TRANSPORT
ISSN 1648-4142 print / ISSN 1648-3480 online
2012 Volume 27(2): 196–205
doi:10.3846/16484142.2012.696214

ON WHEEL–RAIL CONTACT SURFACE PHENOMENA WITH STRUCTURAL CHANGES AND 'WHITE ETCHING LAYERS' GENERATION

Libor Beneš

Dept of Materials Engineering, Czech Technical University in Prague, Karlovo náměstí 13, 12135 Praha, Czech Republic
E-mail: libor.benes@fs.cvut.cz

Submitted 27 April 2011; accepted 21 June 2011

Abstract. The main aim of this work was a study of the microstructure transformations with the residual stress formation that is induced by rolling contact friction and adhesive wore in the wheel–rail system. Several small rail-surface samples, we term them as the 'chips', and a piece of wheel sample were chosen for the analyses of the surface changes on the wheel–rail surface. A multitude of different experiments were carried out in order to analyse the microstructure changes at the surface and the near-surface region of the material samples and, thus, to contribute to the understanding of the complex wheel–rail rolling contact phenomena – and its degradation mechanisms. The formation of nano-structured martensite and carbides on the rail and wheel surface causes the extremely high microhardness valuees and the strong corrosion resistance of the so called White Etching Layers (WEL).

Keywords: rail, railway wheel, rolling contact, surface, microstructure, transformation, white etching layer (WEL), wear, adhesion, degradation mechanisms.

1. Introduction

Increased service life and reliability of rail materials and steel for manufacturing of rail wheels is a highly topical issue, considering the current trend of increasing the number and modernisation of rolling stock, and increasing axle loads and travel speeds on railroads. Added to this is the situation of the Czech Republic ascending to the highly developed European environment, which is closely related to building high-speed railroad corridors which run across the country and represent the main transport routes among a number of European countries. The geographic and geopolitical position of the Czech Republic calls for the need to join the transportation infrastructure to the European network in such an extent and quality that will allow operation of domestic and transit transportation at an internationally recognised level of quality.

In the conditions of the Czech Republic, road and rail transport continue to play the key role as they will serve the majority of transportation needs that the society will have. From this point of view, it is the reliability translated into service life and safety of individual components in the vehicle–rail system that represents one of the prime factors that affect the function of an advanced rail transportation system.

The present era and its matching level of scientific knowledge and technical expertise undoubtedly require that a systemic approach be applied towards the analysis and realisation of these facts. A systemic philosophy and theory are required for a structural description of the characteristic properties which, by means of components and relationships in the wheel–rail system, result in certain processes and state in the system of interest (Burša, Janíček 2009).

As such, the experimental part of the study presented herein focused on the study of structural changes in the surface layers of the rails, with a primary emphasis of the mechanisms of occurrence and the structural characteristics of the 'White Etching Layer'. The methods of X-ray diffraction analysis, light microscopy and transmission electron microscopy (TEM) were used to analyse segments from rails deteriorated by operation. The segments were collected as they exhibit specific surface defects with the possibility of a 'White Etching Layer' (hereinafter referred to as WEL).

2. Surface Layer Degradation in the Wheel–Rail Contact

Research of degradation processes in the surface layers of the contact areas is closely related to the study

Copyright © 2012 Vilnius Gediminas Technical University (VGTU) Press Technika
http://www.tandfonline.com/TRAN

of mechanisms of structural modification in those locations. Research activities have been focusing on this issue on a rather major scale; however, the processes have not been clarified entirely which during an operation of the rails lead to structure modifications in the surface or subsurface area of wheel and rail contact. Another important consideration in the matter is the effort towards reduction of the financial burden that includes the costs for manufacturing and operation of rail vehicles and the tracks, as well as the requirements for maintenance of the vehicles and rail superstructure.

The results of our own research and study of literature on the subject suggest that the area of the wheel and rail contact, being exposed to operation and long-term variable load and wear, exhibits major macro- and microstructure changes. These are especially related to occurrence of corrugation on the rail head (so-called as *riffels*) and to occurrence of so-called WEL on the corrugated rail surface. As WEL is characterised by a high level of hardness and, therefore, by notable fragility, these layers represent a critical location for occurrence of cracks or release of a part of the material separated by cracks, known as *spalling*. A system thus degraded generates considerable vibration and background noise; this is incompatible with the European concept of a 'silent railroad' that is currently subject to frequent discussion. Furthermore, periodical grinding of damaged rails is associated with considerable expense.

In general, the structure components that are the subject to varied contact load are the subject to degradation mechanisms known as *contact fatigue*. This refers to damage the evidence of which are cracks that occur either on or below the surface, depending on the nature of stress. The topic of contact fatigue is rather complicated as it always includes rotation of the major stresses (Wang 2002). What is more, there are many other mechanisms (e.g. wear, degradation by corrosion, corrugation of the surface) each of which needs to be considered in prediction of the *fatigue life* of the entire system.

A typical example of a contacting pair matching this concept is the wheel–rail system. As axle loads and travel speeds in the railroad industry are constantly rising and, in the global view, all railroad administration bodies are striving for more effective methods to prevent the wear of the rail head and the wheel, the issue of *rolling contact fatigue* (RCF) as in the wheel–rail system represents a major concern, see Tournay and Mulder (1996) or Clayton (1996). The subject of these analyses is quite complex and many of the aspects have not yet been theoretically explained in full.

The rail head is the rail component which is exposed to the highest load and wear. Higher operating load is caused especially by the dynamic forces between the wheel and the rail; by heat effects of wheel slip both while braking or transmitting power from the axle to the rail. These factors are subsequently multiplied by the ever-increasing weight load per axle and by the continuous application of higher and high speeds. It is evident that the observed damage of the rail head (*fatigue cracks*) may be initiated both on the surface and in the areas immediately below the surface, see Galliera *et al.* (1995). It may also be inferred that the mechanisms leading to occurrence of these are highly varied and depend on the operational, material, and geometrical conditions of the contact.

