# Influence of gas nitriding conditions on layer structure formation on grey- and white-solidified cast irons

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The wear and corrosion behaviours of nitrided cast iron depend strongly on the material-specific morphology of its graphite. Furthermore, and depending on its specific form, this morphology can have either a positive or a negative effect on the potential for improvement of such behaviour. For this reason, comparative investigations were carried out to determine the influence of gas nitriding (GN) conditions on the layer structure and the properties formed for both grey-solidified (with graphite) and white-solidified (without graphite, but with ledeburitic carbides) cast iron surface layers generated by electron beam remelting (EBR). By varying the nitriding potential within the limits K<sub>N</sub> = 0.1 (no compound layers), K<sub>N</sub> = 1 and K<sub>N</sub> = 4 (with compound layers), as determined on the basis of the Lehrer diagram, the influence of different Si contents in the cast irons on the layer formation and phase composition were investigated. In addition, the influence of the morphology of the two different C-containing modifications - the stable graphite (lamellar, spheroidal) and the metastable ledeburitic carbides (secondary dendrite arm spacing: 3 µm - 7 µm) – on the formation of both the compound and diffusion layers was analysed by means of SEM investigations, hardness measurements and EBSD analysis. This allowed conclusions to be drawn about the microstructurally dependent nitriding mechanism during gas nitriding. For the metastable carbide layers, the subsequent thermal regime of the nitriding process applied (EBR+GN) may have influenced the initial microstructure. For this reason, the tempering behaviour was investigated in separate annealing tests under temperature-time regimes relevant for nitriding (540 °C to 670 °C / 4 h to 16 h). The hardness decrease induced by the coarsening of pearlitic cementite in the ledeburite was strongly dependent on the chemical composition of the cast iron used. At temperatures above 630 °C, the hardness dropped remarkably due to graphitizing. In light of these results, the nitriding parameters could be optimized.

# **KEYWORDS:** CAST IRON - ELECTRON BEAM REMELTING - GAS NITRIDING DUPLEX SURFACE TREATMENT

#### Introduction

Grey-solidified cast iron is interesting for numerous applications due to some of its special properties, such as good castability and associated near-net-shape component manufacture, high damping capacity, high compressive strength and emergency running properties [1,2]. Alloying with silicon (Si) promotes solidification according to the stable Fe-C system. Therefore, the microstructure consists of coarse graphite particles em-

Anja Holst, Anja Buchwalder, Rolf Zenker TU Bergakademie Freiberg, Institute of Materials Engineering Freiberg, Germany bedded in a ferritic or pearlitic matrix. As known from steel, nitriding is a common method for improving both the material's wear and corrosion resistance [3]. However, because the graphite is not nitrided, i.e., the layer is interrupted, the compound layer is inhomogeneous and does not always fulfil its function satisfactorily.

Electron beam remelting (EBR) of cast iron surfaces carried out before the nitriding treatment leads to the dissolution of near-surface graphite and, as a result of the process-related high cooling rate, to solidification according to the metastable Fe-Fe<sub>3</sub>C system.

The eutectic microstructure of white-solidified cast irons is referred to as ledeburite, which exhibits high hardness and has been investigated extensively [4–6].

The morphology of ledeburite depends on the cooling rate. The degree of microstructural refinement can be influenced by varying the energy input during EB remelting.

The higher the cooling rate, i.e., the thinner the remelted surface layers, the finer the remelted microstructure. During nitriding of the ledeburitic surface layers, a closed, homogeneous compound layer is formed [7]. Such combined treated surface layers (remelting + nitriding) are significantly more resistant to abrasive wear and corrosion [8,9].

In order to understand the nitriding mechanism and, in

particular, the role of silicon in nitriding, hypoeutectic pearlitic and ferritic cast irons were investigated in the initial (grey-solidified) and EB remelted (white-solidified) states.

Gas nitriding was carried out with different nitriding parameters (time  $t_N$ , nitriding potential  $K_N$ ) at a constant temperature  $T_N$  of 540 °C.

