than by thermal activation [58], basically exist. However, no microstructural studies confirming this either have yet been performed. On the other hand, the latent heat released by the martensitic phase transformation itself is a cause for local adiabatic heating, which has to be taken into consideration. In any case, the exact reasons for the observed behavior are very complex and no unambiguous conclusions can be drawn at this stage. Also a difference in the strain rate sensitivity of the martensitic phase transformation may exist depending on whether the material is being deformed in tension or in compression, and therefore the correlations between the measured mechanical behavior and the X-ray diffraction measurements should be treated with due caution, as well as the accuracy of the X-ray diffraction measurements.
Figure 26. True stress vs. logarithmic strain curves of the TRIP steel at different temperatures at the strain rates of $10^3 \text{s}^{-1}$ (a) and $10^{-3} \text{s}^{-1}$ (b), with magnifications of the yield point, showing the transition from smooth yielding to discontinuous yielding, and strain rates of $750 \text{s}^{-1}$ (c), and $1250 \text{s}^{-1}$ (d).

Figure 27. Austenite volume fractions as a function of plastic strain determined by XRD from specimens deformed in compression at $\dot{\varepsilon} = 10^3 \text{s}^{-1}$ and temperatures of $-50^\circ\text{C}$, $22^\circ\text{C}$, and $100^\circ\text{C}$, as well as at $\dot{\varepsilon} = 10^{-3} \text{s}^{-1}$ and RT.
Figure 28. Strain hardening rate $\theta = d\sigma/d\varepsilon$ of the TRIP 700 at different temperatures at strain rates $\dot{\varepsilon} = 10^{-3}$ s$^{-1}$ and 1250 s$^{-1}$ (a) and the average strain hardening rate between 1 and 15% of plastic strain (d).

Figure 29. Yield strength and ultimate tensile strength of the TRIP steel as a function of the deformation temperature in the temperature range $-100 < T < 235$ °C (a), and uniform elongation $\varepsilon_{un}$ and total elongation $\varepsilon_f$ of the TRIP steel as a function of the deformation temperature (b). The $\varepsilon_{un}$ and $\varepsilon_f$ values are extracted from the engineering stress-strain curves.

Figure 29a shows the YS and UTS values of the TRIP steel as a function of temperature and strain rate. The curves reveal clear differences in the mechanical behavior of the TRIP and DP steels, in particular at temperatures slightly below the room temperature. Figure 29b shows the dependence of the uniform and total elongations of the TRIP steel.
on strain rate and temperature. From the presented data it is evident that both elongations have maximum values somewhere between 75 and 150 °C, independent of the strain rate, which would therefore also be the optimal ‘cold’ forming temperature for this TRIP steel. To determine the exact temperatures for maximum elongations, more data points in this interval would be needed. In any case, the maximum elongations correspond to the situation where strain-induced α′-martensite transformation persists up to very high strains because of the optimal temperature window with respect to the thermodynamics of the martensite transformation [173-174]. As discussed in chapter 2.2.2, the temperature at which the martensitic phase transformation ceases to occur is the $M_p$-temperature, above which hardening is caused mainly by dislocation interactions and under high strain rate deformation by possible contributions from mechanical twinning. The hardening rate, however, is not any more sufficient to prevent localized necking from proceeding at higher strains, and as a consequence, the elongation values of the studied TRIP steel begin to drop at $T > 115$ °C. This decline continues until $M_s$, above which the elongations either level off or slightly increase again. On the other hand, when moving from the temperature of maximum elongation towards lower temperatures, the thermal stability of the austenite phase and thus the additional strain needed for the α′-martensite transformation to occur decreases, while at the same time the slip stress increases and therefore the stress levels needed for the martensite transformation to initiate are reached more easily, resulting in more intensive work hardening at low strains and decreasing elongation values. Also the stacking fault energy decreases with decreasing temperature and thus promotes the martensite transformation. In the temperature range $-100 < T < -10$ °C, the metastable austenite starts to transform already under elastic stress, i.e., without yielding, which results in a further decrease of the elongations with decreasing temperature.

The stress-assisted and strain-induced regimes of martensite transformation are separated by the $M_S^t$ temperature, which for the present material appears to be approximately -10 °C, corresponding to a local minimum in the strength values [73, 175-176]. Below this temperature, the yield strength of the TRIP steel increases due to the martensitic phase transformation occurring already under elastic stress, whereas above $M_S^t$ austenite first undergoes some plastic deformation, increasing the flow stress and probably providing additional nucleation sites for the martensite, before the transformation can start. $M_S^t$ is also marked by the transition from smooth yielding to discontinuous yielding [73, 175] which, although not too distinct, could be observed also for the TRIP steel in the low strain rate tests (see Figure 26), confirming the $M_S^t$ temperature to be approximately -10 °C. However, the exact values of $M_S^t$ (as well as for $M_S$ and $M_D$) may vary between individual grains due to variations in the local composition, grain size [177-178], grain orientation [179], and local stress distribution. This might also be the reason for the observed serrations in the curves of stress and elongation over temperature especially around the $M_S^t$ temperature. The conclusions concerning the $M_S^t$ temperature were

---

1 The YS and UTS values as well as the $e_\text{un}$ and $e_\text{f}$ values at different deformation temperatures for the strain rates $10^3$ s$^{-1}$ and 1250 s$^{-1}$ are summarized in the Appendix in Tables 7 and 8, respectively.
drawn from the observations concerning the yielding behaviour of the steel, the formation of stress-assisted martensite was, however, not experimentally verified.

5.2 TWIP steels

In order to better understand the dependence of the mechanical behavior of TWIP steels on composition, grain size, stacking fault energy, temperature, and strain rate, thermodynamic stacking fault energy calculations were performed for the five high-alloyed manganese steel grades. Three of the TWIP steels were then mechanically tested at various strain rates and temperatures and their microstructure and texture evolution was studied by means of EBSD. The defect structures of deformed samples were also studied by transmission electron microscopy. The results of the mechanical tests and microstructural characterizations are discussed and compared with recent observations made by other research groups. The results of the compression tests performed on TWIP 4 and 5 were published, for example, in Ref. [183].

