Óbuda University

Doctoral School on Materials and Technologies Sciences

Research on the surface wear and corrosion resistivity

Habilitation Thesis Book

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Content

Introduction

In a very general sense, the damage is a dissipative (irreversible) process, that is known to occur by a variety of mechanisms, depending on the properties of the material and the external conditions. Due to the complexities of the physical process resulting material damage, fatigue, creep, wear or corrosion may act together synergistically in most practical cases. For example, high-temperature components (such as various pressure vessels) are correspondingly subjected to combined creep, fatigue and corrosion. The surface itself is a momentous part of the mechanisms. The material damage is usually initiated on the surface. The surfaces of the structures are exposed to several mechanical and chemical effects. The lifetime of the structures can be predicted by the resistivity of the surface. The most aggressive damage initiation effects are wear [1-9] and corrosion [10-13]. In industrial practice, we can find several methods and processes to increase the surface resistance against these effects. While researching resistivity increasing processes, it has found the surface properties of the materials to be key elements. The surface of the materials is mostly characterised by the chemical composition, structure, surface roughness, and mechanical properties. During several years of research, I have studied how can the damage mechanism propagation depends on the surface characteristic and how can the surface resistivity of the surface against these mechanisms be increased. The foundation of this habilitation thesis book is a collection of the most important self-published articles [KT1 – KT10] of the research.

I. Research antecedent

The surface resistivity research focuses on the wear and corrosion phenomena. Both wear and corrosion processes can be characterized by weight loss as a function of time. The correlation between the wear resistance and the hardness is well known [1-9, O1-O15]. Several types of wear (abrasive, adhesive, fretting, etc.) and corrosion (selective corrosion, pitting corrosion, chemical corrosion, galvanic corrosion, etc.) [10-13, O16-O24] are well defined. There are several known experimental methods to analyse the mechanism of the wear and the corrosion. The wear and the corrosion processes are complex phenomena. In the case of the abrasive wear, the samples move on each other's surface and by this dry sliding, they cut small particles from the surface.

To facilitate the process of increasing wear resistance, the surface hardness needs to be increased as well. The hardness of the materials depends on the chemical composition and the microstructure. There are several processes for surface hardening, like explosive hardening [14-28, O25-O36], surface coatings (PVD), and outer surface treatments [29-47, O37-O43]. Several research results introduce the correlation between surface hardness and wear resistance. I concluded - and published in my research results -, that the steels with the same hardness can show different wear resistance as a function of the microstructure [O9, O12]. Apparently, the hardness and the microstructure collectively describe well the wear resistance of the materials well [KT2, KT4]. Concerning corrosion, some correlation can also be found between the surface properties and the corrosion resistance. The corrosion resistivity depends on the surface chemical composition and microstructure [KT5, KT6]. Understanding the microstructure and its influence on wear and corrosion resistance is of great importance for designers and engineers in selecting wear-resistant materials [2]. Sometimes it occurs that under conditions of wear, the original structure with higher hardness does not exhibit a better wear resistance [3]. Experimental results demonstrate the influence of the local microstructure on the mechanism of the wear and corrosion processes and the dependency of wear and corrosion resistance as a function of the microstructure.

I.1. Comparative wear resistance testing method

In most cases, the wear process takes place under such circumstances when various parameters - depending on the local coordinates - have a significant effect on the kinetics of the wear [4]. Among the position-dependent parameters, the microstructure and the surface roughness of the specimen are especially important. Both are determined by the preliminary manufacturing procedures (heat treatment, plastic deformation, processing technology). In both industrial and research practice, one can find several tribotesters to investigate the wear resistance of the materials. For testing comparative wear resistance, using, a simple ball cratering tribometer grants results in a short time. The reason for the selection of this method was the availability and the optimal experimental time. The experimental setup is based on a ball/plane impact. It used a rolling bearing which is fixed at an angle to support the ball, so it would make it possible for the ball to do a special

circulating movement. The importance of this movement is that in this build-up it doesn't have to take the wear of the ball into our consideration.

For the comparative wear resistance investigation is suitable the ball cratering tribotester (the ball cratering tribometer is shown in Figure 1) is suitable. The experimental parameters in the case of the ball cratering setup are the following: load is constant (N = 0,86 N), and the test time is 5 min. The wear cratering tool is an Al_2O_3 ceramic ball with a radius of R = 10 mm, whereas the rotation speed of the ball was n = 570 rot/min.

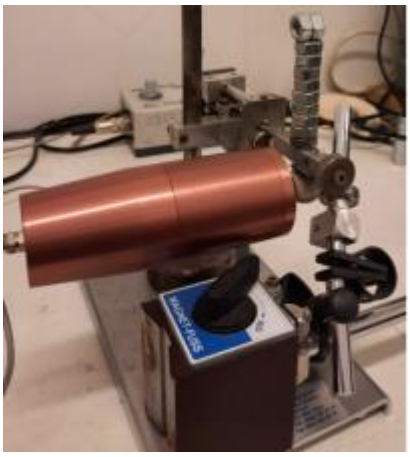

Figure 1. Ball cratering tribotester [O5-O9, KT2, KT10]

The wear coefficient can be determined using the dimensions of the resulting crater and the test parameters according to Equation (4). The h (mm) is the depth of the wear crater determined by Equation (1), where R (mm) is the ball radius, and r (mm) is the radius of the wear crater:

$$h = R - \sqrt{R^2 - r^2} \tag{1}$$

The wear volume V (mm^3) is determined by the Equation (2):

$$V = \frac{\pi \cdot h}{6} \cdot \left(\frac{3}{4} \cdot (2r)^2 + h^2\right) \tag{2}$$