# Experimental

### **Base materials**

The investigations were carried out on commercially available grey-solidified pearlitic (P) and ferritic (F) cast irons with lamellar (GJL), spheroidal (GJS) and vermicular graphite (GJV), with the samples provided in the form of cast flat samples (100 × 60 × 20 mm<sup>3</sup>) by the foundry *Kessler and Co. GmbH* (Leipzig, Germany). Tab. 1 shows the respective chemical composition and the carbon equivalent (CE) calculated from it. Compared to conventional GJL-250 cast iron, small additions of chromium (Cr), copper (Cu) and titanium (Ti) served to modify the formation of the graphite lamellae in order to improve the melting behaviour during EB remelting [10]. The nominal tensile strength was not affected. This was marked in the material designation by the imposed character mod (GJL-250<sub>mod</sub>).

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Identification	Alloy	Element concentration [wt.%]									CE	Microstructure
		С	Si	Mn	P	S	Cr	Cu	Ti	Mg		(P = pearlite;
												F = territe)
P/L_2.0Si	GJL-250 <sub>mod</sub>	3.21	1.97	0.59	0.040	0.020	0.17	0.76	0.24	-	3.88	P, < 1-5 % F
P/S_2.3Si	GJS-600-3	3.61	2.33	0.34	0.030	0.010	0.03	0.67	-	0.04	4.40	P, < 1-3 % F
F/S_2.3Si	GJS-400-15	3.72	2.32	0.12	0.016	0.009	-	-	-	0.04	4.50	F, 5-20 % P
F/V_2.5Si	GJV-300	3.44	2.52	0.15	0.011	0.009	-	0.06	-	-	4.28	F, < 1 % P

Tab.1 - Nominal composition (Fe base) of the investigated alloys and the corresponding<br/>carbon equivalents CE (CE = wt.%C + ½ (wt.% Si + wt.% P))

The alloys were grouped with respect to their matrix microstructure (P - pearlitic, F - ferritic), graphite morphology (L - lamellar, S - spheroidal, V - vermicular) and silicon content, cf. Tab. 1. tic cast irons were 258 HV0.3 (P/L\_2.0Si) and 288 HV0.3 (P/S\_2.3Si), respectively, while that of the ferritic cast iron was 167 HV0.3 (F/S\_2.3Si, F/V\_2.5Si).

The measured mean core hardness values of the pearli-

#### Treatment parameters

The electron beam remelting (EBR) was carried out in a K26-15/80 universal electron beam chamber facility (pro-beam systems GmbH), which operated at a beam power ≤ 15 kW and an acceleration voltage ≤ 80 V. Before remelting, the specimens were preheated in-situ up to a temperature of approx. 450 °C using the EB field technique. The preheating procedure was necessary to prevent cracking of the EBR layer. The remelting behaviour of cast iron materials is mainly determined by their graphite morphology. For this reason, the beam guidance technique (multi-spot technique, meander technique, etc.) and/or EB parameters (acceleration voltage, focal position, local and global frequency) were adapted in order to produce crack- and pore-free remelted layers with surface deformation minimized as much as possible. Once the electron beam had passed, the remelted surface layer solidified again almost instantaneously due to rapid heat dissipation into the cold core material – a phenomenon known as self-quenching.

Before nitriding, the surface deformations resulting from the EBR process were machined and ground with 1200grit SiC paper. Gas nitriding was carried out in a chamber furnace at a nitriding temperature of  $T_N = 540$  °C for both different time periods (4 h, 16 h) and nitriding potentials  $K_N = 0.1 / 1 / 4$  [atm<sup>-1/2</sup>].

The influence of the thermal regime during nitriding on the thermal stability of the remelted microstructure was investigated through examination of the pearlitic cast irons. They were annealed under an argon atmosphere at different temperatures and for different time periods, which are associated with both low and high thermal loads typical of conventional nitriding treatments.

The particular annealing conditions were 540 °C/4 h, 540 °C/16 h, 590 °C/12 h, 630 °C/4 h and 670 °C/12 h. The parameter combination of 670 °C/12 h is unusual for conventional nitriding, and may be considered to be an upper limit.