In consideration of movement of the heat source in the wheel–rail contact, the following may be inferred: The observed defects identified by the ČD (Czech Railways) regulation S-67 in force (Předpis SŽDC (ČD) S67 1997), type: 'areas in the zone of travel ground by recurring slippage of the driving axle' (referred to as *wheel-burns* in English terminology), occur at low speeds of vehicle travel relative to the rail.

An analogous situation, albeit in the reverse order, is the travel mode when the wheel is blocked and is sliding on the rail, with the generated heat exerting stress rather on the wheel than on the rail. In order to explain both these phenomena it is necessary to monitor the relative speed of the wheel relative to the rail, in the area of contact. That of the two monitored surfaces which is moving at a higher rate of speed produces friction heat that is dissipated equally into the rail and into the wheel. It is obvious that the heat source remains longer above the surface that is moving more slowly (in the case of a blocked wheel it is the wheel itself, in the event of slippage, it is the rail). The result is that the surface moving at the lower rate of speed absorbs much more heat that is generated at the point of contact, with the risk of exceeding the transformation temperature of the material. However, once the heat source moves outside the area of contact, the subject location of the surface layer is abruptly cooled by subsequent radiation of heat into the cold mass that surrounds the monitored area of contact.

This translates into structure modifications which result in degradation of the layer on or below the surface on the rail head or on the active area of the wheel. The mechanism described above leads to fragile structure elements, cracks, crumbles, or cavities surrounded by cracks and other types of damage in the area of contact. Given the wide scale of modifications of the heat source generated by the contact of the wheel and the rail, there is an existence of a range of external signs of damage (i.e. *wheel-burn* defects) on the rail head – isolated, multiple, continuous within a section of the rail, etc.

3. White Etching Layers on the Rail Head

The requirements for material characteristics of rail steel, caused by the increasing rate of travel speeds of rail vehicles and the higher axle loads, are constantly rising. The changes of material structure of the rail retroactively affect the mechanisms of rail degradation. While in the past two decades the main criterion for rail service life has been its resistance to abrasive wear (application of Series 75 ČD rails in the Czech Republic), the current prevailing trend is the requirement for resistance to occurrence and propagation of contact-wear defects (i.e. defects such as *head-checking, riffel* and *shelling*) that occur on or near the surface of rails with higher strength characteristics.

These defects differ in the mechanism of occurrence, yet one of their properties is universal – occurrence and propagation of cracks with subsequent separation (flaking, release) of material from the rail surface which, in an extreme situation, may lead to rupture across the entire rail section. Each year, considerable amounts of funds are spent in the world for elimination of these defects and of their consequences. Therefore, explanation of the acting mechanisms and effective factors represents a crucial task towards reduction of occurrence of the contact-fatigue defects in rails mentioned above.

When viewed in a light microscope, the cross-section of an operated rail (Fig. 1) shows *riffle bands*, (surface corrugation of the rail head) in a layer approximately 10÷100 μm thick, without a structural contrast, which is resistant to etching of metallographic samples and shows a white colour after etching with Nital. This layer is often considered or identified as martensite. As this layer exhibits exceptional fragility (referred to as WEL since Chapter 1, the abbreviation derives from the English term 'White Etching Layer'), propagation of defects known as *squat* or *shelling* and occurrence of defects is observed in the rail head. All of the above result in background noise when the wheel travels along the rail and exhibit a detrimental effect on the service life of the rail and of the wheel axle (Lojkowski *et al.* 2001).

Therefore, a number of studies have focused on the origin of these WEL (Clayton 1996; Galliera *et al.* 1995). Until recently, the predominant theory claimed that WEL is caused by the rise of temperature in the wheel–rail contact during enormous friction, or slippage, processes during wheel travel along the rail (Baumann *et al.* 1996). Standard rail steel has a predominantly pearlitic structure with a nearly eutectoid content of carbon and fine plates of pearlitic cementite. Some authors claim that the temperature in the wheel–rail area of contact may extend the boundaries of transformation temperature for generation of austenite (Kout 2001, Mitura 1980) while the subsequent rapid cooling may be the cause for modification into martensite (Jergéus 1992) in a closely delimited area corresponding to the surface.

Although this may, in extreme situations, provide conditions for generation of martensite on the rail surface (heating to the transformation temperature and subsequent rapid cooling), it has been established that this is not the only possible mechanism of WEL occurrence (Djahanbakhsh *et al.* 2001). This discussion was initiated by Newcomb and Stobbs (1984) who used TEM surface analyses of rail surfaces and suggested the so-called model of low-temperature solution of carbides. These authors consider the influence of high contact pressure in the wheel–rail system in the order of GPa, which causes solution of carbides in temperatures as low as 100°C; this process leads to much higher density of dislocations and to segregation of carbon atoms in the cores of such dislocations. Furthermore, they suggest a process of progressive propagation of WEL that differs from the initially accepted single-stage process of WEL occurrence.

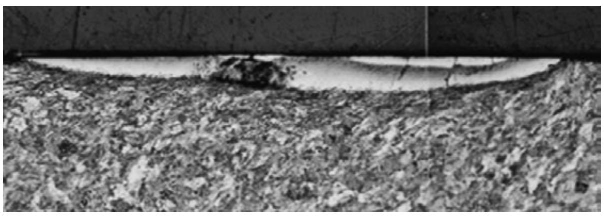

Fig. 1. White Etching Layer (WEL) on the surface of the rail (Wang 2002) – Nital, magn. 100x

This concept has recently received considerable support, especially based on recent findings on the so-called mechanical alloying process (Lojkowski *et al.* 2001). These findings have established that experimental processes may lead to generation of structures analogous to the discussed WEL only by mechanical spreading of metallic dust from pearlitic steel under high pressure at temperatures lower than the transformation temperature required for generation of austenite (Djahanbakhsh *et al.* 2001).

At the same time, it has been found that considerable amounts of carbon dissolve not in the dislocation cores, but in the surrounding ferrite, as documented by modifications of the lattice parameters. This effect may be related to the diffusion processes in the total mass of the material. The theory of mechanical alloying, which is based on low-temperature and purely mechanical alloying is known as the *ballistic theory* (Djahanbakhsh *et al.* 2001), and has been selected as a description of the process.

