MICROSTRUCTURE BASED HARDENING MODEL FOR TRANSFORMATION INDUCED PLASTICITY (TRIP) STEELS

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Keywords: TRIP steels, phase transformation law

1. Introduction

Multi phase steels especially TRIP steels combine a high strength and a high formability, which is a condition for deep drawing parts in the automotive industry. The Transformation-Induced-Plasticity-effect (TRIP) is responsible for an extra work hardening during straining and the energy absorption during this effect enables an application to crash relevant components. But the damage behaviour of TRIP steels is poorly understood and the application of them is restricted.

2. Motivation and aim

This work presents a microstructural numerical investigation of the influence of the TRIP effect and its different contributions to the evolution of the microstructure to get a better understanding of the microstructural damage mechanisms and to ensure an application in automotive parts.

Due to industrial conditions of processing routes TRIP steel sheets show a change of the morphology over the sheet thickness, which therefore results in a change of failure mechanisms within the part. The reasons for this are segregation and banding effects within the sheet.

Fig. 1 shows such resultant failure mechanisms in two different industrial TRIP sheets (left: debonding of hard martensitic bands from the surrounding ferritic matrix, right: cleavage fracture zones surrounded by a ductile network of voids). Typically, TRIP steels consist of a ferritic / bainitic-ferritic matrix with embedded islands of retained austenite and some martensite (see Fig. 2).

During deformation, these retained austenite grains undergo a phase transformation to the harder phase martensite. The locations of these grains in the microstructure are of major importance because they influence the impact of the TRIP effect, the microstructural localization and therefore the macroscopic deformability of the material.

3. Phase transformation law and microstructural finite element model

Different deformation mechanisms occur within different temperature ranges. Below the martensite start temperature $M_s$, martensite can be formed by cooling. Above $M_s$, an additional driving force is needed and phase transformation can be achieved by deformation of the microstructure. Within this range, martensite is formed at same nucleation sites as during cooling.

Above a so called $M^\sigma$-temperature, martensite is formed at intersections of shear-bands consisting of $\varepsilon$ martensite with hcp crystal structure. These shear bands emerge after plastic deformation of retained austenite (see Fig. 3).

The presented model considers the TRIP effect in the strain-induced regime which is based on the phenomenologi-
numerical model of Olson and Cohen. According to Iwamoto et al., the plastic strain rate is defined as the sum of the plastic strain rate which is induced by slip deformation in austenite and martensite and the transformation strain rate which is induced by phase transformation. The constitutive model is based on the publication of Sierra and Nemes. The transformation plastic strain rate is divided into two parts, the deviatoric contribution related to the shape change, and the dilatational contribution related to the volume change. Hence, the overall plastic strain rate tensor is given by the following expression:

\[
\varepsilon_\text{pl}_{ij} = \varepsilon_\text{pslip}_{ij} + \varepsilon_\text{pshape}_{ij} + \varepsilon_\text{ptrans}_{ij},
\]

with the slip deformation plastic strain rate tensor:

\[
\varepsilon_\text{pslip}_{ij} = \varepsilon_\text{pslip} \frac{\partial F}{\partial \sigma_{ij}},
\]

The shape change plastic strain rate tensor is defined as follows:

\[
\varepsilon_\text{pshape} = R \varepsilon_\text{pslip},
\]

with the magnitude of shape change:

\[
R = R_0 + R_1 \left( \frac{\sigma_{ij}}{\sigma_{ij}^\text{m}} \right),
\]

the yield stress of the parent austenitic phase is \( \sigma_\text{ya} \) and the dimensionless constants are \( R_0 \) and \( R_1 \). The increase in martensite volume fraction \( f \) is determined by plastic strain in the austenite and is written in the following form:

\[
f = A (1 - f^\text{a}) \varepsilon_\text{pslip},
\]

with the equivalent plastic strain rate in austenite due to slip deformation \( \varepsilon_\text{pslip} \), and \( A \) is defined as follows:

\[
A = \alpha (f^\text{a})^{-\alpha} (1 - f^\text{a})
\]

with the volume fraction of shear bands \( f^\text{a} \) and the hardening exponent \( n \). \( \alpha \) is a temperature dependent coefficient which is defined by:

\[
\alpha = \alpha_1 T^2 + \alpha_2 T + \alpha_3 - \alpha_4 \Sigma
\]

where \( \alpha_{1-4} \) are material constants, \( T \) is the temperature and \( \Sigma \) is the stress triaxiality. The driving force for martensite formation \( \eta \) is also dependent on the stress state and given by the following equation:

\[
\eta = \eta_1 + \eta_2 \Sigma
\]

with the constants \( \eta_{1-2} \).