The test length (m) is determined by Equation (3), where n (rot/min) is the speed of the ball, t (min) is the test time and R (mm) is the diameter of the ball:

$$S = n \cdot 2 \cdot R \cdot \pi \cdot t \tag{3}$$

The wear coefficient K (mm^3/Nm) is determined by Equation (4), where S (m) is the test length, N (N) is the normal load and V (mm^3) is the lost wear volume:

$$K = \frac{V}{S \cdot N} \tag{4}$$

I.2. Comparative corrosion resistance testing method for austenitic steel

It can find several methods to determine the corrosion kinetic. To compare the corrosion kinetic of the test samples can use a simple method. A standardised method is the Huey Test (ASTM A262) [13], for detecting the intergranular attack in the austenitic stainless steel when the corrosive material is a concentrated FeIIICl at 95°C, for 48 hours. The testing of the austenitic stainless steel intergranular corrosion tendency needs to do heat treatment of the test samples before the corrosion tests. The most commonly used sensitising treatment is 1 h at 675°C. It knows that the most important corrosion forms in the case of austenitic steels are the intergranular and the stress corrosion. The well-known susceptibility of austenitic stainless steels to intergranular corrosion after heat - treatment in the temperature range of 500°–800°C (sensitization) has long been attributed to the depletion of Cr from regions of the alloy matrix adjacent to grain boundaries in which $Cr_{23}C_6$ had precipitated. Those regions of the steel in which the local Cr composition falls below about 12% have a diminished ability to form a passive film and hence corrode preferentially [11]. The fabrication and welding technology affected microstructure can show high corrosion sensitivity [12]. The $Cr_{23}C_6$ precipitation showed in Figure 2. The plastic deformation modifies the grain size, and the welding technology causes heat input that can affect other transformation and precipitation phenomena.

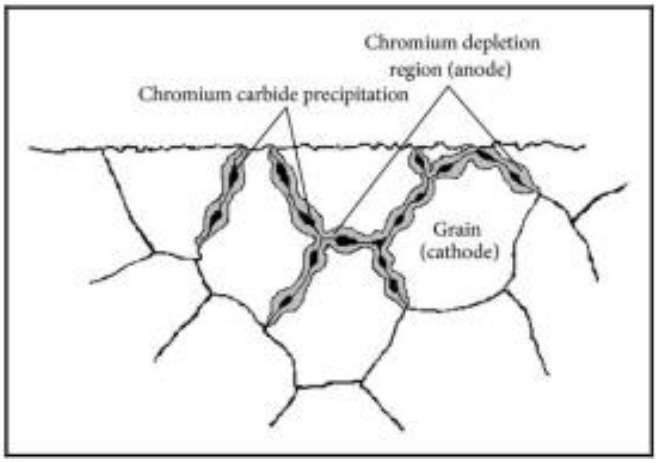

Figure 2. Schematic representation of the carbide precipitation at grain boundaries in austenitic stainless steel [11]

The kinetic determination of the corrosion is a goal of several kinds of research [12, O21-O23, KT5, KT6].

I.3. Explosive hardening

Shock hardening or explosive hardening is a beneficial and common technology [14, 17, 18, 21, 23, O28, KT4]. Explosive hardening of railway frogs from Hadfield steel (Mn steel) is a common technology in the world, which allows for increases in the surface and subsurface hardness of frogs (Figure 4) [23, 25, O28]. This hardening technology is also able to increase the hardness and the wear resistance of the austenitic stainless steel too [27, O34, KT2, KT5]. The austenitic stainless steel has great ductility, low hardness and very good corrosion resistance. It is impossible to increase the hardness by the way of simple heat treating.

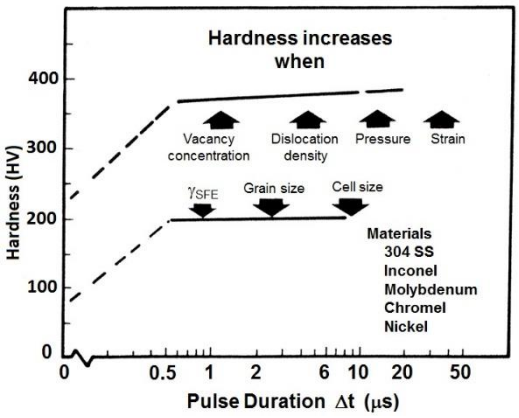

Figure 3. Hardness increasing [25, 27, KT2]

Cold working and ageing heat treatment involve hardness increasing in the case of this steel. That static strain ageing is a well-known phenomenon frequently observed in the body-centred cubic crystal structure (bcc) metals and alloys (Figure 3 shows the explosive hardening effect) [24, 25, 27].

It is known the explosion shock occurs a hardness rising but the parameters of this phenomenon have not been fully understood and studied yet. The effect of strain rate on the ϒ-α' transformation in stainless steel has been of interest to researchers for some years. The early work simply noted if the pulse duration (Δt) increases, also increase the amount of martensite and the hardness of the steel also increases [25].

The conventional work hardening mechanism of Hadfield steel involves mainly dislocation, twinning as well as dynamic strain ageing [22]. Though explosively hardened

Hadfield steel crossings have been widely used in railways around the world, the deformation mechanism of this steel during explosive impact is not well understood [23].

This austenitic steel contains approximately 1,2% carbon and 12% manganese in a 1 to 10 ratio. It was unique in that it exhibited high toughness, high ductility, high work hardening ability and excellent wear resistance. Because of these properties, Hadfield's austenitic manganese steel (AMS) gained rapid acceptance as a useful Engineering material [24]. Figure 4 shows the typical application of the casted manganese steel in the railway crossing nose.

Figure 4. Typical casted manganese crossing [25]

It's known that this steel microstructure and hardness change during forging. The casted samples have an austenitic microstructure, that after cold working, transforms to martensite. The explosion effect also involves a microstructure and hardness changing, with almost the same result as after forging.