Baumann (Baumann *et al.* 1996) does not regard the white etching layers to be 'traditional' martensite either, in reference to Liebelt and Knotthe (1997) who suggests that the maximum temperature limitation of the processes occurring in the wheel–rail contact is temperature increment by 400 K; this value is not sufficient for austenitisation of pearlitic material and for its subsequent quenching into the martensite region.

For a long time, WEL were regarded as amorphous materials in the development of laboratory equipment at the time did not allow identification of their microstructure. The structure is similar to that of the Beilby layer that is used in polishing and grinding of polished sections in metallography. Based on the high hardness and fragility of the material, studies of WEL have observed the occurrence of cracks and crumbling of material from the rail surface with subsequent generation of dappling in the rail head area (Baumann 1998).

4. Research Methodology

The goal of the experimental section of this study was to analyse microstructure changes in the rail head caused by wheels contacting the rail especially in locations with frequent starts of rail driving machinery and on track curves. As indicated in the introduction of this paper, it is an extremely complex tribology system in which the friction and wear processes lead to degradation of material and cause occurrence of critical defects.

The results of the literature search infer that the rail area on and below the surface is exposed to considerable loads which lead to major changes in the material structure and result in generation of major structure gradient in the direction of the normal line towards the surface.

Therefore, special attention is given especially to take-off sections of rails where considerable damage of the travel surface of the rail has been caused by cumulative slippage of wheels. The sequence of processes induced by temperature and deformation was used especially for monitoring the formation of the WEL on the rail head, which is specified by high levels of hardness and resistance to etching of samples.

Samples collected from rail sections where occurrence of WEL was assumed (*samples taken from take-off track sections before semaphores or from areas around train stations*) were subject to metallographic analysis, X-ray diffraction analysis, and to studies using transmission electron microscope (TEM).

In order to provide comparison of the degradation mechanisms being analysed, the selection of experiment material was extended by segments from Mn-austenitic steel used for the crossing elements of rail switches, and by samples taken from the travel surface of the rail wheel rim.

After demonstration of the existence of the WEL that is clearly connected with the initiation and propagation of defects on the rail head, a hypothesis was proposed and subsequently verified for elimination of WEL by cumulative tempering under air at temperatures of 150°C, 200°C, 250°C and 300°C, each for a period of 4 hours. Classification of the structure changes after each of the four-hour cycles required that the samples be documented using the light microscope, SEM, and a diffraction plot to record the progressive change in the microstructure characteristics of the studied sample volume.

The material analyses were complemented with a stress-deformation study of the situation in the wheel–rail contact in the conditions of rail vehicle take-off with a defined amount of wheel slippage on the rail, with a focus on effects on the dominant components of stress by wheel slipping and also in respect to heat stress and the amount of heat effect on the rail surface under take-off with wheel slippage.

5. Metallographic Analysis of Rail Chips

The purpose of partial experiment (metallographic analysis) was to use a standard polished section – a cross-section of the selected chip – for assessment of the microstructure changes observed after application of low-temperature tempering in comparison to the initial condition of the chip.

The selected chip sample consists of heterogeneous layers that are the result of intensive and locally varied plastic deformation with a dominant influence of tangential shear components in the area of contact. Mutual dislocations of the layers are delimited by thin oxide layers in the boundary regions and, in some cases, by pores arranged in rows or by more consistent cracks. The potential defects possess a high degree of variability; this is logically attributable to the random nature of the observed states and their random dynamic causes.

The analyses performed show that the WEL itself consists of several sub-layers; this suggests gradual generation of the layer during operational exposure (see Fig. 2). The figure also shows a tendency to layer delamination along the boundary with unaffected mass in the sub-surface area of the rail head, which is the result of a high hardness gradient and, therefore, of local stress conditions acting on both sides of the boundary area. The cumulative effect of the tangential shearing components in the area of contact causes separation of material masses from the rail head and is the cause of occurrence of the studied defects on the surface.

The selected chip and the documented layer showed a characteristic feature in the locally homogeneous microstructure which may be used to derive one of the possible mechanisms. High local dynamic stress and the resulting plastic deformation have caused, in the final effect, local heating up to the transformation levels $\alpha Fe + M_3C \Rightarrow \gamma Fe_C$. The maximum temperature was high enough (it is necessary to consider transformation hysteresis at the realistic rate of heating, including the possibility of M_3C dissolving in austenite).

The subsequent abrupt cooling of the austenite saturated with carbon resulted in athermal martensitic transformation. The quantity of carbon dissolved in austenite (approximately <0.3% up to >0.5%) was sufficient for transformation of lath martensite (a habitual parameter close to {225} is typically reported). The microstructure shows the typical lath arrangement of the triangular elements.

The registered condition of the initial microstructure suggested that this case was the final condition of the chip as the martensitic structure mentioned above

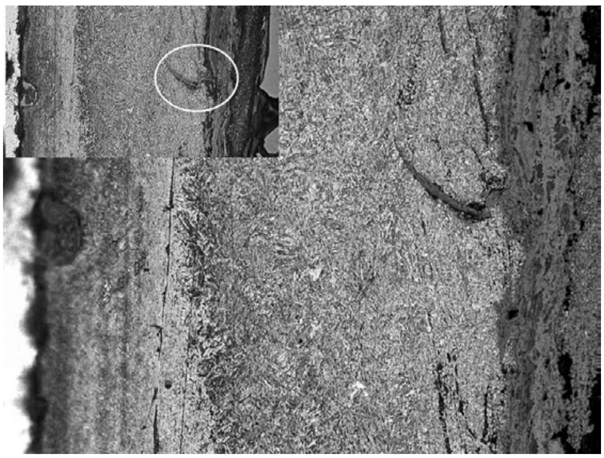

Fig. 2. Microstructure of the studied rail-segment after annealing on 250°C / 4hrs – etched in NITAL, magn. 250x and 50x, respectively

showed no subsequent plastic deformation. The state of an identical area has been documented in various cumulated modes of tempering for an assessment of the effect of secondary application of the heat in temperatures ranging from 150 to 300°C to the initial structure.