Finally, the dilatational plastic strain rate tensor is expressed by:

\[
\varepsilon_\text{ptrans} = \frac{1}{3} \Delta \varepsilon_\text{pslip},
\]

where \( \Delta \) accounts for the volume change which lies in a range between 0.02—0.05 and \( \delta_0 \) is the well known Kronecker delta.

The constitutive law which describes the behaviour of the composite similar material is given by:

\[
\sigma_\text{ij} = D_\text{ij} \varepsilon_\text{ij}^\text{a} - P_0 \frac{\partial F}{1 + v \partial \sigma_0},
\]

where

\[
P_0 = \frac{E}{1 + v} \frac{\partial F}{\partial \sigma_0}.
\]

\[
Q_0 = -P_0 \Sigma + \frac{1}{3} \delta_0 \frac{E}{1 - 2v}.
\]

\( \sigma_\text{ij} \) is the rate of Cauchy stress, \( D_\text{ij} \) is the elastic stiffness tensor, \( E \) is the Young’s modulus and \( v \) is the Poisson’s ratio.

The phase transformation was used as a user defined material subroutine (UMAT) within the commercial finite element package ABAQUS. A TRIP microstructure was simplified to a 2 phase microstructure consisting of retained austenite islands embedded in a ferritic matrix. The principal can be seen in Fig. 4. The simplified microstructure (upper right image) was nearly created from a given SEM image (upper left image). For mesh generation, at first a geometry file was created (IGES file) by using a Voronoi cell generator (see lower left image). Finally, the mesh was built by a geometry export into the meshing tool.
Hypermesh® and the material properties were given according to the upper right image. The final mesh consists of 89102 triangular elements and the calculations were taken out under a plane strain condition. To investigate the influence of different contributions to the TRIP effect, 4 material models were investigated. The isotropic elastoplastic constitutive law of the ferritic matrix was kept constant, but the retained austenite stability was modified by adjusting of certain material parameters. The different assumptions are given in Table I.

### 4. Results and discussion

Fig. 5 shows the martensite volume fraction and Fig. 6 the Von Mises stress distribution after uniaxial tension with applied homogeneous boundary conditions until 10% deformation of the microstructure by consideration of all effects. The position alters stress distribution within the matrix and consequently strain localization which emerges as shear-bands of high plastic strains. That means the position influences the local stresses and therefore the phase transformation.

![Fig. 5. Distribution of martensite volume fraction after 10% deformation (uniaxial tension in x direction with homogeneous boundary conditions) by consideration of all contributions to the TRIP effect](image)

![Fig. 6. Stress distribution after 10% deformation (uniaxial tension in x direction with homogeneous boundary conditions) by consideration of all contributions to the TRIP effect](image)

In addition it can be seen that locations with high martensite contents have higher stresses due to their increased strength (see Fig. 6). This increased strength occurs suddenly because of a rapid phase transformation. During deformation, the phase transformation alters the stress distribution within the matrix and therefore changes the accumulation of plastic strain and damage which postpones the final failure. To investigate the influence of different contributions to the TRIP effect, some parameters in the model definition were set to zero (see Table I). In material A there is no consideration of the influence of stress triaxiality on phase transformation (see equation (8)). As already mentioned, the strain induced phase transformation assumes that martensite forms at the intersection of shear-bands. To investigate this effect the influence of stress triaxiality on shear-band formation was neglected in material B. In material C and D the Greenwood-Johnson effect was investigated, where material C assumes no shape change (see equation (3)) and material D allows no volume dilatation (see equation (9)) during martensite formation.

Material A and D show that stress-state independent martensitic formation and volume dilatation have a minor influence on phase transformation than the influence of stress state dependent shear-band formation and shape change (see Fig. 7). In comparison to all active mechanisms, the martensite fractions in material A, B and D are not as high as by consideration of all effects.