5.2.1 Thermodynamic Stacking Fault Energy Calculations

The stacking fault energy $\gamma_{SFE}$ of the Fe-Mn-Al-Si-C-(Cr) system was calculated for the compositions presented in Table 2 over the temperature range $50 \text{ K} \leq T \leq 600 \text{ K}$ based on Equation 26 and the equations presented in section 2.4.1. The thermodynamic datasets needed for the calculations were acquired from the literature, mainly from the tables published by the Scientific Group Thermodata Europe (SGTE).

The molar surface density \( \rho \) was calculated according to Equation 27, where the lattice parameter \( a \) was determined from XRD measurements on the TWIP steel grades using Ni-filtered Cu-K\( _\alpha \) radiation. The variation in the determined lattice parameter between different TWIP grades was of the order of \( \Delta a = \pm 0.01 \text{ Å} \), which is considered small enough to be neglected. The same average value \( a = 3.60 \text{ Å} \) was therefore used for all grades.

The change in the molar Gibbs energy \( \Delta G^{\gamma\rightarrow\varepsilon} \) upon the \( \gamma_{\text{fcc}} \rightarrow \varepsilon_{\text{hcp}} \) phase transformation was calculated according to Equation 30:

\[
\Delta G^{\gamma\rightarrow\varepsilon} = \chi_{Fe} \Delta G_{Fe}^{\gamma\rightarrow\varepsilon} + \chi_{Mn} \Delta G_{Mn}^{\gamma\rightarrow\varepsilon} + \chi_{Al} \Delta G_{Al}^{\gamma\rightarrow\varepsilon} + \chi_{Si} \Delta G_{Si}^{\gamma\rightarrow\varepsilon} + \chi_{C} \Delta G_{FeMnXCr}^{\gamma\rightarrow\varepsilon} \\
+ \chi_{Fe} \chi_{Mn} \Omega_{FeMn}^{\gamma\rightarrow\varepsilon} + \chi_{Fe} \chi_{Al} \Omega_{FeAl}^{\gamma\rightarrow\varepsilon} + \chi_{Fe} \chi_{Si} \Omega_{FeSi}^{\gamma\rightarrow\varepsilon} + \Delta G_{mg}^{\gamma\rightarrow\varepsilon} \tag{56}
\]

The chemical contributions \( \chi_{\varphi} \Delta G_{\varphi}^{\gamma\rightarrow\varepsilon} \) of the pure elements \( \varphi \) to the change in the Gibbs energy were calculated according to Equation 31. The power series representing the molar Gibbs energies \( G_{\varphi}^{\gamma} \) and \( G_{\varphi}^{\varepsilon} \) as a function of temperature \( T \) were taken from the literature [33-36]. The effect of carbon was included according to the empirical formula proposed by Dumay et al. [34]. In the case of TWIP 5, an additional term for the contribution of Cr was included in Equation 56. The resulting functions describing \( \Delta G_{\varphi}^{\gamma\rightarrow\varepsilon} \) are presented in Table 3. Although recent studies [34] concerning the effect of
silicon on the stacking fault energy $\gamma_{SFE}$ in high Mn-steels have indicated that $\gamma_{SFE}$ seems to first increase for small Si quantities and then decrease at higher ones, the conservative purely thermodynamic formula for $\Delta G_{Si}^{\gamma \rightarrow \epsilon}$ was used.

Only the excess free energies $\Omega^{\gamma \rightarrow \epsilon}$ caused by Fe were considered, whereas the excess energy contributions of other element interactions as well as all ternary interactions and higher were neglected due to their minor contributions to $\Delta G^{\gamma \rightarrow \epsilon}$. The functions describing the excess energy coefficients $\Omega_{Fe\phi}^{\gamma \rightarrow \epsilon}$ were also acquired from the literature as indicated in Table 3.

The magnetic contribution $\Delta G_{mg}^{\gamma \rightarrow \epsilon}$ to the change in the Gibbs energy $\Delta G^{\gamma \rightarrow \epsilon}$ was calculated by

$$
\Delta G_{mg}^{\gamma \rightarrow \epsilon} = G_{mg}^{\epsilon} - G_{mg}^{\gamma}
$$

where $G_{mg}^{\epsilon}$ and $G_{mg}^{\gamma}$ were calculated according to Equations 34 and 35, thus taking into consideration the paramagnetic-to-antiferromagnetic state change of each phase $\epsilon$ and $\gamma$.

**Table 3**: Functions describing the change in the Gibbs energy $\Delta G_{\phi}^{\gamma \rightarrow \epsilon}$ upon the $\gamma_{Fe} \rightarrow \epsilon_{hcp}$ phase transformation for the pure elements Fe, Mn, Al, Si, and Cr, the excess energy coefficients $\Omega_{Fe\phi}^{\gamma \rightarrow \epsilon}$, as well as an empiric formula describing the effect of C on $\Delta G^{\gamma \rightarrow \epsilon}$.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Function in units of J/mol</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\Delta G_{Fe}^{\gamma \rightarrow \epsilon}$</td>
<td>$-2243.38 + 4.309 T$</td>
<td>[33]</td>
</tr>
<tr>
<td>$\Delta G_{Mn}^{\gamma \rightarrow \epsilon}$</td>
<td>$-1000.00 + 1.123 T$</td>
<td>[33]</td>
</tr>
<tr>
<td>$\Delta G_{Al}^{\gamma \rightarrow \epsilon}$</td>
<td>$2800 + 5 T$</td>
<td>[34]</td>
</tr>
<tr>
<td>$\Delta G_{Si}^{\gamma \rightarrow \epsilon}$</td>
<td>$-560 - 8 T$</td>
<td>[34]</td>
</tr>
<tr>
<td>$\Delta G_{Cr}^{\gamma \rightarrow \epsilon}$</td>
<td>$1370 - 10 T$</td>
<td>[34]</td>
</tr>
<tr>
<td>$\Delta G_{FeMnxCr}^{\gamma \rightarrow \epsilon}$</td>
<td>$\frac{1246}{\chi_C} (1-\exp(-24.29\chi_C) - 17.175\chi_{Mn})$</td>
<td>[34]</td>
</tr>
<tr>
<td>$\Omega_{FeMn}^{\gamma \rightarrow \epsilon}$</td>
<td>$2873 - 717 \left(\chi_{Fe} - \chi_{Mn}\right)$</td>
<td>[35]</td>
</tr>
<tr>
<td>$\Omega_{FeAl}^{\gamma \rightarrow \epsilon}$</td>
<td>$3328$</td>
<td>[36]</td>
</tr>
<tr>
<td>$\Omega_{FeSi}^{\gamma \rightarrow \epsilon}$</td>
<td>$2850 + 3520 \left(\chi_{Fe} - \chi_{Si}\right)$</td>
<td>[34]</td>
</tr>
</tbody>
</table>
The values of the magnetic moment $\beta^o / \mu_B$ are of empirical nature. Cotes et al. [29] proposed the following equation for the calculation of the magnetic moment of the austenite phase

$$\beta^\gamma = \chi_{Fe} \beta_{Fe} + \chi_{Mn} \beta_{Mn} - \chi_{Fe} \chi_{Mn} \beta_{FeMn}$$