Tempering of lath martensite at 150°C usually does not yield recognisable changes in the microstructure under application of optical microscopy equipment. As the tempering temperature increases, the characteristic triangular arrangement of martensite laths gradually loosens and nearly disappears after heating at 300°C. It is replaced with a structure which is known as 'sorbite' in carbon steels. It is a mixed structure of ferrite and carbides which have precipitated from the lath martensite, initially in the form of scales or, later (at higher temperatures), in the shape of more or less perfectly coagulated particles.

Table 1 shows the microhardness value of HVm 50 which is determined by puncture into the influenced side of the chip before and after annealing at 300°C/4 hours.

Table 1. Microhardness values HVm 50 measured in the surface layer of the influenced rail-chip, before and after annealing 300°C / 4 hrs

	Surface layer with the 'nanostructure' component	
	Measurement average 1	Measurement average 2
Before annealing (initial state)	721±13	841±14
After annealing 300°C / 4 hours	405±11	364±12

After etching, the overall character of the sorbitic structure exhibits relief both by nature as per the initial orientation of the martensite laths and secondarily by the orientation and local clusters of the carbidic precipitate. The described decomposition processes of lath martensite correspond to the measured microhardness in the respective areas and, to a large degree, to the results of the X-ray diffraction analysis.

6. TEM Analysis of a Sample Taken From the Surface of Rail Steel

Analyses in a transmission electron microscope (*hereinafter referred to as TEM*) were performed in two samples of chips from the subject rail: the samples were identified as 'initial state' and 'state after tempering at 300°C/4 hours'. The following methods were applied in preparation of thin films from the segments taken from the rail surface for TEM observation: the discs were spherically ground on the inside using the *One Side Dimpling* method; then they were thinned with ion polishing in a Gatan PIPS (*Precision Ion Polishing System*) machine. The thinning operation was performed, for the most part, on the inside so that the distance of the studied area from the rail surface was approximately 1 (m. The subsequent observation used a Philips CM12 transmission electron microscope under 120 kV of acceleration voltage.

The sample was a typical microstructure of highly deformed steel with a high density of dislocations. Selective electron diffraction from an area with a diameter of approximately 6 μm shows uninterrupted rings, which suggests a very small grain size. Grain size ranges approximately from 20 to 200 nm. Therefore, the microstructure may be identified as nanocrystalline (see Fig. 3), highly homogeneous; the grains are of irregular shape and are separated by clusters of dislocations.

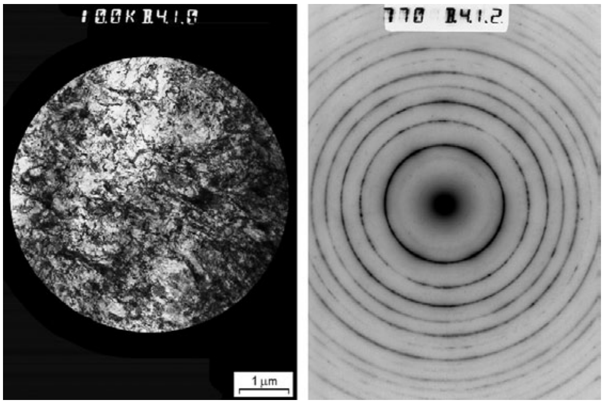

Fig. 3. TEM-image of the region where the selected electron diffraction was carried out; the obtained annular diffraction pattern proved that the microstructure consists of very small ferrite grains with random orientation

Annular diffractions (see Fig. 3) show that only ferrite is present in the material; presence of carbides may be expected; however, the diffraction sensitivity in the situation is not sufficient as to show carbides in the image.

Presence of austenite has not been demonstrated. The deformation of the structure is so strong that the carbide contrast in normal images is obstructed by the contrast from a dislocation cluster.

The density of dislocations cannot be quantified in such a highly deformed state – compared to the publication by Lojkowski *et al.* (2001) where an analogous case showed a numerical representation of dislocation density; however, it was only shown in samples collected from the rail surface at a distance of more than 50 μm from the rail head.

After annealing at 300°C / 4 hours, materials recovery occurs, the density of dislocations declines and the boundaries of individual grains are defined more clearly. Selective diffraction suggests that the quantity of grains per weight unit has reduced considerably.

7. X-Ray Diffraction Analyses

All X-ray diffraction analyses were performed in identical conditions using a Siemens D-500 ray diffractometer, filtered by Fe Ka1+2 radiation in a range of 50÷150° 2Q at stepping of 0.025° 2Q with 20 seconds per step. The analysis of present phases was performed by comparison with the PDF 2 commercial database of X-ray diffraction plots using the SEARCH software by Siemens.

The software application of DIGITAL used the method of approximation of a modified Lorentz formula (‘*Cauchy quadrate*’) in the obtained diffraction plot to calculate the following parameters: intensity [pulses], full width at half maximum (FWHM), and the position of the peak maximum value. By application of Bragg’s law:

$$n\lambda = 2d\sin\Theta, \tag{1}$$

where: n (an integer) is the ‘order’ of reflection; λ is the wavelength of the incident X-rays; d is the interplanar spacing of the crystal and Θ is the angle of incidence, those values were used to calculate the planar distance d_{hkl} of individual planes in each sample and to calculate the lattice parameter a_{hkl} using the formula:

$$\frac{1}{{d_{hkl}}^2} = \frac{(h^2 + k^2 + l^2)}{{a_{hkl}}^2}, \tag{2}$$

where: h, k, l represent the Miller indexes of the specific plane.

Using the inverse proportion, the full width at half maximum (FWHM) may be used for calculation of size of coherent domains L_{hkl}, where diffraction occurs. This parameter may be compared to ‘X-ray’ grain size of sorts and, using the Scherrer equation, the parameter equals to:

$$L_{hkl} = \frac{\lambda}{\text{FWHM}\cos\Theta}, \tag{3}$$

where: FWHM must be stated in radians.

The non-influenced side (see Fig. 4) shows no major change in any of the diffraction line parameters; therefore, the non-influenced side of the chip seems unaffected by cumulative tempering at temperatures below 300°C.