The neglecting of influence of stress-state on shear-band formation results in a significant lower martensite volume fraction. That means that the formation of shear-bands has a major importance for the TRIP effect. A shape-change free phase transformation results in much higher martensite content, because the matrix does not need to oppose a strong resistance against the lattice change.

Like in the previous results it can also be seen that high martensite volume fractions lead to higher local stresses due
to increased strength (see Fig. 8). A strong deviation of stress distribution within the grain gives a heterogeneous stress distribution within the matrix material.

5. Conclusion

The numerical investigations showed a strong influence of the morphology on the phase transformation. Therefore a local consideration of the microstructure is required. The probability of shear-band formation and the shape change during transformation play a major role for the TRIP effect.

The partial transformation within the austenitic grains results in a heterogeneous distribution of the stress fields. Therefore the further increase in martensite content is influenced.

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A model is presented which takes the strain-induced martensite formation in low-alloyed TRIP steels into account. A parameter study is conducted to demonstrate the influence of individual mechanisms (e.g. volume dilatation). A correlation between martensite volume fraction and local stresses is depicted.
QUALITY EVALUATION OF RESISTANCE SPOT WELDS OF HOT-DIP GALVANIZED SHEETS IN CORROSIVE ENVIRONMENT

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1. Introduction

Significant contribution to rapid development of the automotive industry is provided by continuous improvement of quality and performance parameters of produced vehicles. One of the important requirements for materials used for vehicles is their corrosion resistance, which leads to the increasing share of the automobile parts with protective, especially zinc coatings1,2. Galvanized sheet produced by zinc electrolyzing or dipping shows very good resistance to atmospheric corrosion. Surface treatment by galvanizing ensures good corrosion resistance of metal chassis and allows producers to offer more than 10 year life guarantee of the chassis3,4. However, this type of surface treatment causes considerable problems for joining metal chassis. Conventional methods (MIG welding, resistance spot welding), but also unconventional methods (laser welding) lead to evaporation of protective layers near weld due to the heat and loss of the corrosion resistance of the welded parts5. The test samples were cut to size 90 × 40 mm according to STN EN 051 122, length of lapping was 30 mm. Spot welding was carried out on pneumatic spot-welder BPK 20 with welding electrodes CuCr with 5 mm diameter. Selection of the welding parameters was based on the recommendations of British Standard (BS 1140) for spot welding of coated sheets and knowledge of weld growth mechanism6. However, the correction of the parameters was necessary, for the specific spot-welder. Welding parameters were: contact force Fz =2.6 kN, welding time t = 14 periods, welding current 85 kA (ref.7,8). Tensile tests according to STN 05 1122 were used for the evaluation of mechanical properties of spot welds before and after the exposure to the corrosive environments. The test of spots and full penetration welded joints were carried out on a test machine TIRA-test 2300, (VEB TIW Rauenstein).

The samples for microhardness evaluation were prepared according to ISO 6507-1 and ISO 6507-2. Microhardness on the cross scratch patterns was measured according to STN EN 1043-2 for hardness evaluation of particular areas of the weld and its transient phases.

The tests were carried out on Shimadzu Bubo microhardness tester with indenter Vickers under a load of 0.01 kN. Intender load time was 10 s.

The microhardness was measured on cross scratch patterns according to scheme in Fig. 1. Three measurements were carried out: in the base material, heat affected zone and weld metal in DX54D+Z.

Welds were exposed to salty environment of 5% NaCl solution and ecological solution Solmag used during winter season on roads. The above environments simulated conditions of vehicle operation6,7. Gravimetric method was used for the evaluation of the corrosion processes. Potentiodynamic polarization curves were measured for determination of the electrochemical characteristics of these materials. The values of EK – corrosion potential by Tafel, Ik – corrosion current density and Rp – polarization resistance were measured. Electrochemical characteristics of the tested samples were evaluated before exposure to the corrosive environment and after 240, 480 and 720 hrs of exposure according to Tafel and Stern method. A computer-controlled potentiostat/galvanostat VOLTALAB 21 PGP201 (producer Radiometer Analytical, Denmark) with software VOLTAMASTER 4 was used for potentiodynamic measurement; a schematic diagram of appa-

![Image](image-url)
ratus in three-electrode configuration is in Fig. 2. Scanning electron microscope JEOL JSM - 7000F was used for microscopic observations.