### **Examination methods**

Cross-sectional specimens were prepared for the microstructural characterization. All specimens were embedded in Struers *PolyFast*®, ground and polished (final

step: 1 µm diamond suspension). The nitrided specimens were electrolytically coated with a protective Ni layer prior to embedding. To facilitate light-optical examination, the specimens were etched with 3 % Nital solution. The specimens investigated by scanning electron microscopy were not etched, but subjected to an additional polishing step using Buehler MasterMet™. A LEO 1530 FE-SEM (Carl Zeiss) equipped with an HKL Channel 5 EBSD system (Oxford Instruments) was employed, along with a MIRA 3-SEM (Tescan) equipped with an Apollo XP detector (EDAX/Ametek). Light-optical microscopy was used for measuring the EBR layer thickness ( $t_{FBR}$ ) as well as the compound layer thickness (CLT) after nitriding as an average value of a minimum of 20 individual values. Secondary dendrite arm spacing (SDAS) was determined by means of light-optical micrographs according to the BDG quideline P220 (2011) [11]. This value facilitated the estimation of both the degree of microstructural refinement achieved by the solidification process and the cooling rate. The final SDAS values mentioned were mean values from 10 individual measurements. The treated surfaces were characterised by Vickers hardness-depth profiles perpendicular to the surface using nominal loads of 2.94 N (HV0.3) for EBR layers and 0.49 N (HV0.05) for nidrided layers. For EBR layers, the minimum and maximum hardness values were estimated from the profiles measured.

#### **Results and discussion**

# Microstructure and properties of white-solidified EB remelted surface layers

Independently of the EBR parameters used, all cast irons solidified in a metastable manner according to the Fe-Fe<sub>3</sub>C system. While the base materials investigated were hypoeutectic cast alloys, the main fractions of the so-called white-solidified microstructure were primary austenite dendrites (which decomposed to pearlite) embedded in a eutectic matrix (Fig. 1). The eutectic matrix exhibited typical features of ledeburite, i.e., cementite plates and a rod-like eutectic. While with an increasing cooling rate (equal to SDAS↓, Fig. 1a, c), the volume fraction of rod eutectic (austenite rods in a cementite matrix) decreased, the volume fraction of plate-shaped eutectic microstructure increased. Moreover, analysis of the element distribution using energy dispersive X-ray spectroscopy (EDS) showed a strong heterogeneous silicon (Si) distribution in the ledeburitic microstructure. The eutectic cementite could be assumed to be Sifree. Local Si enrichments occurred along the boundary between eutectic cementite and pearlite. Si microsegregations during solidification and the formation of another ternary Fe-C-Si phase [12,13] – which typically outlines the edges of eutectic cementite plates – obviously led to the Si enrichments observed.



**Fig.1** - White-solidified microstructure of remelted surface layers with small and large SDAS of pearlitic P/S\_2.3Si (a+b) and ferritic F/V\_2.5Si (c+d)

The cooling rate was influenced by the energy input (beam current). As expected, the lower the energy input, the thinner the EBR layer thickness ( $t_{EBR}$ ) and the more compact the secondary dendrite arm spacing (SDAS) (Fig. 2). As a result of the preheating, the cooling rate calculated from the SDAS was  $\leq$  900 K/s [14]. Independent of the beam guidance techniques or EBR parameters used, only marginal differences in layer hardness were apparent as a function of EBR layer thickness or SDAS (cf. Fig. 2a, b). The hardness values measured in the remelted surface layers were in the range of 600 - 750 HV0.3. Compared to the initial hardness of the pearlitic (258 - 288 HV0.3) and ferritic (167 HV0.3) base materials, the hardness values in the EBR layers were more than doubled (Fig. 2a) or quadrupled (Fig. 2b) after EBR.