Comparison of the achieved results clearly indicates that a thin layer of Fe oxides was formed on the non-influenced side as a result of tempering. Interestingly, such oxidation products were not identified in the influenced side in any temperature. There is a highly plausible explanation that the presence of the ‘nanostructure’ component in the rail surface increases its resistance to corrosion.

The performed X-ray diffraction analyses and the measured results offer the main results given below.

Presence of residual austenite was not found in any of the chips, not even under annealing up to 300°C. Presence of martensitic doublet in the observed chips was not proven; however, this fact must not be regarded as a rule. The only finding is that the X-ray diffraction method did not show presence of any martensitic structure in the chip samples. The ‘nanostructure’ component which is present on the chip surface disappears at temperatures between 250 and 300°C. However, this temperature does not yield any major changes on the microdistortion level or in the grain size of the polycrystalline component.

Microhardness HVm 50, identified in the section across the chip in the initial state and after annealing 300°C/4 hours, has proven (see Table 1) that general recrystallisation or recovery of the surface microscopic layer with the ‘nanostructure’ component of bcc-Fe had occurred before this temperature was achieved.

8. Occurrence of a White Etched Layer (WEL) in the Wheel–Rail System

The issue of occurrence of contact fatigue in the thrust bearing is analogous to the wheel–rail contact. In spite of this fact, there exist only a few studies on the topic of degradation of surface areas and occurrence of WEL in the wheel–rail system. In this aspect, Baumann (1998) and others have concluded that WEL in the rail consists of a nanocrystalline structure with grain size of 20 nm.

Baumann (1998) suggests that this nanocrystalline structure is the principal fact that prevents WEL from being studied using a light microscope; he applies the same view to explain the exceptional resistance of this layer to corrosion. As the X-ray diffraction technique showed no reflections attributable to the carbidic component, Baumann assumes that the carbides have dissolved in a certain mass of pearlite. This suggested

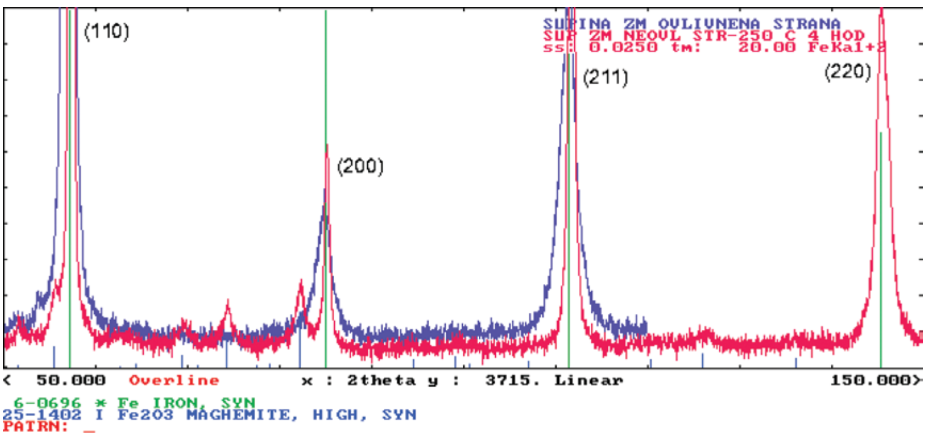

Fig. 4. X-ray diffraction plot made on the non-influenced side of the rail-chip after the next annealing 250°C/4 hrs (including comparison with the influenced side)

mechanism has been confirmed with an experiment that consisted of 70-hour crushing of a mixture of Fe_3C and Fe; the test demonstrably showed disappearance of the Fe_3C compound. An identical result was later confirmed with a thermal analysis which showed that precipitation of carbides occurred by heating of crushed nanocrystalline material.

Baumann (1998) also holds that an extremely high level of deformation and shearing stress leads to dissolution of cementite and diffusion of carbon. Using the Scanning Electron Microscopy, Baumann (1998) had demonstrated that WEL possess their own substructure; therefore, they consist of several sub-layers, indicating that WEL are generated progressively and that the structure of their individual segments may vary. Later (Lojkowski *et al.* 2001) it was shown that pearlitic colonies immediately below the rail surface are dissolved by the said mechanism; this indicates a high degree of plastic deformation which assists in generation of the nanostructure component from ferrous material that contains a small amount of cementite. Partial transformation of the nanocrystalline structure into martensite was also observed.

Other systems which operate on the principle of friction between the contacting faces have previously confirmed that slippage between metallic surfaces may produce a nanocrystalline layer on the surface of either friction faces of the system components. The thickness of this layer is then dependent on the type and material of the friction pair and on the slippage conditions. Typically, the layer is 1 μm thick (Ganapathi *et al.* 1990). Tensile tests have established that high levels of strength and fragility are typical in nanocrystalline materials (Kuznetsov *et al.* 1963).

The obtained results clearly indicate that samples collected from railroad sections operated under high load or from sections operated under lower load levels show notable effects on material microstructure below the rail head – depending on the distance from the surface; this corresponds to load distribution in the wheel–rail contact (Beneš, Schmidová 2003a, 2003b).

Wang (2002) maintains that occurrence of white etched layers was recorded in all locations with notable growth of residual pressure stress. As residual tensile stress was identified in the rail head and bottom as early as in the rail manufacturing process, the author focused on the analysis of residual tensile stress in relation to the observed enormous, permanent plastic deformation and generation of martensite on the rail head during the use. His findings indicate that WEL, a notable change in structure, is caused by high cyclic load and the cyclic nature of wear in the contact between the wheel and the rail. This aspect must not neglect the vibration generated in the wheel–rail system (Lata, Čáp 2010).

9. Microstructure of the White Etched Layer

Application of light and electron microscopy equipment has confirmed that the transition between the standard ferritic-pearlitic structure of the rail material and the white etched layer is sharp at all times. No presence of any transition layer has been identified. This fact only confirms that the plastic deformation during WEL generation is complemented with a thermal load (Liebelt, Knotthe 1997).

Should the only mechanism of this transformation process (transition from a ferritic-pearlitic microstructure into a white etched layer) be enormous plastic deformation, the gradient of the deformation substructure would be observed in the top surface layer, such as in railroad wheels which show deformation of the active surface without any occurrence of WEL.