3. Results

Fig. 3 shows that the force determining the strength of spot welds before their exposure to the corrosive environments was ~ 11% higher in material B in passivated state than in the case A in nonpassivated state. Both materials were exposed to the corrosive environment of 5% NaCl solution, after which increased carrying capacity of welds was observed in material B (passivated state). When the values of $F_{\text{max}}$ before and after the exposure to the corrosive environment are compared, strength increase was found in for both types of spot-welds (for samples A by ±0.45%, and samples B by ±3.1%). Both materials were exposed to the corrosive environment of ecological solution Solmag, after which increased carrying capacity of welds was observed in material B (passivated state).

Fig. 4 shows the macrostructure of the cross-section of the weld. No signs of defects and, heat affected areas and local flashing of the Zn layer in the contact place of welding electrodes can be seen.

Fig. 5 illustrates the formation of the zinc corrosion products on the surface of the test samples after the corrosion tests in NaCl environments. It is possible to observe white zinc hydroxide, products of corrosion, which was confirmed by EDX analysis.

The factors causing formation of white corrosion on zinc layer are condensation of air moisture, and the effect of the electrolyte.

The measured values of microhardness are represented in Fig. 6.

The measurements show that changes in parameters of resistance spot welding do not cause significant microhardness changes in the weld metal or heat affected zone of DX54D+Z steel.

Aggressive environment does not influence microhardness of welds.

Galvanized sheet behaves like a galvanic cell when the moisture is present, where zinc layer with lower electric potential is the anode and steel represents the cathode. Oxygen is very important factor in the atmospheric corrosion because
the thin layers of moisture act as diffusion barriers into the condensed moisture. Therefore, reaching the dew point is not necessary for the electrochemical reaction. As a result, very thin layer of electrolyte solution (0.005 to 0.15 mm) forms on a metal surface already at a low humidity (corrosion humidity ~ 60%).

When the samples were exposed to the environment of 5% NaCl solution for up to 240 hours, their weight increased due to formation of corrosion products, mainly zinc. However, when the exposure of samples to chloride environment exceeded 240 hours, there was a significant weight loss of zinc coating, caused by release of corrosion products into the electrolyte. The protective efficiency of zinc coating is significantly reduced in the aggressive environment. The comparison of weight losses of A and B samples in 5% NaCl solution showed a difference of 53.66 %, Fig. 7. It indicates that the passivated samples are more resistant to mass loss than non-passivated in 5% NaCl solution.

When the samples were exposed to the environment of the ecological solution Solmag, their weight did not significantly change during the time of exposure. The potentiodynamic polarization curves of the samples measured in both states (non-passivated and passivated surface) evaluated by Tafel method are shown in Fig. 8.

Non-passivated zinc-coated sheet (A) and passivated zinc-coated sheet (B) in 5% NaCl solution after 720 hours exposure show the same corrosion current densities \( J_k \), i.e. passivation is effective only in the early phase of exposure, Fig. 8.

The influence of chloride corrosive environment on galvanized sheets is so intense, that after several tens of hours’ exposure there is no significant difference between passivated and non-passivated surface in zinc-coated steel sheet.

An interesting phenomenon occurred between 240 to 480 hours of passivated zinc-coated steel’s exposure in non-aggressive environment, when the corrosion products show a high polarization resistance \( R_p \) to corrosion current flow (1.2–1.6 M\( \Omega \); see Fig. 9). However, after 720 hours of exposure, this passive layer is damaged even in low-aggressive environment and the “run-off” effect of zinc (emissions of Zn into the environment) predominates.

4. Conclusion

Welds on passivated sheets have higher carrying capacity in their initial state than the welds on non-passivated sheets. After exposure of samples to both corrosive environments the welds on passivated samples have higher carrying capacity. This increase is caused by zinc corrosion products, which, despite their low adhesion to the substrate, increase the strength of welds.