(58)

where $\beta_{Fe}$ and $\beta_{Mn}$ are empirical values for the pure elements and $\beta_{FeMn}$ is a second order term, which is basically used as a fitting parameter to account for the deviation from the generic binary mixture behavior that uses weighted sums of the pure elements’ magnetic moments. The influence of Al and Si on the magnetic moment were not taken into consideration. However, Dumay et al. [34] proposed to take into account the influence of carbon by adding an additional term to Equation 58:

$$\beta^\gamma = \chi_{Fe} \beta_{Fe} + \chi_{Mn} \beta_{Mn} - \chi_{Fe} \chi_{Mn} \beta_{FeMn} - \chi_{C} \beta_{C}$$

(59)

According to Huang [37], the values for $\beta_{Fe}$ and $\beta_{Mn}$ are 0.7 and 0.62, respectively, while the values $\beta_{FeMn} = 0.64$ and $\beta_{C} = 4$ were taken from Dumay et al. [34]. The magnetic moment of the $\varepsilon$-phase was calculated according to Huang’s [37] paper and the additional term proposed by Dumay [34] as

$$\beta^\varepsilon = \chi_{Mn} \beta_{Mn} - \chi_{C} \beta_{C}$$

(60)

The magnetic transition temperature $T_{\text{Néel}}^\phi$ was calculated according to the empirical expressions in [34] and [37] as:

$$T_{\text{Néel}}^\gamma = 250 \ln(\chi_{Mn}) - 4750 \chi_{Mn} \chi_{C} - 6.2 \chi_{Al} - 13 \chi_{Si} + 720$$

(61)

and

$$T_{\text{Néel}}^\varepsilon = 580 \chi_{Mn}$$

(62)

The formulas for $\beta^o / \mu_B$ and the Néel temperature $T_{\text{Néel}}^\phi$ are summarized in Table 4. The energy per surface unit of the $\{1 1 1\}$ interface between $\varepsilon$ and $\gamma$ phases was assumed to be $\sigma = 8 \text{ mJ/m}^2$, according to Allain [35], who studied a comparable chemical composition.
In Table 5, the room temperature SFE values are presented for the TWIP steel compositions given in Table 2, while in Figure 30a the SFE of the alloys are plotted as a function of the temperature. The relative magnitudes of the SFE of the five test materials are $\gamma_{\text{TWIP} \, 3} > \gamma_{\text{TWIP} \, 5} > \gamma_{\text{TWIP} \, 4} > \gamma_{\text{TWIP} \, 1} > \gamma_{\text{TWIP} \, 2}$, where the magnitudes of $\gamma_{\text{TWIP} \, 4}$ and $\gamma_{\text{TWIP} \, 1}$ differ only slightly, as their almost identical chemical composition already suggests. Figure 32 shows the individual chemical contributions of each element and the first order excess term contributions to the stacking fault energy $\gamma_{\text{SFE}}$ of TWIP 1 as a function of temperature. In Figure 30b the SFE of TWIP 1 is shown as a function of temperature together with the contributions of the individual magnetic, chemical, and interfacial terms. It can be clearly seen that the magnetic contribution to the SFE is very strong at low temperatures, whereas at temperatures above the Néel temperature of the $\gamma$-phase, which was determined as $T_N^\gamma = 394$ K, the magnetic contribution levels off and becomes almost negligible so that the SFE is dominated by the chemical contribution at high temperatures. At $T_N^\epsilon = 161$ K, another change in the slope due to the change in magnetic ordering of the $\epsilon$-phase can be observed. The overall temperature sensitivity of the SFE is rather moderate at low temperatures, where the magnetic and chemical contributions more or less compensate for each other, but becomes rather high at temperatures above approximately 400 K. This can be seen also from Figure 31, where the first derivative $\frac{\partial \gamma_{\text{SFE}}}{\partial T}$ of $\gamma_{\text{SFE}}(T)$ is plotted over the temperature. The two critical magnetic transition temperatures $T_N^\epsilon$ and $T_N^\gamma$ can be observed in Figure 31 as discontinuities and as changes in the curvature of the $\frac{\partial \gamma_{\text{SFE}}}{\partial T}$ plot.

![Figure 30](image_url)

**Figure 30.** The stacking fault energies $\gamma_{\text{SFE}}$ of the TWIP steel alloys 1-5 as a function of temperature $\theta$ (a), and the stacking fault energy $\gamma_{\text{SFE}}$ of TWIP 1 as a function of temperature $\theta$, broken down into the components of chemical, magnetic and interfacial contribution (b).
Table 5. The stacking fault energies $\gamma_{SFE}$ of the TWIP steel alloys 1-5 at room temperature.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>TWIP 1</th>
<th>TWIP 2</th>
<th>TWIP 3</th>
<th>TWIP 4</th>
<th>TWIP 5</th>
</tr>
</thead>
<tbody>
<tr>
<td>SFE [mJ/m²]</td>
<td>27</td>
<td>20.5</td>
<td>42</td>
<td>27</td>
<td>33.5</td>
</tr>
</tbody>
</table>

Figure 31. First derivative $\frac{\partial \gamma_{SFE}}{\partial T}$ of $\gamma_{SFE}(T)$ of TWIP 1 as a function of temperature $T$.