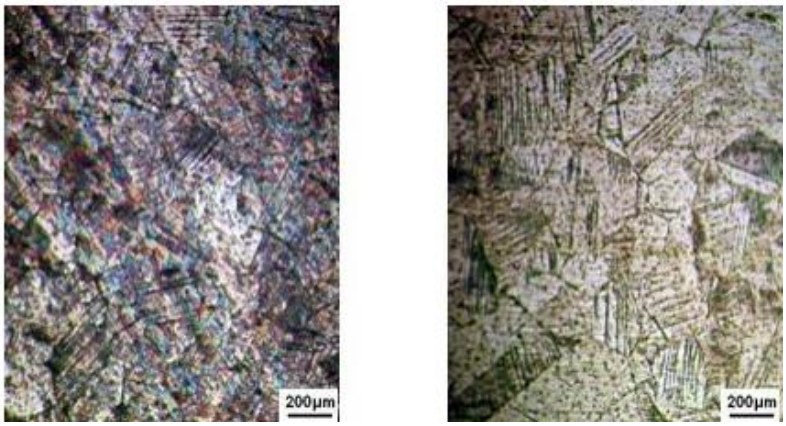

Figure 5. Optical macrographs of the direct hardened Hadfield steel (etching them using a solution containing 4% HNO_3 and 96% C_2H_5OH)

Figure 5 shows the microstructure of the surface layer after direct hardening by two different explosive materials. The microstructures in the case of both samples are martensitic [KT4].

It has been commonly accepted that AMS subjected to rapid work hardening is stable during plastic strain (Figure 6). However, nowadays evidence to suggests that under rapid work hardening and plastic deformation, AMS undergoes strain-induced stress transformation from γ austenite to α ferrite or ε martensite. It demonstrated that the transformation from γ austenite to ε martensite is dependent on the strain rate [24].

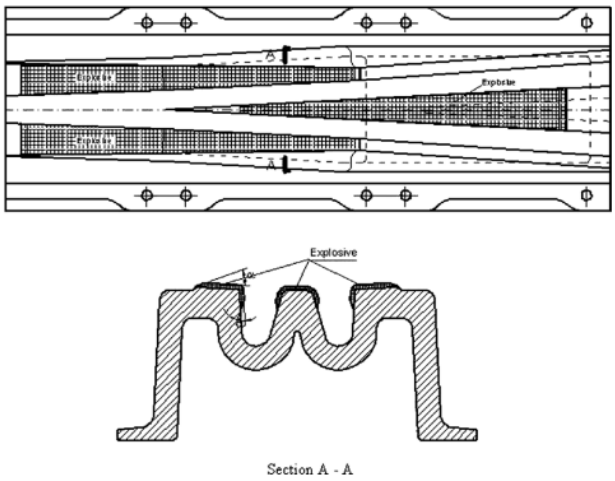

Figure 6. Sketches of Hadfield steel crossing and its section as well as direct explosive hardening places on the surface [28].

Explosive-hardening of metals has long been used industrially and AMS (12-14% Mn contain) have long been singled out for particular use [20]. Hadfield steel has been widely used to manufacture railway crossings because of its excellent work hardening, high strength and toughness properties [21].

Figure 7. Setup of the direct explosion hardening [22]

The setup of the Hadfield frog hardening can see in Figure 6. The experimental setup of the direct explosive surface hardening is illustrated in Figure 7. The direct explosive surface hardening means that the explosive material is located on the steel surface (show Figure 6 and Figure 7)

The pressure of the nascent gases is calculated by the Equation (5) [29]:

$$P = \frac{1}{4} v_d^{\ 2} \rho_0 \tag{5}$$

Where:

- v_d: detonation velocity of explosive [m/s];
- ρ_0: density of explosive [kg/m^3];
- ρ: density of the nascent gases [kg/dm^3];

The explosion impact force on the surface (6) [29]:

$$J = \int p dt \tag{6}$$

Where:

- J: impulse on the surface unit [N/m^2];
- P: nascent gases pressure from the Equation (5) [Pa].

The P pressure quantity depends on the parameters of the explosive material and the effect time depends on the amount (thickness) of the explosive material. The velocity of the collision (v_c) must be lower than the speed of the sound (v_s) (7), which means it needs to use a low-speed explosive for this technology. The interfacial pressure at the collision front also must exceed the material's yield strength to occur a plastic deformation. This is the surface hardening under extreme pressure [21, 29].

$$\frac{v_c}{v_s} < 1 \tag{7}$$

The thickness of the explosive powder can be optimized based on the practice. It is known that it needs a minimal amount of explosive, about 0,017 (g/mm^2) Permont 10T [29]. The hardness increases depending on the pressure of the nascent gases (see Figure 8).

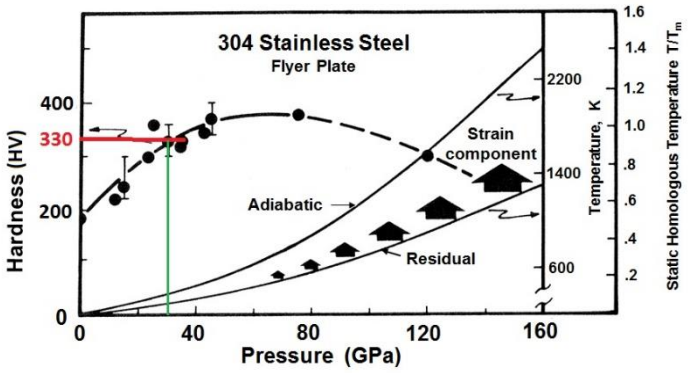

Figure 8. The hardness as a function of the pressure of nascent gases [27]

The explosion kinetic, therefore the nascent pressure (P) is displayed as a function of time in Figure 9. It can be seen that the nascent pressure increases in the first and second periods and in the third period the pressure is constant.