Fig.2 - Influence of the EBR layer thickness t<sub>EBR</sub> on minimum (hollow rhombus ◊) and maximum (solid rhombus ♦) layer hardness and mean secondary dendrite arm distance SDAS (solid circles ●)

During EB remelting, the spatially connected network of fine graphite lamellae of the GJL cast iron was almost completely dissolved, while the graphite morphologies of the GJS or GJV cast irons – which were present in a

more widely separated manner in the microstructure – floated to the surface in the melt, and were thus not completely melted due to their very short exposure time to the electron beam. This graphite was removed during the surface grinding that was necessary before the remelted surfaces could be nitrided. However, such a procedure can lead to a change in the local carbon content, especially in thin EBR layers. This lower carbon content led to a higher volume fraction of primary austenite (pearlite after cooling) in the remelted microstructure and, thus, to somewhat lower layer hardness values (cf. Fig. 2).

For further studies regarding low-temperature annealing and nitriding, EBR layers were selected for each cast iron with:

low SDAS↓: 3.0 μm - 3.7 μm (t<sub>EBR</sub> = 0.1 mm - 0.3 mm) and
 high SDAS↑: 5.4 m - 6.7 μm (t<sub>EBR</sub> = 1.0 mm - 1.3 mm).

Due to its poor melting behaviour (pore formation, surface deformation), only EBR layers with SDAS↑ (4.9 - 6.7 µm) could be provided for P/L\_2.0Si.

# Influence of thermal regime of nitriding on the microstructure of EBR layers

Starting from the initial levels of hardness after EB remelting (hollow symbols in Fig. 3), the hardness decreased continuously with increasing thermal load.



**Fig.3** - Influence of the thermal regime during annealing on the hardness of EBR layers of pearlitic cast iron. The massive loss of hardness at 670 °C/12 h was due to graphitizing

SEM investigations of the microstructures annealed up to 630 °C/4 h revealed that the changes in hardness (Fig. 3) were mostly related to changes in the microstructure of pearlite.

The lamellar morphology of the pearlite (Fig. 4a) was lost due to the coarsening of the pearlitic cementite (Fig. 4b). The morphology of eutectic cementite (EC) in the EBR layers was hardly affected by annealing up to a thermal load of 630 °C/4 h.

As determined in [15], the presence of Si in the pearlitic **a**-iron (Fe) and along the boundary between eutectic cementite and pearlite significantly retarded the coarsening of the pearlitic cementite and stabilized its lamellar morphology.

Higher silicon content hindered the coarsening of pearlitic cementite more effectively, as indicated by the different changes in hardness of P/S\_2.3Si and P/L\_2.0Si (Fig. 3). The decrease in hardness was larger for P/L\_2.0Si than for P/S\_2.3Si up to 630 °C/4 h.

The higher chromium (Cr) and manganese (Mn) contents in P/L\_2.0Si (cf. Tab. 1) - which are known to dissolve in cementite and, consequently, stabilize pearlitic cementite against coarsening during annealing [16,17] - could not compensate for the lower Si content. At a thermal load of 630 °C/4h, the pearlitic cementite only became coarser, but was largely retained (Fig. 4b). a) Annealed at 540 °C/16 h







Fig.4 - SEM images (backscattered electron contrast) of the annealed microstructure of remelted P/L\_2.0Si. The microstructural constituents, i.e., eutectic cementite (EC), pearlite (P) and ferrite (F), are labelled. Cementite appears dark in contrast, while ferrite appears bright

At the maximum thermal load of 670 °C/12 h, the hardness dropped remarkably due to graphitizing. The EBR layer of P/S\_2.3Si was almost completely graphitized at 670 °C/12 h (Fig. 5b), whereas the remelted P/L\_2.0Si was only partially graphitized under the same conditions. The microstructure of the remelted P/L\_2.0Si contained coarse temper graphite as well as coarse eutectic cementite plates (Fig. 5c). In contrast to the P/S\_2.3Si, the higher thermal stability of the eutectic cementite in P/L\_2.0Si should have resulted from the lower Si content and the higher Cr and Mn content (cf. Tab. 1).

Refining the microstructure (SDAS<sub>↓</sub>) by changing the EBR parameters increased the amount of possible graphite nucleation sites, while the redistribution of Fe and Si was accelerated by more rapid interface diffusion. For this reason, graphitization and ferrite grain growth were more advanced in the smaller remelted microstructures, i.e., the temper graphite particles and the ferrite grains were larger in the smaller remelted layers than in the remelted layers with larger SDAS after 670 °C/12 h (Fig. 5a,b).