The measured values of microhardness shown in Table 5 are in accordance with the observed microstructures and their components. The maximum values were measured in the sample in WM (261 HV 0.01). This value corresponds to

![Fig. 7. Weight loss in corrosion environments; a) material A; b) material B](image1)

![Fig. 8. Tafel method for determination of \( J_k \)](image2)

![Fig. 9. Stern method for determination of \( R_p \)](image3)
Microhardness of a dominant bidentic compound in the weld metal. The change of microhardness in the weld metal was continuous and it corresponded to mixing the weld metal of particular types of steel.

Microscopic analysis proved high-quality of welds, which also showed that optimum technological parameters of welding were used. Local flashing of zinc layer was observed in the contact place of welding electrodes. After exposure of samples to corrosive environments, zinc corrosion products, so-called “white corrosion”, formed, as a result of condensation of air moisture or effect of the electrolyte.

The weight of samples exposed to ecological solution Solmag during corrosion tests increases in consequence of accrued zinc corrosion products, and the corrosion process is slower than when exposed to NaCl solution. In the relatively aggressive corrosive environment of NaCl there occurs significant weight loss. The potentiodynamic electrochemical measurements show that both types of sheets have the same corrosion current densities after 720 hours of exposure. That indicates that passivation is useful only in the early phase of exposure of materials to corrosive environment. Corrosion products with a high polarization resistance Rp form on the surface of the passivated samples, keeping their favorable properties up to approximately 500 hours of exposure to corrosive environment.

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The contribution dealt with the evaluation of quality of resistance spot welds. Steel sheets of DX54D+Z were used for experiments. Welds on passivated sheets have higher carrying capacity in their initial state than the welds on non-passivated sheets. This increase is caused by zinc corrosion products, which increase the strength of welds. The weight of samples exposed to ecological solution Solmag during corrosion tests increases in consequence of accrued zinc corrosion products, and the corrosion process is slower than when exposed to NaCl solution. Both types of sheets have the same corrosion current densities after 720 hours of exposure.
THERMAL EFFECTS ON STEELS AT DIFFERENT METHODS OF SEPARATION

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Keywords: microhardness, plasma cutting, laser cutting, flame cutting, Vickers hardness

1. Introduction

As regards material separation methods, more sophisticated and effective methods are used today beside the classical ones. The new everyday development and changes of modern electronic technology and machine tool engineering brings new generations of many different products. Global development of these methods plays a vital role in the industrial process.

The cutting of material is usually the first operation that has to be done when producing some component. There are different types and methods of steel cutting. The most common methods of material splitting are cutting by water jet, laser, plasma and flame. In some cases, the heat caused by cutting influences the outer layers of the processed material. Among two other common methods of cutting there are cutting by tools with defined geometry and abrasive processing – such as grinding or sanding. Some progressive methods could be also applied using physical, electrical and chemical processes as well as other sources of energy for removal of the material.

All these methods of material splitting force us to explore their impact to the qualities of cut material, especially to the layer that is in direct contact with the cutting tool. Learning about changes in structure and qualities of processed materials could assist in development of material separation and processing technologies.

1.1. Testing of micro-hardness

Development of hardness measurement methods for individual structural parts – measurement of micro-hardness – was an important requirement for studying the structure of solids. Micro-hardness testing devices are inbuilt in metallographic microscopes or they are used as stand-alone devices.

Measurement of micro-hardness is done by pushing of the diamond body, shaped in Vickers or Knoop pyramid, into the surface of measured body by forces ranging from 1 g (0,0009807 N) to 1000 g (9,807 N). While common testing of (macro) hardness according to Vickers is done by means of load of 10 N to 1200 N, the forces weaker than 10 N were used first in 1932, when such testing was done at the National Physical Laboratory in UK.

The accuracy of measurement is lower as the applied forces are weaker (smaller size of the dint). Moreover, as the load force weakens and enters the area of micro-hardness, the Vickers hardness is no more depending on the power of applied load (unlike in macro-hardness that is measured according to geometric similarity of individual dints).

1.2. Micro-hardness according to Vickers

Test of hardness according to Vickers is prescribed by European standard ČSN EN ISO 6507-1, for three different areas of testing load (Tab. I).