Figure 32. The chemical contribution of each element and the first order excess term contributions to the stacking fault energy $\gamma_{SFE}$ of TWIP 1 as a function of temperature $T$, as well as the interfacial energy term.
5.2.2 Mechanical behavior of the high manganese TWIP steels

The microstructure of all TWIP steels was completely austenitic at room temperature, even after largest degrees of deformation, which was confirmed by X-ray diffraction measurements. TWIP 1-3 were studied with tensile tests at strain rates ranging from $10^{-3}$ s$^{-1}$ to 1250 s$^{-1}$, and temperatures varying between -50 °C ≤ T ≤ 80 °C. At room temperature, the differences in the quasi static strength levels of TWIP 2 and TWIP 3, whose compositions differ with respect to both aluminum and silicon, are relatively small, as revealed by the true stress vs. logarithmic strain curves shown in Figure 33a. However, the strain hardening rates $\dot{\gamma} = d\gamma/d\varepsilon$, which are shown in Figure 33b, reveal some differences in the hardening behavior of these two materials. TWIP 3 hardens clearly stronger up to about 24 % of plastic strain, but at higher elongations TWIP 2 exhibits higher strain hardening rate than TWIP 3, indicating that there are some differences in the strain induced hardening mechanisms between these two steels. As can be seen from Table 5 and Figure 30, the SFE of TWIP 3 is clearly higher than that of TWIP 2 (42 as opposed to 20.5 mJ/m$^2$). According to the SFE ranges reported by Grässel et al. [71], Frommeyer et al.[72], and Allain et al. [18] for the different deformation mechanisms, the value $\gamma_{SFE} = 42$ mJ/m$^2$ might already indicate a change in the governing deformation mechanism from twinning to dislocation slip. The flow stress levels of TWIP 1 are clearly lower than those of TWIP 2 and TWIP 3, but the hardening behavior of TWIP 1 is rather similar to that of TWIP 2. Basically the main difference between TWIP 1 and TWIP 2 is that TWIP 2 contains also niobium, which is supposed to have an effect on the grain size of the material. All three grades show total elongations slightly below 50 % (~70 % in engineering strain) and flow stresses increasing from about 350 MPa to 850 MPa (350-550 MPa in engineering stress). The average strain hardening exponent $n$ of the Hollomon equation is about 0.35 for all three studied steel grades, although the stress-strain response of the TWIP steels does not seem to exactly obey the Hollomon equation, which generally describes the stress-strain behavior with a more concave curvature.

![Figure 33](image-url)

**Figure 33.** True stress vs. logarithmic (true) strain curves of the TWIP 1-3 steel grades at strain rate $10^{-3}$ s$^{-1}$, $T = 22$ °C (a), and the corresponding strain hardening rates $\dot{\gamma} = d\gamma/d\varepsilon$ (b).
Further differences as well as systematic correlations in the mechanical behavior of the studied materials were observed when the testing conditions, i.e., temperature and strain rate, were changed. The yield strength (YS) and ultimate tensile strength (UTS) values of the three materials are plotted against the logarithmic strain rate in Figure 35, and the corresponding true stress vs. logarithmic strain curves are shown in Figure 34. The yield strength was determined from the engineering stress-strain curves as the intersection point of the lines fitted to the elastic and plastic regions (between about 2 and 10 % of plastic strain) of the curve, while the UTS value was determined as the stress maximum in the engineering stress-strain curve. For quantitative comparability of the strain rate sensitivities, linear regression lines were fitted to the data points in the low and intermediate strain rate range between $10^{-3} \text{ s}^{-1} \leq \dot{\varepsilon} \leq 750 \text{ s}^{-1}$, with the slope $m = d\sigma / d\ln\dot{\varepsilon}$ as a measure of the strain rate sensitivity. All three TWIP steels show quite linear and positive but rather weak strain rate sensitivity in this regime, with $m$-values ranging between 3.4 and 9.3. No significant difference between the strain rate sensitivity factors for YS and UTS was observed. Quite often a decrease in the strain rate sensitivity is observed at higher strain levels, as could be seen also for the DP and TRIP steels discussed in section 5.1, which is due to the thermal softening caused by the adiabatic heating of the material. This, however, could not be seen in the present results. As discussed in section 2.3.5, strain rate sensitivity is often defined also as $m = d\ln\sigma / d\ln\dot{\varepsilon}$. Table 6 shows also these $m$-values for the three TWIP steels for better comparability with published data. At a strain rate of about $1000 \text{ s}^{-1}$, a steep upturn in the flow stress with increasing deformation rate is usually attributed to viscous drag effects on the moving dislocations, as discussed in section 2.3.3.

As Figure 35d shows, at all strain rates TWIP 1 has clearly the lowest yield strength and UTS, while TWIP 2 and TWIP 3 show only relatively small differences in their strength values. The YS values of TRIP 3 are generally somewhat higher than those of TRIP 2, but for the UTS values the case appears to be the opposite, which can be attributed to the slightly different strain hardening behavior of these two steel grades due to their different compositions and stacking fault energies.
Figure 34. True stress vs. logarithmic strain curves of TWIP 1 at different temperatures at the strain rates of $10^{-3}$ s$^{-1}$ (a), and 1250 s$^{-1}$ (b), TWIP 2 at $10^{-3}$ s$^{-1}$ (c), and 1250 s$^{-1}$ (d), and TWIP 3 at $10^{-3}$ s$^{-1}$ (e), and 1250 s$^{-1}$ (f).
**Figure 35.** YS and UTS of the TWIP steels as a function of strain rate for TWIP 1 (a), TWIP 2 (b), and TWIP 3 (c). Slope $m = d\sigma / d\ln \dot{\varepsilon}$ is the strain rate sensitivity factor. YS and UTS of all materials are plotted in (d).

**Table 6.** Strain rate sensitivity factors $m$ of the three TWIP steels.

<table>
<thead>
<tr>
<th>Material</th>
<th>YS $m$-value</th>
<th>UTS $m$-value</th>
<th>YS $m$-value</th>
<th>UTS $m$-value</th>
</tr>
</thead>
<tbody>
<tr>
<td>TWIP 1</td>
<td>6.8</td>
<td>6.6</td>
<td>0.0085</td>
<td>0.0031</td>
</tr>
<tr>
<td>TWIP 2</td>
<td>6.0</td>
<td>9.3</td>
<td>0.0066</td>
<td>0.0072</td>
</tr>
<tr>
<td>TWIP 3</td>
<td>3.4</td>
<td>7.9</td>
<td>0.0087</td>
<td>0.0027</td>
</tr>
</tbody>
</table>
Figure 36. YS and UTS of the TWIP steels as a function of the deformation temperature at strain rates (a) $10^{-3} \text{s}^{-1}$ and (b) $1250 \text{s}^{-1}$, as well as the corresponding uniform elongations $\varepsilon_{un}$ and total elongations $\varepsilon_f$ at strain rates (c) $10^{-3} \text{s}^{-1}$ and (d) $1250 \text{s}^{-1}$.