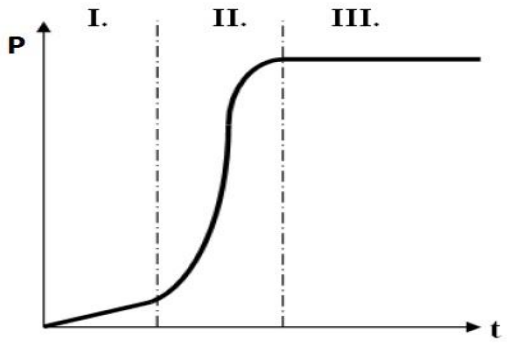

Figure 9. The kinetics of the detonation
(I.: burning period, II.: explosion period, III.: detonation period) [28]

I.4. Surface coating

Improving the wear resistance of the steel is a pressing requirement from the industrial side. For tool steels, the better wear resistance means a longer lifetime, which directly influences the production rate and ultimately, the price of the manufactured products. Thus, high productivity requires tool steels with a long lifetime and high performance. The required properties of tool steels are high strength at elevated temperatures, high impact strength, hot wear resistance, good toughness, and good hardenability. Despite high strength tool steels having good hot working capabilities, at elevated temperatures during the manufacturing process they suffer from surface oxidation, decarbonization, chemical interactions between oxides, tribological contacts as well as

thermal and mechanical loads [30-35]. High strength tool steels are widely used in hot-working processes (above 200°C). To ensure consistent production quality at high volume productions, the surface roughness and wear properties of such tools need to be superior. For surface modification commonly used to improve the wear resistance of tool steels, several processes are suitable [36-40, KT7-KT10]. Common industrial processes to improve the wear resistance and the mechanical properties of the surface of tool steels are nitriding, surface hardening, laser surface treatment and surface coating [31, 41]. Different research groups have shown the improvement possibilities of lifetime and performance of hot work tools by applying surface treatment processes, such as plasma nitriding, PVD coating or a combination of treatments [42-45]. During plasma nitriding, a nitride layer is formed on the surface of the used metal, which significantly improves the hardness of the surface [46]. PVD is used to develop layers of CrN, TiN, CrAlN or TiAlN depending on the chemical composition of the steel to improve surface hardness [47]. The improvement of PVD technology allows the development of multi-layer systems and the surface modification after nitriding [48]. Coated and heat-treated surface characterization was published by a different research group [49–52]. The improvement of wear resistance can be attributed to the combined effects of surface roughness, surface hardness and stiffness. Based on fracture mechanics and crack propagation theories the damaging process usually starts with the failure of the surface [53]. Researchers found that with a refinement of the surface roughness the fatigue limit of the surface layer declines. The literature discusses multiple methods to modify the surface characteristic of different materials, however, all the methods and characterization methodology are specially optimised to produce applicable results [54–56]. Further studies investigate the improvement possibilities of wear resistance of tool steel. These studies indicate that surface roughness by itself does not provide a definite indication of wear resistance. Furthermore, to obtain the desired wear properties, several layers are required with different stiffness and elasticity [57–59]. TiN monolayer deposited by PVD has inferior mechanical and wear properties compared to TiN/Ti or TiN/TiCN multilayers [60, 61]. The monolayers are more brittle than the multilayer systems and the damaging process starts sooner [62]. Thus, the development of a multilayer system by PVD to improve wear resistance is more advantageous. Tribological property investigation of TiAlN, CrAlN, Al_2O_3, SiC and B_4C PVD coatings has shown that the performance of the treated tools depends on several parameters such as substrate

chemical composition and microstructure, substrate mechanical properties, deposition method and chemical composition of the deposited layer [62–68]. The friction and wear properties can be improved by the deposition of TiN, TiAlN, AlTiN and CrAlN layers by PVD [28, O37-O43]. Another way to increase the performance of tool steel is nanolayer technology and duplex coating treatments [66]. The duplex PVD process is also a suitable technology to create high-strength surface layers [67, 68].

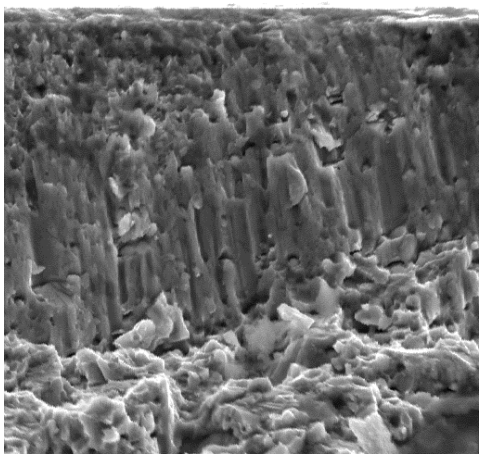

Figure 10. SEM Cross-section of the TiN coating

Figure 10 shows a TiN coating cross-section with the columnar structure. Different layers of the surface as a function of the coating thickness and chemical composition (nano multilayer PVD coating) increase the wear resistance and lifetime of the tools [67]. Based on the literature review and the material science studies, it can be concluded that increasing the lifetime of tool steels must be a complex treatment and the substrate's mechanical and chemical properties are extremely important [68].

II. New scientific results

Thesis 1.
I developed a new explosive hardening process specification, namely the *indirect explosive surface hardening process*, and demonstrated that this process results in a bigger hardness of the surface than the conventional direct process in the case of the tested steels (X120Mn12, X5CrNi1810) when the setup hole is between 1,5-10 mm [KT1, KT2, KT3, KT4].

Direct hardening is a common surface hardening of the Hadfield steel, usually, a rapid process usually used to increase the surface hardness of the casted Hadfield steel railway crossing. It is known the explosion shock affects a hardness-rising phenomenon, but the parameters of this process have not been well understood yet. The effect of strain rate on the γ-α' transformation in stainless steel has been studied for some years ago. The direct hardening setup is shown in Figure 11 and the indirect explosive surface hardening process is shown in Figure 12.

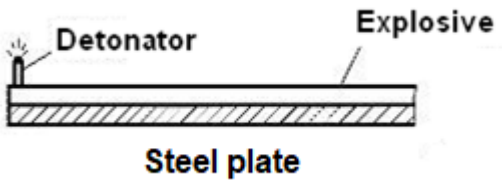

Figure 11. Direct explosive hardening setup

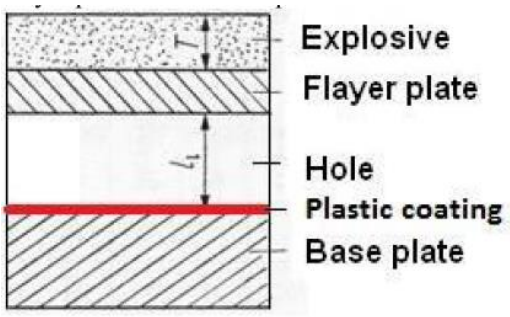

Figure 12. Indirect explosive hardening setup [KT2, KT4].