Fig.2 - Light-optical micrographs of remelted P/S\_2.3Si (a+b) and P/L\_2.0Si (c) annealed at 670 °C for 12 h, where temper graphite formation occurred

In summary, the low-temperature annealing of the cast  $re T_N \leq 540$  °C and a nitriding time  $t_N \leq 16$  h without chanirons investigated showed that the EBR layers with small ges to their microstructures or hardness. SDAS ( $\geq$  3 µm) could be nitrided at a nitriding temperatu-

# Microstructure and properties of grey- and white-solidified cast irons after nitriding

At the chosen nitriding temperature of 540 °C, the nitriding potential  $K_N$  was selected according to the Lehrer diagram [18] (i.e., the phase diagram for the formation of iron nitrides, which is valid for pure iron) in such a manner that nitriding was performed in the  $\alpha$ -Fe field ( $K_N = 0.1$ , without compound layer), in the  $\gamma'$  field ( $K_N = 1$ ) and in the  $\epsilon$  field ( $K_N = 4$ ) for both short ( $t_N = 4$  h) and long nitriding times ( $t_N = 16$  h).

As expected, no compound layer was formed after nitriding using  $\rm K_{_N}$  = 0.1 in the pearlitic and ferritic cast irons

investigated.

With the exception of P/L\_2.0Si, the ledeburitic surface layers were decarburized considerably during nitriding without compound layer (Fig. 6a), such that after a process time of 16 h, in particular, the first hardness values HV0.05 measured at a distance of 10 µm from the surface had a significantly lower value than the remaining remelted and nitrided surface layer (Fig. 6c).

In the P/L\_2.0Si additionally alloyed with Mn, Cr and Cu (Tab. 1), the eutectic cementite was not decarburized as much and no decrease in hardness was measured directly below the surface (Fig. 6b,c).



Fig.2 - SEM images (backscattered electron contrast) of remelted F/V\_2.5Si (a) and P/L\_2.0Si (b) with the corresponding hardness-depth profiles after gas nitriding at 540 °C/16 h, K<sub>N</sub> = 0.1. While the near-surface pearlitic and eutectic cementite of the F/V\_2.5Si was decarburized, the eutectic cementite of the P/L\_2.0Si was still largely intact

The phase constituents of the compound layers of the EBR layers, which were formed during nitriding with K<sub>N</sub> = 1 and 4 [atm<sup>-1/2</sup>] for 16 h, are summarized in Tab. 2. In

simplified terms, Fe carbonitrides and Fe nitrides are not distinguished conceptually in the following, but are referred to as Fe nitrides ( $\gamma'$ ,  $\epsilon$ ).

Tab.2 - Composition of the compound layer of the microstructural constituents (eutectic cementite; pearlite) in theremelted pearlitic and ferritic cast irons after nitriding at 540 °C for 16 h

K <sub>N</sub> [atm <sup>-1/2</sup> ]			1	4			
Identification	Alloy	Cementite	Pearlite	Cementite	Pearlite		
P/L_2.0Si	GJL-250 <sub>mod</sub>	γ'/ε	Ύ	-	-		
P/S_2.3Si	GJS-600-3	γ'/ε	Ύ	3	(ε)/ε+γ'		
F/S_2.3Si	GJS-400-15	γ'/ε	(γ')/ε	-	-		
F/V_2.5Si	GJV-300	γ'/ε	(γ')/ε	3	3		

During nitriding in the  $\gamma'$  field of the Lehrer diagram (K<sub>N</sub> = 1), all compound layers formed on the remelted and nitrided cast irons were characterized by a strong lateral inhomogeneity in their phase constitution. As shown in Fig. 7,  $\epsilon/\gamma'$  double layers were locally present in nitrided eutectic cementite, which was identified by EBSD phase maps. The nitrided pearlitic regions of the pearlitic cast irons alloyed with Mn and Cu were transformed to  $\gamma'$ . For the ferritic cast irons with low Mn contents, both  $\gamma'$  nitrides and  $\epsilon$  nitrides were detected in the nitrided pearlitic regions.