The yield strength and UTS of the three TWIP steels increase only moderately with decreasing temperature, as seen in Figures 36a and 36b, where the YS and UTS values are plotted as a function of the deformation temperature at strain rates $10^{-3} \text{s}^{-1}$ and $1250 \text{s}^{-1}$. This behavior is characteristic for pure fcc metals and is in agreement with the observations on TWIP steels reported by other research groups [71].

[1] The YS and UTS values at different deformation temperatures for the strain rates $10^{-3} \text{s}^{-1}$ and $1250 \text{s}^{-1}$ are summarized in the Appendix in Table 7.
Just as at room temperature, TWIP 1 shows the lowest strength values over the whole covered temperature range at all studied strain rates. At low strain rates, the temperature dependence of the yield strength of TWIP 2 and TWIP 3 is almost identical, while the ultimate tensile strength of TWIP 2 increases clearly faster with decreasing temperature and reveals the strongest temperature sensitivity of all three materials. At high strain rates, such a clear correlation between strength and temperature for TRIP 2 and TRIP 3 is not so evident, as Figure 36b shows.

In Figures 36c and 36d, the uniform elongation $e_{un}$ and the total elongation $e_f$ of the studied materials are plotted as a function of the (initial) test temperature. The strain rates in Figures 36c and 36d were $10^{-3}$ s$^{-1}$ and 1250 s$^{-1}$, respectively. In general, both the uniform and total elongation first increase with decreasing temperature, reach a maximum value, and then decrease again with even more decreasing temperature. This typical behaviour of fully austenitic TWIP steels is usually explained as follows: with decreasing temperature, the amount of deformation twins formed during plastic deformation increases, because the twinning stress increases less rapidly with decreasing temperature than the slip stress. Since the critical twinning stress $\tau_c$ and the stacking fault energy $\gamma_{sf}$ are coupled by a proportionality constant (the exact expression of which depends on the material; see chapter 2.1.4), one can equivalently say that the stacking fault energy $\gamma_{sf}$ is lowered into a region, where twinning as a deformation mechanism becomes energetically favoured over dislocation motion. Stress induced deformation twins are generated at high strain levels preferentially in the regions of localized deformation, i.e., in the necking area, where also the stresses are highest facilitating the nucleation of twins [182]. Since the twin boundaries on {111} twin planes act as strong dislocation barriers [71], local hardening of the material in these regions sets in, the deformation shifts to a weaker region, and as a consequence, necking of the material is being retarded. At the temperature where the elongations show maximum values, the twin formation proceeds most gradually during the deformation of the material. At even lower temperatures, most of the twinning occurs in the earlier stages of deformation and the TWIP effect is lost, leading to decreasing elongation values with decreasing temperature [71].

From Figure 36c it is obvious that at low strain rates all three studied TWIP steels reveal maximum elongations somewhere between -30 °C and 0 °C. For the exact determination of the maximum values for each chemical composition, more data points in that temperature interval would be needed. At high strain rates, TWIP 1 shows maximum elongations at -25 °C ≤ T ≤ RT, TWIP 2 at -50 °C ≤ T ≤ 0 °C, and TWIP 3 at -25 °C ≤ T ≤ 0 °C. TWIP 3 shows clearly the lowest uniform elongations at all strain rates and temperatures, and by tendency also the lowest total elongation values. TWIP 1 and TWIP 2 show roughly the same levels of $e_{un}$ and $e_f$ at low strain rates for -25 °C ≤ T ≤ RT, but at lower temperatures the ductility of TWIP 1 decreases drastically while that of TWIP 2 decreases only moderately. The total elongation values of TWIP 3 are more or less comparable to those of TWIP 1. The less dramatic decrease in the elongation values of TWIP 2 in comparison to those of TWIP 1 and TWIP 3 at -50 °C may be explained by its lower SFE. TWIP 1 has a stacking fault energy of $\gamma_{SFE} = 20.5$ at RT, which decreases to

1 Table 8 in the Appendix summarizes these values.
\( \gamma_{SFE} = 17.9 \text{ mJ/m}^2 \) at -50 °C. Allain et al. [18] proposed mechanical martensite transformation to set in at stacking fault energies below 18 mJ/m². At lower temperatures mechanical martensite transformation therefore might partly compensate for the reduced mechanical twinning rate at high levels of strain. Pineau [162] confirmed these assumptions, proposing that the formation of \( \varepsilon \)-martensite might set in at low \( \gamma_{SFE} \) in a similar manner as observed in austenitic stainless steels of comparable SFE. However, no XRD of TEM measurements were conducted on the specimens deformed at low temperatures confirming this.

At high strain rates, two separate effects start to influence the mechanical behaviour of TWIP steels. On one hand, adiabatic heating of the specimen starts to play an important role at high strains. On the other hand, extensive twin formation has been observed by several researchers under high strain rate conditions, exceeding the amount of twins found in the microstructure after low strain rate deformation [71, 183]. Some of the observed results can be explained with these two effects. Under high strain rate deformation at 80 °C, TWIP 1 exhibits the highest uniform and total elongation values of 48 % and 64 %, respectively. The elongation values of TWIP 1 remain the highest of the three materials also with decreasing temperatures down to -25 °C. At temperatures lower than that, however, the same phenomenon as in low strain rate tests is observed, i.e., the elongation values of TWIP 2 decrease more moderately towards lower temperatures than those of TWIP 1. Unlike at low strain rates, the elongation values of TWIP 3 do not drastically decrease at \( T < -25 °C \).