Table I

<table>
<thead>
<tr>
<th>Area of testing load</th>
<th>Symbol of hardness</th>
<th>of Previous marking (ISO 6507-1:1982)</th>
</tr>
</thead>
<tbody>
<tr>
<td>F [N] ≥ 49,03</td>
<td>≥ HV 5</td>
<td>Test of hardness according to Vickers</td>
</tr>
<tr>
<td>1,961 ≤ F &lt; 49,03</td>
<td>0,2 to &lt; HV 5</td>
<td>Test of hardness according to Vickers with low load</td>
</tr>
<tr>
<td>0,09807 ≤ F &lt; HV 0,01</td>
<td>&lt; Test of micro-hardness according to Vickers</td>
<td></td>
</tr>
</tbody>
</table>

The penetrating body – made of diamond shaped as a regular tetragonal pyramid with the square base and with preset vertex angle (136°) between opposite walls – is pushed against the surface of testing body. Then, the diagonal size of the dint left after load removal is measured (Fig. 1).

\[ F = 0,1891 \times \left( \frac{F}{d^2} \right) \]

where F – Testing load in N, d – Arithmetic average of two diagonals \(d_1, d_2\) in mm.

Fig. 1. The basic principle of hardness testing according to Vickers
Vickers’ hardness is then expressed as the ratio of the testing load applied to dint area in form of regular tetragonal pyramid with square base and the vertex angle equal to the angle of penetrating body (136°).

Hardness measurement by applying low load and testing of micro-hardness require exact measuring of dint diagonals. Evaluated testing sample has to have smooth and even surface, without flakes, lubricants and foreign particles. Evaluated sample is thus prepared as a metallographic scratch pattern, i.e. no deformation or thermal influence to the surface must occur. Generally, the preparation is done by wet grinding and then by polishing on diamond pastes or by electro-polishing. Exact method of sample preparation is chosen according to respective materials.

Thickness of testing body or layer has to be at least 1,5 times longer than the diagonal length of the dint

2. Experimental part

There were three methods of material cutting selected in the experimental part. All three methods have a thermal impact to the outer layer. Steel 11373 was selected for testing of thermal influence in the cutting area. The testing samples were made by using the technology of cutting by laser, plasma and flame. The sample preparation was done in companies specialized to these technologies of material cutting.

2.1. Micro-hardness testing according to Vickers

A device used for micro-hardness testing is called micro-hardness tester. The modern digital device can measure \( d_1 \) as well as \( d_2 \) dimensions directly in digital form; then a computer computes the average of \( d_1 \) and \( d_2 \), which is followed by HV computing and further data processing, so that different needs are met. At the same time, the whole micro-hardness testing process is controlled digitally so that highly effective and exact measurement can be obtained.

Micro-hardness testing of structural components and phases is one of the most important tools used in metallographic industry. By means of micro-hardness testing, individual structural components could be identified relatively exactly, such as lower and upper bainit, lowly and highly carbonated martenzit etc.

2.2. Sample preparation

As the sample surface must have a metallic shine and the surface roughness must not exceed Ra 0.2, the sample preparation is very important step. In general, the sample could be fixed in a gripping or set into spaces and after grinding by soft grindstone, the sample has to be sanded by metallographic sand papers of different numbers so that the dint could be clearly seen.

2.3. Laser

Testing samples were made by means of TRUMPF laser. The power used for preparation of the testing sample was 2400 W amd pressure 0,6 bar. The temperature in controlled cutting process reaches up to \( 10^6 \) °C, (cutting speed was \( 2.7 \) m min\(^{-1} \)). Maximum power of respective laser is 5 KW with accuracy of 0,1 mm.

The original structure of measured sample was a pearlitic one, respectively ferrite-pearlitic, with hardness ranging from 170 HV to 180 HV. Cutting by laser has a thermal impact on the original pearlitic structure and changes it to bainitic structure. The change is caused by great difference in temperature between the cutting area and the surrounding environment. The temperature in the cutting area is approximately 106 °C. The standard temperature of the surrounding environment is 22 °C, which causes very quick cooling creating the non-equilibrium state (hardening) in the cutting area and thus creating the structure of bainit. The measured hardness value in the cutting area was approximately 340 HV (Fig. 3) that corresponds to structure of upper bainit. Due to the characteristics of the tested material (11373), especially due to its low contain of carbon (0,17 %), lower values of...
hardness can be expected. The measured values show that the depth of the influenced layer is up to 1–1,5 mm. The changes of hardness are insignificant in greater distance from the cutting line (Fig. 3).