The setup of the direct hardening which was used is shown in Figure 11 when the explosive charge detonates in contact with metal, creating a detonation shock to the steel surface. The hardening depends on the explosive parameters detonation velocity and density of the explosive. The detonation is controlled and reproducible. The mechanical properties change through the pressure of the detonation.

In the case of the indirect surface hardening process, a setup was used (shown in Figure 12) when between the explosive and the metal there was some millimeter distance. For this investigation, a 1 mm thick mild steel (S235JR) was used as a flayer plate (see Figure 12). This setup is similar to the explosion cladding technology. The used explosive

was also Permon 10T (powder). In the case of indirect hardening, it needs to use a thin layer (plastic coating see Figure 12) needs to be used on the surface of the base plate, which prevents the joining of the base plate and the hardener flayer plate.

We measured the surface hardness ($HV_{1,2}$) of the direct and indirect hardened test samples. The explosive hardening is made by a Permon 10T explosive material. The used materials were annealed Hadfield steel X120Mn12 (110 $HV_{1,2}$) and stainless steel X5CrNi1810 (215$HV_{1,25}$). The used explosive material and the setup parameters are summarized in Table 1.

Table 1. Parameters of used explosive hardening

Explosive PERMON 10T (powder)			
The specific volume of the gas	928 dm^3/kg	Hole size (indirect hardening)	1.5 mm
Detonation rate	3200 m/s	The thickness of the explosive	30 mm
Density	850 kg/m^3	Weight	319 g

Table 2. Surface hardnesses

Materials	Base metal ($HV_{1,2}$)	Direct hardening ($HV_{1,2}$)	Indirect hardening ($HV_{1,2}$)
X120Mn12	110	238,4	472,348
X5CrNi1810	215	263	322

The summarised experimental results of the direct and the indirect explosive surface hardening are shown in Table 2 and Figure 13.

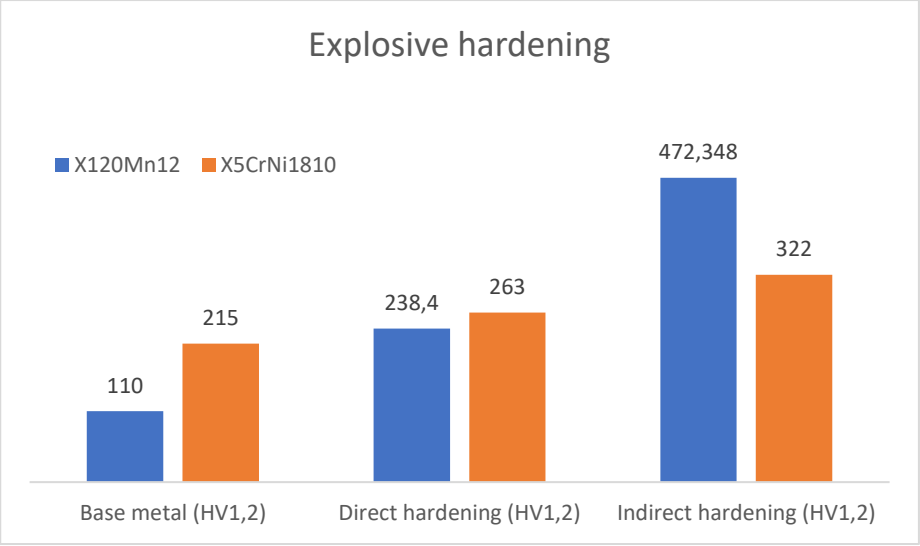

Figure 13. The results of the explosive hardening

The research results verify the thesis because the indirect explosive surface hardening results in a higher hardness increase than the traditional direct explosive hardening process. It can suggest the indirect explosive surface hardening in the practice.

Thesis 2.

I demonstrated that the obtained surface hardness in the case of the tested steels (X5CrNi1810, X120Mn12) depends on the hole size (up to 10 mm) of the indirect explosive surface hardening setup [KT1, KT2, KT3, KT4].

The setup of the direct explosive surface hardening is shown in Figure 11 while the indirect explosive surface hardening setup is shown in Figure 12. The obtained hardness values as a function of the hole size of the indirect explosive surface hardening are summarized in Table 3 and shown in Figure 14.

Table 3. The earned hardness as a function of the hardening hole size of the indirect explosive surface hardening

Steel	**X120Mn12**	**X5CrNi1810**
Hole size	**Hardness HV_{30}**	
0	215	263
2	222	322
3	217	330
4	243	335
7	not tested	340
10	not tested	348

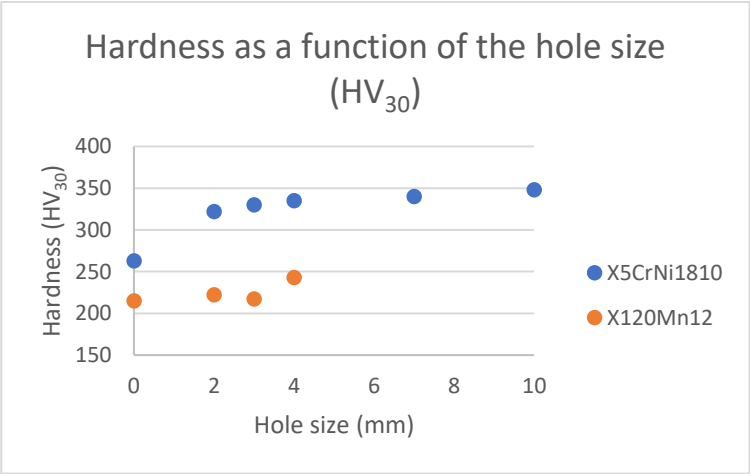

Figure 14. The surface hardness is a function of the hole size

The surface hardness increases as a function of the setup hole sizes because the resulting hardness depends on the plastic deformation rate and the plastic deformation depends on the collision energy of the flayer plate.