The observed, extremely high level of microhardness (approximately 800 HVm) in the white etched layer indicates that WEL cannot consist of 'traditional' martensite as its microhardness, adequates to a steel with such carbon content, would be expected to be somewhere around 500 HVm.

The observed changes in structure suggest generation of heat flow (resulting from recurring wheel slip on the rail) and subsequent austenitisation in the area of contact. This rapid rise of temperature is replaced by abrupt cooling by the relatively colder mass in the adjacent sections of the rail, which may lead to generation of martensite in close proximity to the surface. It is a thermally induced transformation the procedure of which is: *ferrite/pearlite → austenite → martensite*.

It is a general concept that the boundary temperature for austenitisation of non-alloyed steel is 723°C. However, in reality, this temperature limit only represents the theoretical austenitisation temperature for a balanced phase transformation under typical pressure levels. The rate of heating, pressure, the alloy elements, and other parameters affect the transformation temperature greatly. Some authors (Bever 1986) state that a higher rate of heating increases the temperatures of A_1 and A_3. The effect of hydrostatic pressure on changing the balanced parameters of A_1, A_3, and the temperature of M_s has been confirmed.

High pressure levels cause many physical and mechanical properties of material to rise to abnormal levels (Ortner *et al.* 1992; Ermolaev 2000). For example, authors states that as the hydrostatic pressure rises, the temperature of beginning of the martensitic transformation drops on a nearly linear scale; however once pressure is relaxed, the parameter of temperature M_s assumes its normal value. The authors – Hellier *et al.* (1986), Bain (1924) – claim the value of nominal contact stress on the rail surface to be nearly that of the hydrostatic pressure, over 1400 MPa. It is the hydrostatic pressure which may have a major effect in reducing the transformation temperature A_3 within the area of contact, see Eckstein (1971) as an example.

The 'hot spots' on the surface during rolling (in the conditions of *dry rolling*) may achieve very high heat flow – Quinn and Winer (1987) indicates temperatures in excess of 900°C. Therefore, assumptions may be derived in saying that the austenitisation process commences in the most exposed locations of the wheel–rail contact which also show extreme levels of material wear. As both the wheel and rail exhibit considerable weight, the rate of cooling is equally high, as is the opportunity for occurrence of martensite.

Due to the rapid dissipation of heat into the surrounding mass of the rail, the specified maximum temperatures may only propagate in a very thin layer below the surface; therefore, the observed white etched layers are not located in depths exceeding several tens of micrometres. Wang (2002) mentioned a distance of approximately 16 μm from the rail surface where residual austenite, or asymmetrical distortions of reflex profiles can be identified.

The conditions of thus induced austenitisation clearly do not offer a sufficient time window for processes initiated by diffusion to completely dissolve the cementite, causing locally varied content of carbon in the austenite. This is shown by an analysis of carbon dissolved in martensite (Wang 2002) on the surface of the samples or cementite particle residues observed in TEM. It has been well established that martensitic transformation is a diffusion-free transformation process so the carbon content in martensite corresponds to the carbon content in austenite; therefore, there is no change in position of the carbon atoms (Bever 1986; Bain 1924). It is generally valid that during the martensite transformation, the dissolved carbon causes distortion of the lattice and causes the occurrence of a tetragonal bcc-structure the c_m/a_m ratio of which is proportional to the carbon content. However, experimental determination of martensite tetragonality in the observed rail segments is extremely difficult as the analysed, highly fragmented chips cannot meet the prerequisite of a homogeneous sample.

It is the increment of this distortion with higher content of carbon, generated by dissolution of cementite, that leads to rapid increase of hardness (Baumann 1998). Wang (2002) supports the increased quantity of carbon dissolved in various stages of *rail corrugation* by the change in the martensite ratio c_m/a_m as seen by application of the so-called synchrotron radiation method; the said c_m/a_m ratio shows notable changes across the deformed band (*riffle band*) on the rail surface.

10. Nanostructure and Stabilisation in the White Etched Layer

The obtained results of comparison of reflection widths recorded in the WEL analysis suggest that domain size is strongly reduced by high level of deformation induced by operation. This fact is also brought up by Wang (2002) with assuming that the effects of cyclical deformation and wear mechanisms caused by passage of rail vehicles may lead to gradual transformation. This transformation occurs in the cementitic lamellas in the pearlite of the original rail or wheel into a mixture of coarse deformed particles which the effects of enormous deformation further divide into ultra-fine grain structures.

It has been proved previously that slippage in contacting metals may lead to, in one or both contacting surfaces, occurrence of a nanocrystalline layer. The thickness of this layer is highly dependent on the materials of the contacting pair and on the conditions of friction (Ganapathi *et al.* 1990). Kuznetsov *et al.* (1963) observed total dissolution of cementite and reduction of grain size up to 20 nm in steel containing 1.2% of carbon; the material had been exposed to high torsional deformation. Valiev (1997) even prepared several pure metals and alloys with nano- and sub-micrometre grain sizes using techniques based on application of enormous plastic deformation.

Baumann (1998) also notices transformation of pearlitic structure and reduction of grain size in the order of nanometres, which is caused by high plastic deformation in the dapples in a thin layer of the rail surface. The authors of the study also simulated occurrence of a similar situation in a ball crusher with so-called mechanical alloy addition; however, they were not searching for correlation of the obtained results and the mechanisms of martensite generation.

A specific parameter of nanostructure (Wang 2002) is the ratio between the surface (boundary) and the mass. Mechanical properties of solid materials are highly affected by the density of dislocations, grain size, and the relationship between the boundary and the mass of the phase in question. Therefore, in the case addressed, the identified nanostructure bears a part of the responsibility for the high level of microhardness in the white etched layer and for its high resistance to corrosion. The same reason allows a light microscope to observe only a solid homogeneous structure of the layer, free of any contrast.

The principle of stabilisation based on refinement of grain size is the fact that grain boundaries act as barriers to movement of dislocations. Any dislocation oriented towards the adjacent grain must change its direction of travel. On the other hand, defects in the atomic arrangement on the grain boundary result in discontinuity of slippage planes of each grain to another.