For  $K_N = 4$ , only the ferritic cast iron F/V\_2.5Si with the high Si content and the low Mn content had macroscopically homogeneous compound layers. For the cast irons alloyed with Mn and Cu, the nitrided pearlite was transformed into  $\varepsilon$  and  $\gamma'$  nitrides at  $K_N = 4$ . Furthermore, above the nitrided pearlite, outwardly directed Fe-nitride growth occurred at the surface (see arrows in Fig. 7). These surface nitrides may have been associated with the formation of pores that appeared near the surface along the eutectic cementite plates and along former  $\alpha/Fe_3C($ ) phase boundaries in pearlite.



Fig.7 - Influence of the Si content and the nitriding potential KN on the phase constitution (blue: α; yellow: θ (Fe<sub>3</sub>C); violet: ε; green: γ') in the compound layer of the EBR layer, measured with EBSD (black: non-indexable areas especially in nitrided pearlite due to pore formation and very fine microstructures)



**Fig.8** - SEM images of remelted F/V\_2.5Si after gas nitriding at 540 °C/16 h, K<sub>N</sub> = 1 with Si nitride precipitations in the compound layer (CL) and diffusion layer (DL): (a) cross section, (b) transition CL – DL and (c) end of DL

After all nitriding treatments and independently of K<sub>N</sub>, nanodimensional precipitations existed in the compound layer (CL) and the diffusion layer (DL) (Fig. 8). These precipitations were amorphous and only occurred where Si enrichments had been detected, i.e., in particular along the α/Fe<sub>3</sub>C (θ) phase boundaries and in α-Fe matrix. The exact chemical composition of the precipitations has not yet been defined. There is a high probability that the precipitates were Si nitrides (Si<sub>3</sub>N<sub>4</sub>) as described by 19 - 21 for binary Fe-Si alloys.



**Fig.9** - Influence of Si content, nitriding potential K<sub>N</sub> and nitriding time t<sub>N</sub> on the mean compound layer thickness (CLT) of nitrided grey-solidified (a) and white-solidified (b) cast irons

Nitriding the grey-solidified cast irons with  $K_N \ge 1$  led to the formation of a compound layer, which was inhomogeneous with respect to the layer thickness.

No compound layer was formed on the graphite due to the absence of iron. Since the nitrogen could penetrate particularly rapidly to greater depths along the small gap between the graphite and the matrix, the local compound layer thickness (CLT) varied greatly and depended on the size of the graphite particles located near the surface (Fig. 9a).

The absence of graphite in the white-solidified remelted surface layers resulted in dense compound layers with a homogeneous compound layer thickness. Both the dendrite arm spacing of the EBR microstructures (SDAS↓ ~ 3.5 µm; SDAS↑ ~ 6 µm) and the Si content (2.0 - 2.5% Si) of the various cast iron materials had only a marginal influence on the compound layer thickness (Fig. 9b). Compared to all other cast irons, P/L-2.0Si contained the largest amount of other nitride-forming alloying elements, such as Mn, Cr and Ti (cf. Tab. 1). This resulted in the lowest compound layer thicknesses of all of the nitriding variants. The mean compound layer thicknesses of the EBR layers for K<sub>N</sub> = 1 were greater than those of the grey-solidified surfaces, while for K<sub>N</sub> = 4, they were lower than the mean compound layer thicknesses of the grey- solidified surfaces - independent of

the nitriding time (Fig. 9).