In the case of indirect hardening, the hardness of the base plate (Hadfield steel) is increased higher than in the case of direct hardening. The reason for this difference is that in the case of the indirect hardening the hardener plate interfered with the base plate surface at a very high velocity.

The experimental results verify the thesis, that the indirect explosive surface hardening influences the hardness depending on the setup hole size.

Thesis 3.

I demonstrated that the indirect explosive surface hardening technology increases the wear resistance of the tested steel (X5CrNi1810) as a function of the indirect explosive surface hardening setup hole size [KT1, KT2, KT3, KT4].

The comparative wear resistance of the hardened samples was tested by a ball cratering tribometer in case of standard parameters (load, rotation per minute, investigation time, etc.). The wear coefficient K (mm^3/Nm) was determined by Equation (4), where S (m) was the test length, N (N) was the normal load and V (mm^3) was the lost wear volume:

$$K = \frac{V}{S \cdot N} \tag{4}$$

The wear resistance R (Nm/m^3) is the reciprocal value of the wear coefficient:

$$R = \frac{1}{K} \tag{8}$$

The results of the wear resistance test are summarised in Table 4.

Table 4. Wear resistance test results

Hole size of the hardening setup	**Hardness HV_{30}**	**Wear coefficient K (mm^3/Nm)**	**Wear resistance R= 1/K (Nm/mm^3)**
2	222	$8{,}93\ 10^{-20}$	$1{,}12\ 10^{19}$
3	217	$8{,}29\ 10^{-20}$	$1{,}21\ 10^{19}$
4	243	$6{,}33\ 10^{-20}$	$1{,}58\ 10^{19}$

Figure 15 shows the wear resistance as a function of the hardening setup hole size and the hardness. The submitted research results verify the thesis, that the indirect explosive surface hardening technology increases the wear resistance of the tested steel as a function of the hole size.

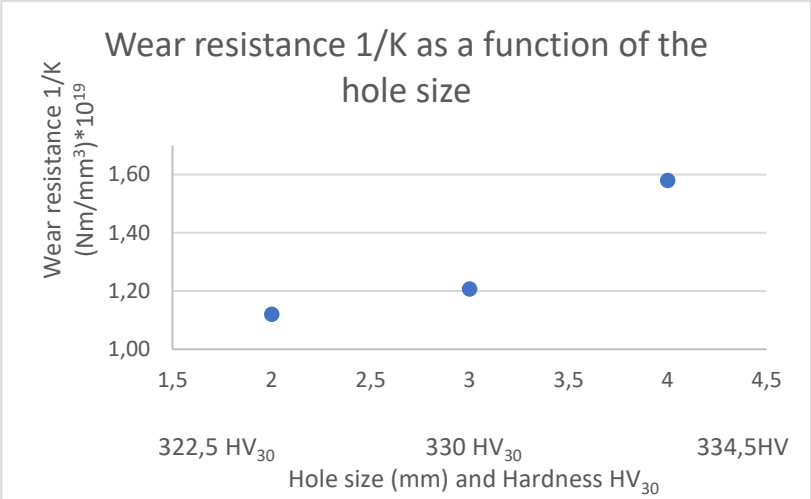

Figure 15. Wear coefficient as a function of the hole size and the hardness

Thesis 4.

I demonstrated that the stainless steel (X2CrNi18-9) corrosion resistance dependence on the surface roughness is bigger in the case of the heated at 800 °C than the non-heated stainless steel corrosion resistance dependence [KT5, KT6].

In the experimental study, the welding heat effect was modelled by heat treating. It heated all samples for one hour at 800ºC and cooling was by air. After the heat treatment, the test samples' surfaces were ground with different quality grinding papers. The measured surface roughnesses find in Table 5.

Table 5. The grinding parameters and the surface roughness

Test sample Identification number	**N 1**	**N 2**	**N 5**	**N 6**
Grinding paper	P120	P180	P320	P400
Surface roughness Ra(µm)	2,353	1,412	0,677	0,54

The surface roughness and heat caused changes in both microstructure and corrosion resistance of the tested samples. This process was modelled by laboratory experiments ASTM A262. Under this load, the sheets showed a weight loss that is measurable with analytical scales. The measured weight loss volumes (g) are shown in Table 6.

Table 6. Corrosion test results

Test sample Identification number	**Weight loss (g)**	
	Without heat treating	**Heat-treated**
N 1	0,4217 g	0,9608 g
N 2	0,4005 g	1,2904 g
N 5	0,4468 g	1,0636 g
N 6	0,561 g	1,7667 g

Tested samples were examined by microscopy and the results are shown in Figures 16-19. On the surface it observed pitting corrosion phenomena through visual testing, it used

stereo microscopy. It was observed that in the case of the tested samples the unsuitable surface roughness caused reduced corrosion resistance. In the case of the heat-treated samples, the heat treatment decreased corrosion resistance. The corrosion was recognizable by visual testing (Figure 16-19) and measurable by weight loss control.