Fine-grain material subsequently shows higher levels of hardness and strength than coarse-grain structures as the combined length (or area) of grain boundaries which prevent movement of dislocations is higher in fine-grain material. Yield limit σ_y is affected by grain size as per the famous relation:

$$\sigma_y = \sigma_0 + K_y\, d^{-1/2}. \quad (4)$$

This equation, known as the Hall-Petch relationship, uses the average grain size d, and constants for the material type in question, σ_0 and K_y. Although this equation does not apply to extremely fine-grain polycrystalline materials (Wang 2002), it clearly indicates that the mechanical properties are greatly affected by grain size.

The described effect of grain size also plays an important role in non-homogeneous properties of white etched layers. As the contact stress changes by distance from the surface and any single WEL has shown several distinguishable bands, the bands located closer to the rail head (i.e. the 'sub-layers' in the subject WEL) are smaller in grain size, higher in microhardness level, and their microstructure appears more compact than the bands in the lower areas of the subject WEL.

The higher number of dislocations shown in this paper by the results of TEM-analyses of the rail chips is another mechanism which assists in stabilisation of the white etched layer. As the surface energy reduces, dislocations are first attracted to the surface (Bureika, Mikaliūnas 2008). The number of dislocations subsequently rises by the effect of plastic deformation that is caused by load and wear of the rail surface.

While the number of dislocations rises, martensite hardness increases as well. Carbon atoms dissolved in the octahedrical positions of martensite reduce the size of domains and increase the size of structural microstress, as demonstrated by the results of the X-ray analyses. Hoffmann *et al.* (2001) attributes the effects of carbon dissolved in martensite in relation to domain size: higher size of domains corresponds to higher density of dislocations.

Therefore, it can be observed that extreme microhardness and strength of martensite which was generated in austenitisation and subsequent abrupt cooling of the surface of the contact area between the wheel and the rail (Kalousek, Klein 1976) can be attributed to the mechanisms of stabilisation by dislocation, precipitation, and deformation, of stabilisation by reduction of grain size, and by increased micro-deformation within the martensite structure itself.

11. Conclusions

The processes occurring in the wheel–rail system lead to concentration of stress during certain modes of the rail vehicle travel (take-off, braking); stress is concentrated in certain distances from or in certain areas within the rail head surface. Whenever the degree of deformation in a given area exceeds the critical point, the first event is cracking of the cementite lamellas into coarse particles within the assumed surface layer. Subsequently, during the cyclic load, the cementite lamellas separates into finer segments which, in turn, achieve the morphology of extremely fine particles in the size of several nanometers. As the TEM analysis of these areas suggests, very high density of dislocations has been observed here as well.

At the same time, the effect of the discussed heat source in the wheel–rail contact results in heating required for austenitisation of the mass element in the rail head. However, the time interval of the contact between the wheel and the rail is too short, and cementite cannot dissolve completely. The effects of the transmitted load result in considerable deformation of the austenite.

The subsequent cooling transforms the austenite in the areas of ultra-fine martensite. Therefore, the structure of the white etched layer includes this ultra-fine martensite, nanocrystalline cementite and residual austenite *(it should be noted, however, that the experimental part of this study did not provide evidence of residual austenite being present)*.

The white etched layers described here may be identified on the surface of worn rails and railway wheels, as well as in a range of other components such as piston rings, gears, during machining processes *(cutting and grinding technology)*, and in various operating conditions *(e.g. friction mechanisms)*.

This topic has been covered by many experimental and simulation studies. Nonetheless, there are differences in the individual observed mechanisms which lead to occurrence of various types of these layers. Therefore, the results obtained by the analyses presented herein may only be compared or applied to WEL which occur in similar material types and under similar load conditions.

From the transport engineering point of view, there is generally known, that the discussed considerable heat-affection of subsurface materials layers, and consequently risen quenched microstructures with high hardness level, can be indicated as an origin of heat and/or fatigue cracks, as well as the contact fatigue and falling of the material out from the railway wheel–rail tread, mostly in case of railway vehicles with disk brakes.

However, the mechanism of conditions resulting in microstructure changes in a rail and railway wheel material is not hitherto explicated in satisfactory way, and that is why the presented research is to be significantly applied in the railway-operation and practice, especially as a prevention and diagnostics.

Moreover, we can not discount an idea that the rising thermally-induced cracks following the circumferential direction of the wheel surface can change their orientation into radial direction, with a possible catastrophic failure (brittle fracture) of a wheel and imminent derailment of a railway vehicle, or a train.

These problems are of particular interest since the increase of axle loads in freight traffic as well as increase of velocity in passenger trains within recent years led to an increase of wear of the rails, which appeared e.g. in an increase of frequency of structure modifications and defects in their surface layer.

From the fracture mechanics point of view, the catastrophic brittle fracture of a rail and wheel can occur when reaching the critical size of these defects (radial cracks). Therefore, it is necessary to study these white etching layers, their characteristics and mechanisms of origin, as well as the 'thermal effects' on principle and to clarify the causes of their generation in order to achieve the higher safety of railway operation.

References

Bain, E. C. 1924. The nature of martensite, *Transactions of the American Institute of Mining, Metallurgical and Petroleum Engineers* 70: 25–35.

Baumann, G.; Fecht, H. J.; Liebelt, S. 1996. Formation of white-etching layers on rail treads, *Wear* 191(1–2): 133–140. http://dx.doi.org/10.1016/0043-1648(95)06733-7

Baumann, G. 1998. *Untersuchungen zu Gefügestrukturen und Eigenschaften der "Weissen Schichten" auf verriffelten Schienenlaufflächen*. Köster Bln. 132 S. (in German).

Beneš, L.; Schmidová, E. 2003a. White etching layer formation on rail treads, in *Proceedings of the 20th International Colloquium 'Materials, Technologies, Design, Maintenance – their Applications in the Field of Transportation'*, 102–105.

Beneš, L.; Schmidová, E. 2003b. Microstructure changes on the surface of rail tread, in *Proceedings of the 20th International Colloquium 'Materials, Technologies, Design, Maintenance – their Applications in the Field of Transportation'*, 106–109.