At -50 °C, the total elongations of the TWIP steels deformed at high strain rates exceed the total elongations of the same materials deformed at low strain rates. This can also be explained by the higher amount of twins being formed under high strain rate deformation and with the adiabatic heating, which increases the actual specimen temperature towards the maximum elongation temperature of the material at higher strains. With increasing initial test temperatures, however, the case becomes the opposite, and the elongations observed under low strain rate deformation start to be larger than those observed at high strain rates. The drastic decrease in the elongation values at high temperatures and at the same time at high strain rates can again be attributed to the effects of adiabatic heating. The increase of temperature to 80 °C causes an increase in the SFE of about 4-5 mJ/m² for all studied TWIP grades. However, calculating the adiabatic temperature rise according to Equation 40, where \( \beta = 0.9 \), \( C_p = 0.46 \text{ kJ/kg K} \), \( \rho = 7.8 \text{ g/cm}^3 \), and using \( E_{mech} = 38 \text{ J/mm}^3 \) as the total mechanical energy generated under the actual test conditions, temperature increase of 95 °C is obtained at the end of the test¹. Since the slope \( \frac{\partial \gamma_{SFE}}{\partial T} \) of the \( \gamma_{SFE} \) vs. \( T \) curve is already quite steep in this temperature region (see Figures 30 and 31), the increase in the stacking fault energy caused by the total temperature increase is quite substantial, and can be as much as 25 mJ/m². Also the steep slopes in the elongation vs. temperature curves at high strain rates for temperatures ca. \( T > 20 °C \) can in general be explained by the change in the slope of the \( \gamma_{SFE} \) vs. \( T \) curve.

¹ Under comparable deformation conditions in a separate test, a similar value of 90°C was obtained from thermocouple measurements spot welded to the surface of an AISI 304 steel. Since the heat conduction from the specimen requires some time, the temperature increase is measured with a little time delay.
Based on the thermodynamic stacking fault energy calculations presented in section 5.2.1, the relative magnitudes of the stacking fault energy for the three test materials are $\gamma_{\text{TWIP 3}} > \gamma_{\text{TWIP 1}} > \gamma_{\text{TWIP 2}}$. TWIP 2 has the lowest stacking fault energy because its Al and Mn contents are the lowest, and TWIP 3 has the highest $\gamma_{\text{SFE}}$ because of its much higher Al concentration than the other two materials. The effect of Nb on $\gamma_{\text{SFE}}$ was not taken into consideration, although it indirectly also affects $\gamma_{\text{FCC}}$ by binding C in carbides. Assuming that niobium carbides, i.e., NbC and Nb$_2$C are formed, then based on the present quantities of the two elements basically all C could be ‘tied’ in NbC precipitates. According to the equation for the effect of C on the stacking fault energy presented in Table 5, this would result in a decrease of $\gamma_{\text{SFE}}$ by about 6 mJ/m$^2$. This scenario, however, is rather unlikely, and actually no large precipitates were observed in the hot rolled steels by SEM/XPS studies, although the effect of Nb on $\gamma_{\text{SFE}}$ cannot be ruled out per se. On the other hand, Nb also strengthens the material via precipitation hardening. This is also one possible reason why the strength levels of TWIP 2 and TWIP 3 are generally higher than those of TWIP 1, which does not contain niobium at all.

Of the three TWIP steels, TWIP 2, which had the lowest SFE, revealed the most stable mechanical performance over the entire temperature and strain rate range and showed the most beneficial strength and ductility combination. TWIP 1 revealed the lowest strength values under any condition, while TWIP 3 in general showed the lowest elongation values. TWIP 2 showed also superior UTS/YS ratios when compared to TWIP 3.

The combined effects of temperature and stacking fault energy on the twinning and other hardening mechanisms of the studied alloys seem to be, however, very complex, and therefore definite conclusions from the observed behaviours are very difficult to draw at this stage.

5.2.3 Microstructure and texture evolution

In order to get better insight into the active deformation mechanisms of the high manganese TWIP steels during straining, electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM) studies were performed. Special attention was paid to the effects of temperature, strain rate, and grain size on twinning and the microstructure/texture evolution in general. The EBSD measurements were conducted on tensile test specimens of TWIP 1, while the TEM studies were performed on compressive specimens. In order to study the microstructure at engineering plastic strains of $e = 0.11$, 0.22, and 0.35 in addition to the fractured (engineering strain of $e = 0.64$) and as-received materials a series of interrupted tests was performed at RT at a strain rate of $10^{-3}$ s$^{-1}$. The compressive tests were conducted on disc-shaped specimens as described in section 4.2.1 in the direction normal to the sheet plane at strain rates $10^{-3}$ and 600 s$^{-1}$.

At low deformations the thickness of deformation twins is too low to be resolved by the EBSD technique. However, after 11 % of tensile elongation, thin dark lines are present in some of the grains as the EBSP quality maps in Figure 37b reveal. Barbier et al. [89] attributed the low band contrast of these lines to the overlapping of the twin and matrix patterns. By deconvolution of the two patterns, the lines can be identified as twins. This
procedure was adopted in the present study and it confirmed the findings of Barbier et al. [89]. With increasing strain, the nano-twins form stacks that can be resolved and indexed as twin domains by the EBSD technique. However, the actual thickness of twins is easily being overestimated due to the limitations in the spatial resolution of the EBSD technique. In fact, matrix laths remain between the stacked nano-twins, as TEM studies proved.

Only few deformation twins were identified by the EBSD technique at this stage of deformation, as can be seen in Figure 37a, where twin boundaries are colored in yellow. Only one twin system was active in any single grain at this stage. Twinning occurred preferentially in grains with a tensile orientation close to $\{\bar{1}1\bar{1}\}$. These observations are in good agreement with the results reported by Yang et al. [184]. During further straining, the twinning activity increases and secondary conjugate twin systems become active, as can be seen in Figure 37c. Twins on conjugate slip planes form only after the tensile axis has rotated near to, or beyond, the $<001>$ -- $<\bar{1}1\bar{1}>$ symmetry line, where the primary and conjugate systems are equally stressed, as predicted also by theoretical considerations [58]. Figure 37d shows the twinned microstructure at fracture, where overlapping of twins with slip bands can be observed easily from the strong curvature of the twins, especially in heavily deformed sub-grain sections.

Figure 37. Orientation map (IPF-x coloring scheme) (a) and EBSP quality map of the TWIP 1 steel after engineering strains of 0.11 (b), 0.35 (c) and 0.64 (d).
Figure 38 shows a semitransparent orientation map overlaying an EBSP quality map of the TWIP steel after the material had been deformed to fracture at a strain rate of $10^{-3}$ s$^{-1}$ at $T= -50 \, ^{\circ}C$. The fracture strain was $e_F = 0.3$ in terms of engineering strain. The number of twins in some of the grains, especially in those oriented close to the [111] direction with respect to the loading direction, exceeded that of the sample deformed to 35 % of plastic strain at room temperature. Increased twin activity already at low strain levels is usually believed to be the reason for premature fracture of TWIP steels at low temperatures [71]. However, the total number of twins in the specimen deformed at -50 °C did not reach the same level as in the sample deformed to fracture at RT. This can obviously be explained by the fact that the texture has not yet evolved to the point where enough grains are favorably oriented for twinning. For example the orientation close to [001] is unfavorable for twinning, which is also apparent from Figure 38a, and the deformation in these grains is dominated by dislocation glide. The observation is in agreement with the results of Barbier et al. [89], who showed that at RT the rotation of grains into orientations that are favorable for twinning continues up to fracture. The effect of premature fracture at low temperatures can therefore be explained as a combined effect of increased twin activity in favorably oriented grains and at the same time lack of grains oriented favorably for twinning.

