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Microstructural changes and strain hardening effects in abrasive contacts at different relative velocities and temperatures

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ABSTRACT

Strain hardening is commonly used to reach the full potential of materials and can be beneficial in tribological contacts. 2-body abrasive wear was simulated in a scratch test, aimed at strain hardening effects in various steels. Different working conditions were examined at various temperatures and velocities. Strain hardening effects and microstructural changes were analysed with high resolution scanning electron microscopy (HRSEM), electron backscatter diffraction (EBSD), micro hardness measurements and nanoindentation. Statistical analysing was performed quantifying the influence of different parameters on microstructures. Results show a crucial influence of temperature and velocity on the strain hardening in tribological contacts. Increased velocity leads to higher deformed microstructures and higher increased surface hardness at a lower depth of the deformed zones at all materials investigated. An optimised surface hardness can be achieved knowing the influence of velocity (strain rate) and temperature for a "tailor-made" surface hardening in tribological systems aimed at increased wear resistance.

1. Introduction

Strain hardening (or work hardening) is a widely used technique for the local change of mechanical properties of special materials. Changes in the microstructure due to an impinging load, e.g. within specialised production processes, cause strain hardening which can enhance the durability of materials [1–3]. Strain hardening raises the hardness and toughness of metallic materials due to deformations and is mostly gained at temperatures below the recrystallisation temperature. Strain hardening at higher temperature strongly depends on the stress levels, the dislocation movement and the temperature, although hardening mechanisms are less distinct at elevated temperatures [4,5]. Also, the strain rate, i.e. the deformation velocity cannot be neglected. Primarily, the microstructure and the induced or persistent inner stress level (residual stresses) affect the materials response [2,6]. Microstructural properties like present phases, their grain size, the degree of previous deformation, the type and amount of different grain boundaries present and the grain misorientation influence the strain hardening of materials [7–9]. On atomistic scale in polycrystalline microstructures several mechanisms can occur, strongly influenced by crystal lattice and dislocation movements [10]. Dislocation movements can be blocked by grain and sub-grain boundaries and at precipitations and inclusions. Inclusions can also develop more dislocations within a so called Frank-Read-source [10,11]. In general the dislocation density and resulting strain fields are key parameters influencing hardening of materials [12]. Also Lomer-Cotrell clouds can be evolved during hardening and can remain immobile and therefore increase the dislocation density [11,13].

Strain hardening within ferritic phases is mainly caused by decreased grain sizes and their influence on the mechanical strength according to the Hall-Petch equation [4,14]. This is strongly dependent on slip systems in polycrystals and their relation to dislocation motion, the formation of grain- and subgrain boundaries and its accompanying hardening effect [9,15–17]. Pearlite hardens via reduced interlamellar distances in grains [18]. The breaking and cracking of cementite lamellae also influence the interlamellar distance and decrease the interparticle distance, which entail hardening due to substructural restraint of dislocations [19,20]. Austenitic steels are likely to exploit good strain hardening behaviour [21]; the ability of the austenitic materials to increase their hardness strongly depends on the chemical composition [22] and the degree of deformation. The dislocation motion and substructure formation, including twinning and the formation of subgrain boundaries, are the main hardening mechanisms in austenitic steels [23–25]. Misorientation can be used as an indication for hardening in austenitic steels [26], since it is linked to the dislocation density and the degree of deformation [16,27]. Austenitic steels also are known to harden via the formation of martensite during deformation due to its higher thermodynamical stability [28,29]. This deformation induced martensite transformation is strongly dependent on the strain level, the governing stresses and the initial texture, which is given by the chemical compositions [30].

As presented, strain hardening and accompanying effects were widely studied, but deformation mechanisms and strain hardening

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effects under abrasive wear conditions lack of systematic investigation of temperature and relative velocity influence [31,32]. Therefore a novel high temperature scratch test was developed at the Austrian Centre of Competence for Tribology (AC2T research GmbH). This test rig enables the examination of fundamental abrasive wear phenomena at different load levels, velocities and temperatures up to 1000 °C [33–35]. Post-test HRSEM investigations, EBSD measurements and nanoindentations were performed, providing detailed insights in hardening mechanisms and microstructural changes at different phases. E.g. phase texture change, orientation and misorientation due to deformation, grain and sub-grain boundary formation can be detected [36].

The aim of this study is a detailed investigation of temperature and velocity influence on strain hardening of different steels and their microstructures. Different hardening mechanisms were pointed out at various microstructures and statistically analysed regarding their influence in different steels.

2. Experimental

2.1. Materials data

Three different steels were examined in this study giving insight to a spectrum of materials and microstructures and their strain hardening behaviour in tribologically affected zones. The chemical compositions of all materials investigated are given in Table 1. Material A is a ferritic-pearlitic carbon steel with 0.45 wt% C (1.1191 or AISI 1045) which was tested in normalised condition. Material B is a typical titanium stabilised, austenitic stainless steel with 18 wt% Cr and 12 wt% Ni, according to 1.4571 or AISI 316 Ti, in normalised condition. Material C is a cast steel (white cast iron) with high amounts of chromium (30 wt %) and carbon (1.4 wt%, 1.4777) in as-cast condition.

Table 1
Chemical composition of the materials investigated.