**Fig.11** -Influence of Si content, nitriding potential  $K_N$  and nitriding time  $t_N$  (4h: hollow symbol; 16h: solid symbol) on the increase in hardness  $\Delta$ HV0.05 ( $\Delta$ HV = HV<sub>DL</sub> – HV<sub>EBU</sub> or HV<sub>BM</sub>) of nitrided cast irons in the initial (green symbols) and remelted states

The diffusion layer thicknesses (DLT) were determined by means of investigating the depths at which the nanodimensioned precipitations were visible under the SEM. The DLT increased with increasing nitriding time and/or nitriding potential up to  $K_N = 1$  (Fig. 10). With a further increase in nitriding potential up to  $K_{N}$  = 4, the diffusion layer thickness remained almost constant (Fig. 10). Within the diffusion layer, the inclusion of nitrogen and the precipitation of special nitrides (e.g. Si, Cr, Ti nitrides etc., cf. [22]) resulted in an increase in hardness, which is shown in Fig. 11 as the difference between the initial hardness of the EBR layer (HV  $_{\rm FBR}$ ) or the base material (HV  $_{\mbox{\tiny BM}})$  and the maximum hardness of the diffusion layer (HV  $_{\scriptscriptstyle \rm DL}$ ). The greatest increase in hardness due to nitriding was measured by increasing the nitriding potential in the range of K<sub>N</sub> = 0.1...1. Nitriding at a nitriding potential of  $K_N = 4$  did not lead to any significant increase in hardness. An analogous relationship resulted for the nitrided pearlitic cast irons (Fig. 11a,b). In the case of nitri-

ded ferritic cast irons, the increase in nitriding potential to K<sub>N</sub> = 4 led to a further increase in hardness (Fig. 11c,d).

#### Conclusion

The influence of gas nitriding conditions by means of the variation of the time and nitriding potential on the formation of nitriding layers on white-solidified (EBR) and pearlitic and ferritic cast irons was investigated.

The white-solidified surface layers were generated by electron beam remelting. By means of the targeted variation of the EB process parameters, and in particular the energy input (beam current), crack- and pore-free remelted surface layers with a layer thickness of  $\leq$  1.3 mm could be produced. Since the pearlitic and ferritic cast irons investigated were hypoeutectic alloys, the white-solidified remelted layers consisted of primary austenite dendrites (decomposed in pearlite) and a eutectic matrix. The thinner the remelted surface layers, the smaller the secondary dendrite arm spacing (SDAS) measured. Regardless of the base material of the cast iron (pearlitic or ferritic), the EBR layer hardness was 600 - 750 HV0.3. For successful implementation of the combination treatment, the thermal stability of the EBR layer during subsequent nitriding was of crucial importance.

Therefore, annealing experiments were carried out under typical nitriding conditions. It was shown that Si stabilized the pearlitic and eutectic cementite against coarsening upon low-temperature annealing. The graphitizing of white-solidified surface layers at temperatures above 630°C, however, was promoted by increased Si content and an increased degree of microstructural refinement. The annealing tests revealed that nitriding treatment at a temperature of  $T_N \le 540$  °C for a nitriding time of tN  $\le 16$  h did not lead to any changes in either the microstructure or the hardness of the EBR layer.

During nitriding, nanodimensional precipitations were formed in both the compound layer and the diffusion layer, which led to an additional increase in hardness. These precipitations were detected in Si-enriched mi-crostructural areas, i.e., in particular along the a/Fe3C ( $\theta$ ) phase boundaries and in the a-Fe matrix. From this information, it was deduced that these were silicon nitrides. Due to their amorphous nature, they could not be detected by X-Ray diffraction (XRD). No Si-nitride precipitations occurred in Si-free eutectic cementite.

In compound layer-free nitriding ( $K_N = 0.1$ ), the EBR layer was decarburized to a considerable extent, with the exception of the P/L\_2.0Si alloyed with Mn, Cu and Cr. The compound layers of the investigated cast irons formed during nitriding with a nitriding potential of KN = 1 exhibited a strong lateral inhomogeneity in their phase compositions. This laterally inhomogeneous constitution of the compound layers was also ob-served for  $K_N = 4$ , with the exception of F/V\_2.5Si. With its high Si content and low Mn content, it was only this F/V\_2.5Si ferritic cast iron that formed macroscopically homogeneous compound layers at  $K_N = 4$ .

The microstructural refinement of the EBR layers (SDAS↓ ~ 3.5 µm, SDAS↑ ~ 6 µm) and the Si content (2.0 - 2.5 wt.% Si) of the cast irons investigated had only a marginal influence on the compound layer thickness, the diffusion layer thickness and the increase in hardness within the diffusion layer.

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