Figure 38. Semitransparent orientation map (IPF-x coloring scheme) overlaying an EBSP quality map of the TWIP 1 steel after deformation to engineering strain of $e = 0.3$ at a strain rate of $10^{-3}$ s$^{-1}$ and $T= -50 \, ^{\circ}C$ (a).
Deformation at a strain rate of 1250 s\(^{-1}\) up to fracture resulted in the microstructure presented by the EBSP quality map in Figure 39. A high density of deformation twins is evident in addition to dark areas with low band contrast, representing high local dislocation densities.

The microstructure of the fine-grained TWIP steel shown in Figure 40 shows basically no evidence of twinning at all, only few thermal twins are visible. Similar grain size dependencies have been reported in the literature for other FCC metals. The effect of grain size on the twinning stress obeys in most cases the Hall-Petch relationship described by Equation 5 with a slope \(k_T\) larger than the slope \(k_S\) for slip, i.e., the twinning stress is more sensitive to grain size than the slip stress [62, 64]. However, the exact reasons for this behavior are still under discussion. Armstrong and Worthington [64] attributed the higher grain size sensitivity of the twinning stress to microplasticity, i.e., to dislocation activity occurring before the onset of globally plastic deformation, whereas the yield stress is associated with the onset of global plastic deformation.
5.2.4 Defect structures

The microstructure of the hot-rolled TWIP 1 steel consisted of tangled dislocations with fairly high densities already at this point. Minimal twinning was observed in a few grains. After straining to an engineering strain of $e = 6.5\%$ at a strain rate of $10^{-3}\text{ s}^{-1}$, an increase in dislocation density was evident, but only few individual twins were observed, most grains not exhibiting mechanical twinning at all. After $12\%$ of plastic strain at the same strain rate the dislocation density increased further, starting to show arrangement in cell structures or networks. However, there was still no evidence of marked increase in the number of deformation twins and only few individual twins were found in some of the grains. The situation was clearly different in the specimens deformed at strain rates around $600\text{ s}^{-1}$; some grains revealed a large number of twins already after plastic strains of about $5.6\%$. However, a large number of grains still did not show any evidence of twinning. The width of the twins was about 10-40 nanometers, as can be seen in Figure 41a, which shows a part of a highly twinned grain in a specimen deformed to $5.6\%$ of plastic strain at a strain rate of $580\text{ s}^{-1}$. Figure 41b shows a heavy pile-up of dislocations at low angle grain and twin boundaries in the same specimen. From the selected area diffraction pattern in Figure 41c, taken from the bright region in 41b, the $180^\circ$ rotation of the crystal planes around the $<111>$ twin axis can be seen. At a strain rate of $600\text{ s}^{-1}$, no major changes in the defect structure occur after $11.9\%$ of plastic strain when compared to the lower strain level, but the cell-like structure is revealed more clearly. Deformation twins still cannot be found in all grains, while some grains show extensive twinning. The observations from the TEM studies are generally in good agreement with those made with EBSD, showing strong grain orientation dependence for twinning as well as an increase in the twinning propensity with increasing strain rate.

Figure 40. Semitransparent orientation map (IPF-x coloring scheme) overlaying an EBSP quality map of a small-grained TWIP 1 steel after the material had been deformed at a strain rate of $10^{-3}\text{ s}^{-1}$ to 0.65 engineering plastic strain.
Figure 41. Defect structures in a TWIP 1 specimen deformed at a strain rate of 580 s\(^{-1}\) to \(\varepsilon = 5.6\%\). Deformation twins in a grain near the edge of the specimen (a), deformation twins and a strong pile-up of dislocations against grain and twin boundaries (b), and SAD-pattern of b (c).
6 SUMMARY AND CONCLUSIONS

The objective of the present study was to characterize the mechanical behavior of three typical representatives from the group of advanced high strength steels (AHSS) – a dual phase DP 600, a transformation induced plasticity TRIP 700, and five experimental high alloyed manganese twinning induced plasticity (TWIP) steels – at varying strain rates and temperatures. Both DP and TRIP steels are commonly used in various structural parts and components of cars, which makes their stress-strain response at high strain rates particularly interesting and important. TWIP steels, on the other hand, are a more recent innovation also with very promising properties for the automotive industry. The results of the mechanical studies were discussed in consideration of the theory of thermally activated dislocation motion in combination with mechanical twinning and phase transformation thermodynamics as well as adiabatic heating effects. Due to the special interest in the ductility of the studied materials, the majority of mechanical tests in this work were performed in tension. However, complementary compression test were conducted as well.

In high strain rate testing with the Hopkinson split bar technique the effects of inertia, wave propagation, and specimen geometry require special consideration when interpreting the test results. Especially tensile tests on sheet steels pose some challenges to the test setup design, e.g., assuring homogeneous deformation of the specimen, accounting for possible contribution of the deformation around the shoulder area, and specimen fixation. Therefore, in order to obtain as reliable measurement results as possible from the high strain rate tests on sheet steels, preliminary investigations on the influence of specimen geometry on the tensile Hopkinson split bar test results were conducted, both experimentally and by elasto-plastic FE simulations.

In the studies on the DP and TRIP steels, the tensile tests revealed positive but rather moderate strain rate sensitivity for both steels. The DP steel, however, appears to be clearly more sensitive to strain rate changes than the TRIP steel. This is explained by the effects of deformation induced heating on the martensite transformation based hardening of the TRIP steel, partially offsetting the direct effects of strain rate.