		Chemical composition [wt%]							
Material	Materials number	Fe	C	Si	Mn	Cr	Ni	Mo	Ti
A	1.1191	bal.	0.45	0.25	0.65	–	–		–
B	1.4571	bal.	0.08	–	1.80	18	30	2.50	0.30
C	1.4777	bal.	1.40	2.00	1.00	30	–	0.50	–

2.2. High temperature scratch test

The high temperature scratch test was designed to determine the scratch behaviour and hot hardness of materials at elevated temperatures [33–35]. The test rig is basically assembled of a 2-dimensional sample positioning system and a loading unit. This carries a resistance heated sample holder enabling testing up to 1000 °C. The setup is implemented in a vacuum chamber (<5 mbar) to prevent oxidation at elevated temperatures. Sample positioning and the relative movement between indenter and sample is realised in x-direction with a servo drive. Typical sample size is about 70 × 30 × 15 mm. The whole scratching process is fully automatised for optimal reproducible test sequences at set loads (up to 500 N) and temperatures. Various sliding velocities can be set from 10 mm/min up to 10 mm/s in x-direction. For this study the normal force was set to 100 N and a sliding velocities of 1 mm/s and 0.16 mm/s (10 mm/min) were chosen to show the influence of different strain rates on the microstructural changes due to deformation. All materials were tested at room temperature (RT) and 500 °C to investigate temperature dependence of wear behaviour.

To determine the hardness decrease due to elevated temperatures hot hardness tests (HHT) were performed using the Vickers method with 10 kg (98.1 N) load at several temperature levels (RT, 100, 300, 500 °C) within the same test rig. Evaluation of indents was done afterwards by optical microscopy (OM).

After scratch testing wear scars were analysed with HRSEM to get an overview of present wear mechanisms. A more detailed description of the scratch test and wear mechanisms present at the given materials can be found in previous works by the authors [33–35].

2.3. Analyses of the deformed zones within metallographic cross-sections

Metallographic cross-sections of the scratches, i.e. deformed zones were prepared after testing. Specimens at all tested parameters were cut, embedded in conductive plastics, grinded and polished up to 1 μm diamond particle size. On the cross-sections changes in the microstructure were investigated by HRSEM. For this microscopical analysis Material A was etched with 3 % HNO_3 in ethanolic solution; Material B with a mixture of 40 % HCl and 10 % HNO_3 in aqueous solution and Material C was etched with 10 % $FeCl_3$ in a 10 % HCl solution. To determine the hardness increase and further the mechanical properties of each phase present, nanoindentation measurements were performed with a Hysitron® Triboindenter TI900 equipped with a diamond Berkovich indentation tip (tip radius 100 nm). Loading cycles were set to 5 s to reach the peak load of 10 mN and subsequent unloading for 5 s. The load vs. depth curves were analysed to evaluate the elastic modulus and hardness using the procedure described in [38]. Indents of linescans were performed every 6 μm to a depth of 100 μm from the surface (15 indents). For statistical reasons 3 lines were done. The different phases were analysed separately to figure out hardening behaviour of specific phases. These indents were set approximately 5 μm from the scratch surface in different deformed microstructural phases.

Additionally, a statistical analysis of nanoindentation data and its correlation to different microstructural parameters like grain size, interlamellar spacing, grain misorientation and martensite content was performed.

2.4. Electron backscatter diffraction (EBSD)

For a detailed investigation on the ongoing microstructural deformation and hardening mechanisms during fundamental abrasive contacts EBSD-measurements were performed. In order to identify the influence of the temperature, samples with 0.16 mm/s scratch velocity tested at room temperature (RT) and 500 °C were chosen for investigation. To show the influence of the scratch velocity RT samples with 1 mm/s scratch velocity were analysed as well. Within this study EBSD measurements were performed with an Ametek-EDAX Electron backscatter detector integrated in the HRSEM Zeiss® Supra 40 VP. The preparation of the corresponding cross-sections included conventional grinding and polishing up to 1 μm. To ensure a sufficient surface quality fine polishing with a Struers® oxide polishing suspension was performed.

Kikuchi-Patterns have been acquired at 20 kV acceleration voltage and a specimen tilt angle of 70° using the EDAX™ TSL OIM Data Collection Software. For all scans a clean-up procedure including grain dilation with minimum grain size of 5 pixel has been applied within the software. Scans of the deformed zones were recorded from the scratch traces on the surface towards bulk. Measurements of the unaffected microstructure were taken in the middle of the samples without any load influence. In order to obtain detailed measurements from the deformed zones and the initial material condition high resolution EBSD scans with a step of 100 nm within an area of 50 × 100 μm (magnification 3000) have been acquired. Evaluations were performed via image quality (IQ) mappings with grain boundaries, inverse pole figure (IPF) maps, phase analyses and the kernel average misorientation (KAM) profiles.

3. Results

3.1. Microstructural analysis and hot hardness of the materials investigated

Microstructures of all materials investigated are given in detail at the HRSEM and EBSD results (Sections 3.4–3.5). Here a short overview should be given. Material A is a ferritic-pearlitic carbon steel with about 45 % ferrite and 55 % pearlite. The average grain size of this steel is 2–20 μm. Material B reveals typical austenitic microstructure with twin grain