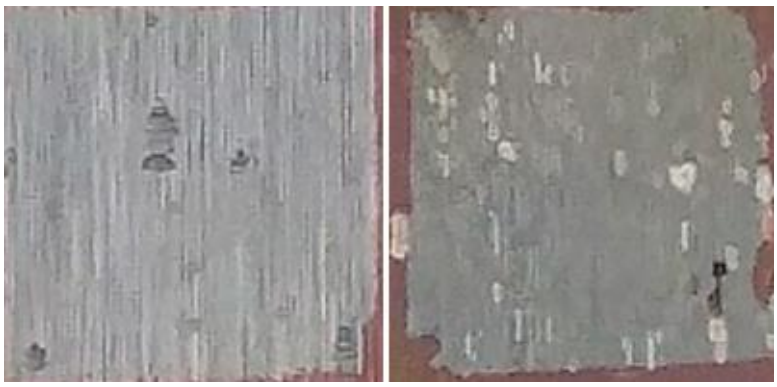

Figure 16. N. 1. samples: a) without heat treatment, b) heat treated

Figure 17. N. 2. samples: a) without heat treatment, b) heat treated

Figure 18. N. 5. samples: a) without heat treatment, b) heat treated

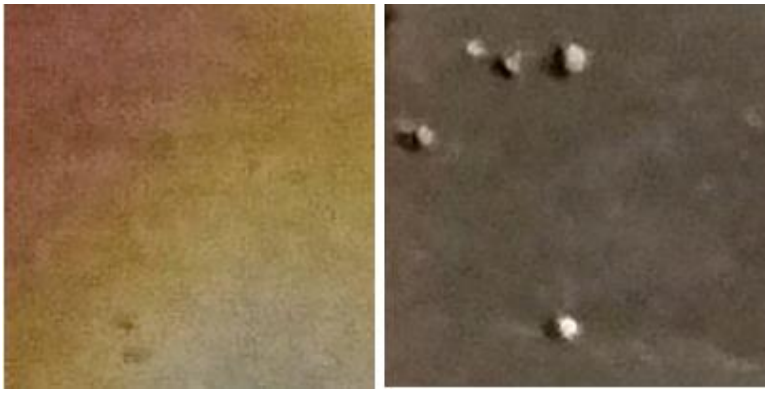

Figure 19. N. 6. samples: a) without heat treatment, b) heat treated

It was detected a relationship between the surface roughness and the weight loss. Figure 20 shows the relationship between the surface roughness and the weight loss in the case of the heat-treated samples and other samples without heat-treatment.

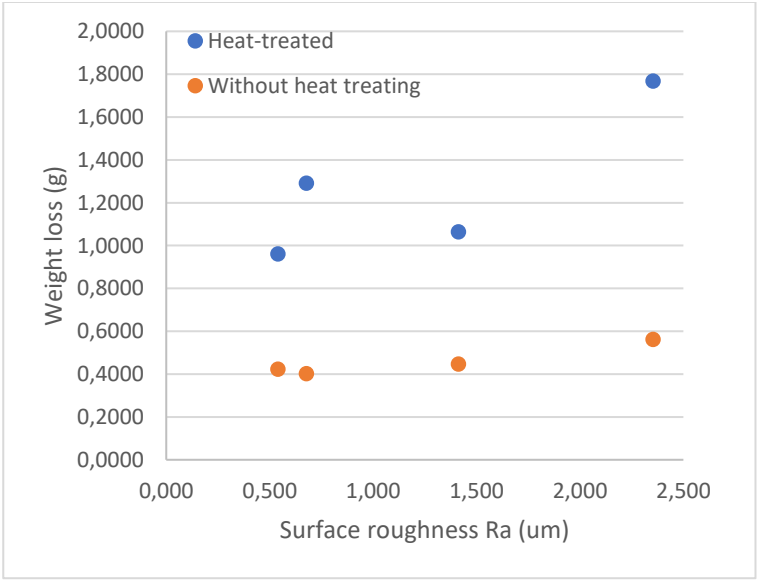

Figure 20. The weight loss of the tested samples

Based on the experimental results, it can conclude that the corrosion resistance decrease is more significant in the case of the heat-treated samples.

Thesis 5.

I demonstrated, that the wear resistance (1/K) of the CA-PVD (Cr/CrN, TiN/TiAlN) coating is bigger in the case of the plasma nitriding underlayer application [KT7, KT8, KT9, KT10].

Table 6 shows the chemical composition of 1.2344 (X40CrMoV5-1) tool steel. The hardness of the steel sample was 229 HB. The tested disk-shaped samples had a Ø 15 mm × 3 mm dimension.

Table 6. Chemical composition of the tool steel

Element	C	Si	Mn	Cr	Mo	V
Concentration %	0.40	0.25	0.45	5.25	2.31	0.65

All samples were quenched and tempered. The austenitization was carried out at 1030 °C for 10 min, whereas N_2 gas was used as cooling media at 9 bar pressure. The quenching was followed by tempering at 550 °C, 580 °C and 540 °C, respectively, for 2 hours at each temperature in N_2 at 1,5 bar.

Test samples N2, N5, and N6 were subsequently subjected to plasma nitriding. The samples were placed into the plasma nitriding furnace right after heat treatment, where a cleaning cycle was performed.

Test samples N3, N4, N5, and N6 were treated in a Cathodic Arc Physical Vapour Deposition (CA-PVD) vacuum chamber having multiple cathodes. The temperature of the substrates was held at approximately 400 °C inside a vacuum chamber. CA-PVD was employed to prepare TiN/AlTiN and Cr/CrN multilayers. The nitriding and the CA-PVD coating are made by Surface Modification Technologies Pvt. Ltd. Vasai, India.

Wear resistance tests were carried out on a ball-cratering tribotester, shown in Figure 1. The used load was N_c = 0.86 N and the test time was 5 min. The wear cratering tool was an Al_2O_3 ceramic ball with a radius of R = 10 mm, whereas the rotation speed of the ball was n = 570 rot/min. The wear coefficient K (mm^3/Nm) was determined by Equation (4), where S (m) was the test length, N (N) was the normal load and V (mm^3) was the lost wear volume:

$$K = \frac{V}{S \cdot N} \tag{4}$$

The results of the microhardness, the surface roughness and the wear coefficient are collected in Table 7. Microhardness does not show a significant difference between plasma nitrided and non-plasma nitrided coatings. The wear coefficient in the case of the plasma nitrided and non-plasma nitrided coatings show a relevant difference.

