

Influence of short heat pulses on properties of martensite in medium carbon steels

Krste Cvetkovski¹, Johan Ahlström*, Birger Karlsson²

Chalmers University of Technology, Department of Materials and Manufacturing Technology, SE-41296 Gothenburg, Sweden

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ABSTRACT

The process of tempering a martensitic medium carbon steel was investigated with the aim to study resulting material properties. The experimental results were used to model residual stresses caused by local heating, with the finite element method. Tempering was followed for times from 0.1 s up to 1 h by using laser heating and conventional salt bath furnace treatments within the temperature interval 500–700 °C. In addition, the thermal expansion was evaluated using dilatometry. Experiments showed that the initial stages of martensite decomposition, associated with loss of crystal tetragonality, proceed almost instantly. An initial large decrease of hardness within the first tenth of a second of the tempering process was measured, followed by only limited further softening with increased tempering time. Thus for the current material the tempering time had limited influence on hardness, governed primarily by the peak temperature during the heating process. Finite element modelling of rapid local heating and cooling showed that the tempering behaviour and associated dilatation effects yield a peak temperature dependent residual stress field with a broad tensile stress distribution for the case of un-tempered martensite. However, for tempered martensite the residual stress field depends primarily on the heating rate and peak temperature and shows large gradients with tensile stresses in the surface and compressive below. Thereby, for both cases, residual stresses were obtained but with completely different residual stress gradients.

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1. Introduction

Pearlitic steels are commonly used for a wide range of applications due to their strength–wear properties and cost efficiency. In industrial processes like laser or electron beam welding, laser flash hardening, adiabatic shearing and in operating environments such as wheel–rail contact in railway traffic, the material is sometimes subjected to rapid heating and cooling within a localised volume [1–6]. Such local heating often has a short duration and is normally followed by self-cooling with large cooling rates. If the heating pulses cause peak temperatures above some 400–600 °C, softening of the pearlitic microstructure takes place resulting in easier plastic deformation and distortion under loading [7]. Even more important are heating cycles extending over the austenitisation temperature (T_{A1}) which may lead to martensite formation during rapid cooling. The volumetric

increase caused by martensite formation leads to stress fields and may cause local cracking. Subsequent re-heating in the affected area above T_{A1} allows for new formation of martensite, but, for lower temperatures tempering occurs with associated decrease of volume. A key issue is the speed of martensite softening and how internal stresses develop during tempering in these shock heating and cooling sequences. As the kinetics of martensite tempering is not well known for such short heating and cooling histories the present study was undertaken. Methods used to predict the material properties after tempering, for example the Hollomon–Jaffe parameter are suitable for long tempering times when appropriate model calibration is performed [8–12]. However for tempering of seconds or less, it may not be valid to extrapolate material properties from long time tempering experimental results [13].

In this study, laser- and salt bath heating have been used to study martensite tempering and resulting mechanical properties, for time periods from a tenth of a second up to 1 h. Dilatometric experiments have also been made to determine associated phase transformations. Finally the effects of rapid heating and cooling of martensite and tempered martensite have been analysed with finite element (FE) simulation to evaluate residual stresses within a component, for the case of local heating.

* Corresponding author. Tel.: +46 31 772 1532; fax: +46 31 772 1313.

E-mail addresses: krste.cvetkovski@chalmers.se (K. Cvetkovski), johan.ahlstrom@chalmers.se (J. Ahlström), birger.karlsson@chalmers.se (B. Karlsson).

¹ Tel.: +46 31 772 1244.

² Tel.: +46 31 772 1242.

Table 1
Chemical composition of material in wt%.

C	Mn	Si	Cr	Mo	V	Cu
0.50	0.75	0.37	0.12	< 0.08	< 0.06	< 0.11

2. Experimental

Ferritic–pearlitic steel with some 90 vol% fine pearlite and 10 vol% proeutectoid ferrite was used as a starting material. The nominal chemical composition of the steel is presented in Table 1. Martensitic material was obtained by austenitisation at 900 °C in a tube furnace with protective argon atmosphere for 5 min and thereafter quenched in brine. The Vickers hardness after this treatment was 780 ± 10 HV30.

Samples for laser experiments were coin-like with a thickness of 3.0 mm and diameter of 22.5 mm. To attain rapid heating and cooling of the material, a 2.8 kW diode laser was used for heating a spot on the sample surface. To have uniform laser radiation absorption between samples, the radiated surface was ground and then polished with 3 μ m diamond suspension. The laser beam was guided via an optical fibre and defocused to a diameter of 6 mm at the centre of the coin surface. The temperature within the innermost 2 mm diameter of the laser exposed area was measured with a pyrometer, using a sampling frequency of 60 kHz. The samples were mounted in a temperature controlled copper sample holder within a vacuum chamber, evacuated down to 10^{-8} bar during the experiments.