Bever, M. B. 1986. *Encyclopaedia of Materials Science and Engineering*. The MIT Press. 5000 p.

Bureika, G.; Mikaliūnas, Š. 2008. Research on the compatibility of the calculation methods of rolling-stock brakes, *Transport* 23(4): 351–355. http://dx.doi.org/10.3846/1648-4142.2008.23.351-355

Burša, J.; Janíček, P. 2009. Systemic approach to modelling of constitutive behaviour of various types of matter. Part I – Basic and simple constitutive models, *Engineering Mechanics* 16(4): 271–285.

Clayton, P. 1996. Tribological aspects of wheel–rail contact: a review of recent experimental research, *Wear* 191(1–2): 170–183. http://dx.doi.org/10.1016/0043-1648(95)06651-9

Djahanbakhsh, M.; Łojkowski, W.; Bürkle, G.; Baumann, G.; Ivanisenko, J.; Valiev, R. Z.; Fecht, H. J. 2001. Nanostructure formation and mechanical alloying in the wheel/rail contact area of high speed trains in comparison with other synthesis routes, *Materials Science Forum* 360–362: 175–182.
http://dx.doi.org/10.4028/www.scientific.net/MSF.360-362.175

Eckstein, H. J. 1971. *Wärmebehandlung von Stahl – Metallkundliche Grundlagen*. Deutscher Verlag für Grundstoffindustrie. 316 S. (in German).

Ermolaev, G. N. 2000. Mechanical properties of materials in the range of martensitic transformations at high pressures, *The Physics of Metals and Metallography* 89(4): 399–402.

Galliera, G.; Amici, G.; Ghidini, A.; Bertelli, R. 1995. Fatigue behaviour of railway wheels affected by sub-surface defects in the tread. Control methods and manufacturing process, in *Proceedings of the 11th International Wheel-Set Congress*, Paris, 69–76.

Ganapathi, S. K.; Aindow, M.; Fraser, H. L.; Rigney, D. A. 1990. A comparative study of the nanocrystalline material produced by sliding wear and inert gas condensation, *MRS Proceedings* 206: 593–598.
http://dx.doi.org/10.1557/PROC-206-593

Hellier, A. K.; Corderoy, D. J. H.; McGirr, M. B. 1986. A study of subsurface rail/wheel contact stresses with application to modelling rail fatigue, in *Proceedings of the 2nd Symposium on Contact Mechanics and Wear of Rail/Wheel Systems*, 421–434.

Hoffmann, B.; Vöhringer, O.; Macherauch, E. 2001. Effect of compressive plastic deformation on mean lattice strains, dislocation densities and flow stresses of martensitically hardened steels, *Materials Science and Engineering: A* 319–321: 299–303.
http://dx.doi.org/10.1016/S0921-5093(01)00978-9

Jergéus, J. 1992. *Wheel Flats and Martensite Formation in Railway Wheel Treads a Literature Survey*. Report F156, Chalmers University of Technology, Division of Solid Mechanics, Göteborg, Sweden. 19 p.

Kalousek, J.; Klein, R. 1976. Investigation into causes of rail corrugations, *AREA Bulletin* 77: 429–448.

Kout, J. 2001. *Poruchy oběžných ploch železničních kol brzděných kotoučovou brzdou*. Habilitační práce. VÚŽ Praha. 47 s. (in Czech).

Kuznetsov, V. D.; Savitskiy, K. V.; Sukharna, N. N. 1963. Some features of the structure of white layers, *Physics of Metals and Metallography* 15(1): 145–147.

Lata, M.; Čáp, J. 2010. The steepness of the rising branch of adhesive characteristics between wheel and rail, *Transport* 25(1): 17–21. http://dx.doi.org/10.3846/transport.2010.03

Liebelt, S.; Knotthe, K. 1997. *Ermittlung der Temperaturen beim Gleitkontakt – Anwendung auf das Rad/Schiene – System*. TU-Berlin, interner Bericht, Veröffentlichung in Weir geplant (in German).

Lojkowski, W.; Djahanbakhsh, M.; Bürkle, G.; Gierlotka, S.; Zielinski, W.; Fecht, H.-J. 2001. Nanostructure formation on the surface of railway tracks, *Materials Science and Engineering: A* 303(1–2): 197–208.
http://dx.doi.org/10.1016/S0921-5093(00)01947-X

Mitura, K.; Matušek, P.; Faja, R. 1980. Tepelné poškozování železničních kol způsobené intenzivním brzděním, *Železniční Technika* 10: 24–28 (in Czech).

Newcomb, S. B.; Stobbs, W. M. 1984. A transmission electron microscopy study of the white-etchinglayer on a rail head, *Materials Science and Engineering* 66(2): 195–204.
http://dx.doi.org/10.1016/0025-5416(84)90180-0

Ortner, B.; Kühlein, W.; Stüwe, H.-P. 1992. Martensite formation by relaxation of hydrostatic pressure, *Zeitschrift für Metallkunde* 83(6): 405–408 (in German).

Předpis SŽDC (ČD) S67 – Vady a lomy kolejnic. 1997. (in Czech).

Quinn, T. F. J.; Winer, W. O. 1987. An experimental study of the "hot-spots" occurring during the oxidational wear of tool steel on sapphire, *Journal of Tribology* 109(2): 315–319.
http://dx.doi.org/10.1115/1.3261358
http://dx.doi.org/10.1016/S0921-5093(97)00183-4

Tournay, H. M.; Mulder, J. M. 1996. The transition from the wear to the stress regime, *Wear* 191(1–2): 107–112.
http://dx.doi.org/10.1016/0043-1648(95)06693-4

Valiev, R. Z. 1997. Structure and mechanical properties of ultrafine-grained metals, *Materials Science and Engineering: A* 234–236: 59–66.

Wang, L. 2002. *Microstructure and Residual Stress State in The Contact Zone of Rails and Wheels*: Dissertation Thesis. TU Berlin. 166 p. Available from Internet: http://opus.kobv.de/tuberlin/volltexte/2002/449/pdf/wang_lihong.pdf