Up to about 7% of plastic deformation, DP 600 shows clearly higher strain hardening rates than TRIP 700, but at larger strains the situation changes. The difference can be attributed to the different strain hardening mechanisms acting in these two materials. The strain hardening behavior of the TRIP steel depends also quite strongly on temperature – the average strain hardening rate increases with decreasing temperature – to which the DP steel seems to be quite insensitive. At RT no significant difference in the strain hardening rate was observed between TRIP steel specimens deformed at different strain rates, although in general lower martensitic phase transformation rates $\partial V^{\gamma \rightarrow \alpha} / \partial \varepsilon$ were revealed for the material deformed at high strain rates by X-ray diffraction measurements, leading to the suggestion that in TRIP steels twinning compensates for the reduction in martensitic phase transformation rate at high strain rates. EBSD and TEM
studies on fully austenitic high manganese steels indeed revealed a strong increase in the twinning propensity at high strain rates.

The TRIP steel clearly outmatches the DP steel in elongation and total energy absorption. The DP steel, however, absorbs more energy over the whole studied range of temperatures and strain rates, when deformations only up to 10 % are considered. In automotive applications, this in fact could be closer to the real case, where only limited deformation is allowed because of passenger safety reasons.

The observed mechanical behavior of the DP 600 steel at different temperatures and strain rates is typical for metals and alloys that deform and strain harden primarily by dislocation glide and dislocation interactions. The general equivalence of high strain rates and low temperatures, quantitatively expressed by an Arrhenius type equation, readily describes the mechanical behavior of the DP steel over the tested temperature and strain rate ranges. The mechanical behavior of the TRIP steel appears to be more complex because of the different phase transformation modes and mechanisms in the austenite phase, which largely control the hardening behavior of this type of steels. The characteristic temperature $M_S$ of the TRIP 700 steel, where the austenite-to-martensite phase transformation mode changes from stress-assisted to strain-induced, appears to be close to -10 °C. Maximum elongations and thereby the maximum TRIP effect for this steel was observed between 75 and 150 °C, which would therefore be the optimal temperature range for industrial forming operations.

Five fully austenitic experimental high manganese TWIP steel grades, differing slightly in their contents of Mn, Al, Si, and Nb, were produced based on thermodynamic stacking fault energy (SFE) calculations. The stacking fault energies of the steels varied between about 20.5 and 42 mJ/m² at room temperature, causing differences in their mechanical behavior, especially when the testing conditions, i.e., temperature and/or strain rate were changed.

Of the three TWIP steels tested in tension, TWIP 1, which does not contain niobium, showed clearly the lowest strength levels at all testing conditions. At room temperature, TWIP 2 and TWIP 3, which due to their different contents of aluminum and silicon have the largest difference in their stacking fault energies, revealed rather similar strength levels. The strain hardening behavior of TWIP 3 with the highest SFE, however, showed at all temperatures some differences compared with TWIP 1 and TWIP 2. For example, TWIP 3 revealed in general the strongest strain hardening up to about 20-25 % of plastic strain, but for higher plastic strains the opposite was the case.

The strain rate sensitivity of all three TWIP steels turned out to be relatively weak or at most moderate in the strain rate range $10^3$...750 s⁻¹. At the strain rate $\dot{\varepsilon} \sim 1000$ s⁻¹, however, all studied TWIP steels showed the commonly observed steep upturn in their strength levels.

When tested at subzero or slightly elevated temperatures, the strength values of the studied TWIP steels increased steadily with decreasing temperature, while both uniform
and total elongations showed maximum values in the temperature region of -50 °C < T < 0 °C, where mechanical twinning has the strongest effect on the mechanical properties of austenitic steels. At high strain rates combined with high initial testing temperatures, all TWIP steels showed a clear deterioration in their elongation values, which was attributed to the increase in the stacking fault energy due to the higher initial test temperature in combination with adiabatic deformation heating, causing dislocation slip to replace twinning and thus resulting in a reduced TWIP effect. TWIP 3 showed generally the weakest strength-ductility combination, most probably because its SFE was too high to truly benefit from twinning. TWIP 2, which has the lowest stacking fault energy $\gamma_{SFE}$ of the studied steels, revealed the most beneficial elongation-strength combination, which persisted down to low temperatures, probably because of the mechanical martensite transformation setting in when the SFE was lowered into the appropriate region. Additions of Nb seemed to increase the strength values of the TWIP steels without compromising their ductility.

The microstructure and texture evolution of the hot rolled TWIP steels was studied using electron backscatter diffraction and the defect structures by transmission electron microscopy. For this purpose, specimens were uniaxially deformed to different degrees of plastic deformation at various temperatures and strain rates. EBSD measurements were conducted on tensile specimens, while TEM studies were performed on compressive specimens. The EBSD studies clearly revealed texture dependence of the twinning propensity. Lowering the deformation temperature was shown to result in decreased elongation values due to premature twinning and ‘underdeveloped’ texture. The observation showed that the explanation for the reduced elongation values at low temperatures usually given in the literature [18, 71-72], i.e., reduced elongations due to extensive twinning at low strain levels, seems to be too conservative. In fact, no extensive twinning was observed at low temperatures and twinning only occurred in favorably oriented grains. Fracture occurs before grains have rotated into orientations favorable for twinning and therefore no beneficial TWIP effect can be observed.

An increase in strain rate resulted in increased twinning propensity, which is in good agreement with the observed mechanical behavior. A reduction in the grain size was observed to strongly suppress mechanical twinning.
APPENDIX

Table 7. Yield strength and UTS of the studied DP, TRIP and TWIP steels at different deformation temperatures at strain rates $10^{-3}\text{ s}^{-1}$ and $1250\text{ s}^{-1}$ (values in MPa).

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Table 8. Uniform elongations $e_{un}$ and total elongations $e_{f}$ of the studied DP, TRIP and TWIP steels at different deformation temperatures at strain rates $10^{-3}$ s$^{-1}$ and 1250 s$^{-1}$ (values in %).

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[3] Hopkinson B., A method of measuring the pressure produced in the detonation of high explosives or by the impact of bullets, Philosophical transactions of the Royal Society in London, A 240 (1914)
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Publication IV

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Deformation behavior of TRIP and DP steels in tension at different temperatures over a wide range of strain rates


Publication V

S. Curtze, V.-T. Kuokkala, M. Hokka, and T. Saarinen

Microstructure and Texture Evolution of High Manganese TWIP Steels

In DYMAT 2009 (2009) 1007-1013
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September 7-11, 2009, Brussels, Belgium