Tempering treatments for times longer than 10 s were made in a salt bath furnace using sodium nitrate for tempering at 500 °C and barium chloride salt for higher temperatures. Samples were thin rectangular plates with 2 mm thickness. A type S thermocouple inserted within a reference sample was used to obtain the actual sample temperature. Hardness indentation was made on a cross-section to avoid influence of thin decarburised layers on the surface.

For laser tempered samples, hardness (HV5) was measured along two orthogonal lines on the surface both crossing the centre of the irradiated area. On a perpendicular section containing one of the surface lines with surface hardness indents, hardness mapping was made with a distance between indents of 0.5 mm. Retained austenite in martensite was measured with X-ray diffraction technique using an Xstress 3000 X-ray equipment with vanadium K_{β} blocking filters.

Dilatometry was performed for martensite and pearlite by means of a push-rod dilatometer, Netzsch Dil 402 C, with a tungsten thermocouple in protective atmosphere. Thermal expansion and solid state transformations of as-quenched and tempered material were studied using single and multi-runs. Analyses were made for different maximum temperatures up to 990 °C, all starting at 30 °C (isotherm, 30 min before each run). Heating and cooling rates were 5 °C/min and 10 °C/min respectively. Alumina (Al_2O_3) was used as a reference. Samples were cylindrical with diameter 6 mm and length 50 mm.

3. Finite element model

3.1. Software and definitions

Simulations were performed with the finite element (FE) method in the commercially available software Abaqus 6.9 Standard by Dassault Systèmes. The software has the ability to link user-defined subroutines into the compilation. Three such subroutines were used in the current work: UEXPAN for implementation of phase

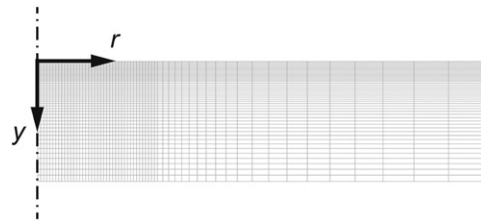


Fig. 1. Mesh and definition of directions.

dependent thermal expansion, DFLUX to model the heat distribution from the laser source and HETVAL to model martensite tempering and to keep track of phase contents [14]. An axisymmetric mesh with 3000 elements was used, see Fig. 1; the element size close to the heated surface was 0.025 mm thick and 0.075 mm wide. Coupled temperature–displacement analyses were run with four-node bilinear displacement and temperature elements (CAX4T). Each of the 3111 nodes has three degrees of freedom: temperature and displacement in radial and depth directions.

4. Analysis

4.1. Temperature control and thermal properties

The initial sample temperature was set to 20 °C. On mounting into the sample holder the sample was assumed to be heated to the temperature of the holder, which in accordance with the experimental setup was set to 100 °C. The laser heating of the most central part of the surface was done with different powers for different cases, and adjusted to achieve a top-hat temperature–time curve in conformity with the experiments. It was found in the experiments that the peak temperatures dominate the following material behaviour by controlling the degree of tempering. Initial cooling was modelled by film cooling to the sample holder; however, the hottest parts of the specimen were mainly self-cooled. As the material models are formulated as rate-independent, the exact cooling curve is not required when temperature gradients are low. Mechanical interaction between the sample and the sample holder was neglected. Final cooling to 20 °C was accomplished by setting a temperature boundary condition to the nodes. A final temperature of 20 °C was required to allow comparisons with experimental hardness data.

The power from the laser was applied on the top surface at $0 \leq r \leq 3.0$ mm with scaling of absorption coefficient to achieve a narrow temperature range during the time of exposure within the pyrometer sampling area, $0 \leq r \leq 1.0$ mm. Temperature dependent specific heat and thermal conductivity were implemented for the actual chemical composition of the coin sample based on data from Thermo-Calc (Thermo-Calc software), reported in earlier publications [15].

4.2. Mechanical properties and dilatation

The elastic response of the material was implemented with slight temperature dependence of the Young's modulus E , decreasing from 210 GPa at 20 °C down to 150 GPa at 700 °C according to data from JMatPro6.0 (Sentese software) calculated for the actual chemical composition. The plastic behaviour was described using a material model with isotropic hardening pre-defined in Abaqus. This model has the capability to account for both temperature and solution dependent variables, in this case a tempering parameter explained below. Underlying data were compiled from own experiments (compression tests at temperatures from 25 to 275 °C), extrapolated to higher temperatures

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