2.4. Flame

The original structure of measured sample was pearlitic, respectively ferrite-pearlitic, with hardness ranging from 160 HV to 165 HV. Cutting by flame creates the temperature of 3000 °C with accuracy of 0,5–1 mm. Environment temperature was 22 °C. The temperature difference between the cutting area and its surrounding environment is not high enough to have an impact on nearest environment around the cutting line. The cutting is followed by cooling process on open air, thus lowering the cooling speed. This enables the creation an equilibrium state of coarse-grained pearlitic structure with globular cementite. Baling of cementite is caused by the surface tension. This change in structure causes decrease in hardness to value of 165 HV as seen in Fig. 4.

2.5 Plasma

Plasmatic torch is a device for creating highly heated gaseous stream of ions and electrons (plasma). This jet stream is heated to very high temperature (over 30.000 °C) by arc discharge and it streams in high speed from a narrow jet. The plasma ray can be deflected by means of magnetic field and accelerated by means of electric field. The splitting of testing material was done by means of higher (30.000 °C) and lower temperature (20.000 °C).

Cutting by plasma creates the temperature about 30.000 °C in the cutting area and generator was used 100 A, Ar/H2. Environment temperature was 22 °C. Despite the great difference between the temperatures in the cutting area and the surrounding environment, no significant changes on the surface layer around the incision line have occurred. This could be caused by very low content of carbon (0,17 %) and by slower speed of cooling that has not reached the critical speed and thus a non-equilibrium state (hardening) was not reached. On the opposite side, the lower speed of cooling occurred causing the creation of very fine pearlitic structure with hardness around 255 HV. Micro-hardness values could be seen in Fig. 5.

![Fig. 3. Change of micro-hardness (cutting by laser)](image)

![Fig. 4. Change of micro-hardness (cutting by flame)](image)

![Fig. 5. Change of micro-hardness (cutting by plasma, 30 000 °C)](image)

![Fig. 6. Change of micro-hardness (plasma cutting, 20 000 °C)](image)
slightest speeding up of the cooling, the speed of cooling would reach the critical value and thus the process would create a non-equilibrium structure (bainit). By applying the lower temperature in the cutting line, the slight decrease of hardness occurred as could be seen in Fig. 6.

3. Conclusion

This work is focused mainly on issues connected with using of different methods of thermal material cutting and their impact to the surface layers. Three basic types of cutting were prepared. Laser cutting technology was used together with cutting by plasma and by flame. In cutting by plasma, two different cutting temperatures were applied, i.e. 30,000 °C and 20,000 °C. Testing samples were sanded and then prepared to micro-hardness testing. The testing was done on Micro-hardness tester DM-2D. The measurement results were displayed in graphic and then evaluated.

The results show that testing bodies cut by laser shown the hardness of 340 HV. Maximum thickness of the layer influenced by cutting was 1–1.5 mm (plasma and flame 3 mm). Deeper layers were not influenced by cutting procedure and the hardness remained unchanged. In plasmatic cutting, the tested material was influenced by the cutting procedure very slightly. The thickness of the influenced layer was almost the same as in laser cutting. The lowest micro-hardness values were shown in cutting by flame. The depth of influenced layer was more than three times greater than in cutting by laser or by plasma. This was caused especially by selected material containing only 0.17 % of carbon. Steels with such low carbon content are practically non-hardenable.

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REFERENCES


D. Maňas, M. Maňas, M. Staněk, S. Šanda, and V. Pata (Tomas Bata University in Zlin, Zlin, Czech Republic): Thermal Effects on Steels at Different Methods of Separation

The submitted article describes the effects of a thermal separation of material on the properties of the surface layers. The separation of material is an integral part of a preparation of all final products. During cutting the surface layer of the separated material is effected, which then has an influence on the sequence of the following operations. Three methods of thermal separation of material were selected for the experiment described in the article - a method of laser, plasma and flame cutting. The temperatures for plasma cutting were 30 000 °C and 20 000 °C. The material selected for these three methods was a common steel 11 373, which is industrially produced and processed. The effect on the material was shown by measuring micro-hardness using the micro-hardness tester DM 2D.