Table 7. Microhardness, surface roughness and wear coefficient of the samples

Number of the test samples	**Test sample identification**	**Microhardness ($HV_{0.01}$)**	**Surface roughness Ra (µm)**	**Wear coefficient K (mm^3/Nm)**	**Wear resistance 1/K (Nm/mm^3)**
N1	Hardened	606	0,01045	$6{,}32 \cdot 10^{-9}$	$1{,}5 \cdot 10^{8}$
N2	Hardened and Plasmanitrided	1140	0,05745	$1.95 \cdot 10^{-9}$	$5{,}12 \cdot 10^{8}$
N3	Hardened and PVD coated TiN/TiAlN	2938	0,23335	$8{,}46 \cdot 10^{-10}$	$1{,}18 \cdot 10^{9}$
N4	Hardened and PVD coated Cr/CrN	2775	0,1768	$4{,}23 \cdot 10^{-11}$	$2{,}36 \cdot 10^{10}$
N5	Hardened, plasma nitrided and PVD coated TiN/TiAlN	2679	0,1895	$7{,}57 \cdot 10^{-10}$	$1{,}32 \cdot 10^{9}$
N6	Hardened, plasma nitrided and PVD coated Cr/CrN	2756	0,1700	$3{,}68 \cdot 10^{-12}$	$2{,}71 \cdot 10^{11}$

Figure 22 shows that the wear resistance 1/K in the case of the plasma nitrided under layered CA-PVD coatings (N5, N6) is bigger than in the case of the non-plasma nitrided under layered CA-PVD coatings (N3, N4). The reason for the difference can be interpreted by the underlayer's mechanical properties. Test sample N1 has the smallest wear resistance,

it is the hardened base steel. The Cr/CrN PVD coated test samples (N5, N6) have high wear resistance, even the N6 sample with a nitrided underlayer is the best one with the wear resistance value. The TiN/TiAlN PVD samples (N3, N5) show the same tendency because the N5 sample with a nitrided underlayer has a higher wear resistance than the N3 sample without a nitrided underlayer.

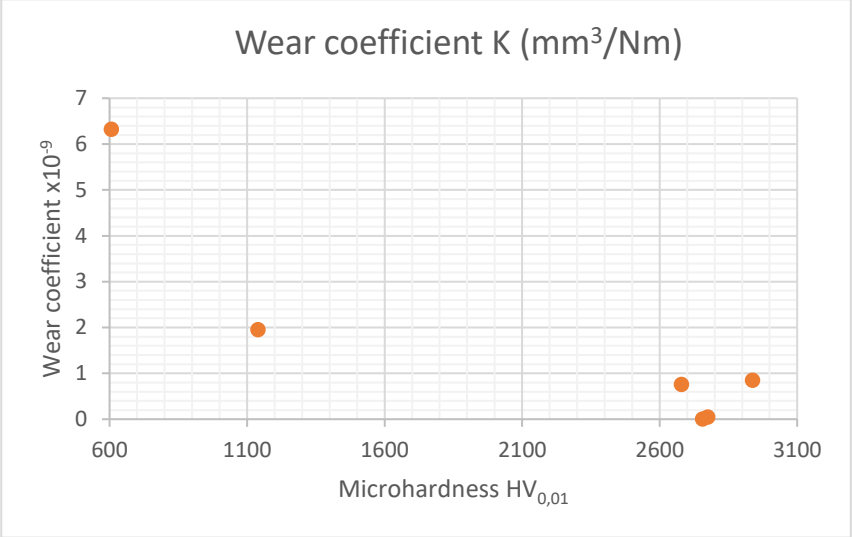

Figure 21. The wear coefficient as a function of the hardness

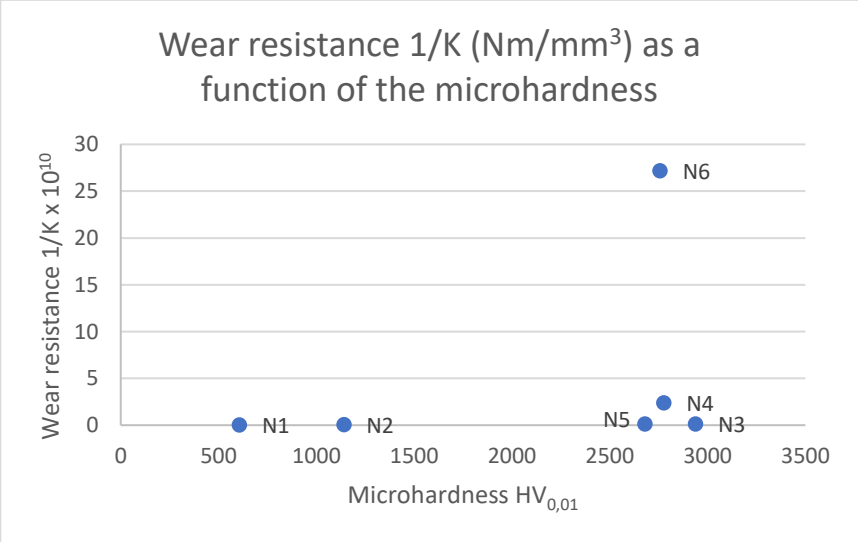

Figure 22. The wear resistance (1/K) as a function of the microhardness

Based on the experimental results, it can conclude that the nitrided underlayer increases the wear resistance of the TiN/TiAlN and the Cr/CrN PVD coatings.

II. The effects and echos of the research results

Numerous research focuses the welding. The explosion welding process is not a new but not well-known process. The published research results in the explosion welding and hardening affected a high number of references and motivating new research. In the Óbuda University Bánki Donát Faculty of Mechanical and Safety Engineering several TDK, BSc and MSc researches work realized (István Sikari Nágl BSc 2013, Bálint Völgyi BSc 2013, Tamás Rigó TDK 2013, etc.). The corrosion resistance research results have also great interest, several references and Ádám Szigeti have got a First price of the „most innovative BSc thesis work" racing (2017). The research of the surface coating processes and the process results testing affected also several references, MSc diploma work of Haidarh Shbanah (2022) and the international partnership from 2017 with Umesh Mhatre (Surface Modification Technologies Pvt. Ltd. India).

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