Solution strengthened ferritic ductile iron ISO 1083/JS/500-10 provides superior consistent properties in hydraulic rotators

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Abstract: Consistent mechanical and machining properties are essential in many applications where ductile irons offer the most cost-effective way to produce structural parts. In the production of hydraulic rotators, dimensional tolerances are typically 20 µm to obtain designated performance.

For castings where intermediate strength and ductility is required, it is common knowledge that conventional ferritic-pearlitic ductile irons such as ISO 1083/500-7 show large hardness variations. These are mainly caused by the notoriously varying pearlite content, both at different locations within a part and between parts in the same or different batches. Cooling rate variations due to different wall thickness and position in the molding box, as well as varying amounts of pearlite-stabilizing elements, all contribute to detrimental hardness variations.

The obvious remedy is to avoid pearlite formation, and instead obtain the necessary mechanical properties by solution strengthening of the ferritic matrix by increasing silicon content to 3.7wt% –3.8wt%. The Swedish development in this field 1998 resulted in a national standardization as SS 140725, followed in 2004 by ISO 1083/JS/500-10.

Indexator AB decided 2005 to specify JS/500-10 for all new ductile iron parts and to convert all existing parts. Improvements include reduction by 75% in hardness variations and increase by 30% in cutting tool life, combined with consistently better mechanical properties.

Key words: ferritic ductile irons; ferritic-pearlitic ductile irons; properties; hydraulic rotators; solution strengthening

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1 Introduction

The inventions by Morrogh and by Millis in the end of the 1940s has resulted in an undisputable progress in the development of cost-effective means to produce structural parts of more or less complicated geometries, by casting of ductile iron parts followed by machining to final dimensions and desired surface finish. The ductile iron family of this “first generation” (developed during the second half of the 20th century) actually offers a metallic matrix composite, having various steel matrices ranging from a fully ferritic matrix (0.02 wt% C) to a fully pearlitic matrix (0.8 wt% C). The nodular shape of the graphite dispersoids enables the mechanical properties to be governed mainly by the matrix. This is in contrast to the case for grey iron where the lamellar graphite flakes act as more severe stress concentrators, leading to negligible ductility almost regardless of the intrinsic properties of their ferritic or pearlitic matrices.

In both cases, a silicon level of about 2wt%–3wt% promotes graphite precipitation during grey solidification according to the stable phase diagram for the ternary Fe-C-Si system, instead of metastable (white) solidification with cementite (Fe₃C) precipitation. Mechanical properties of ductile irons can thereby be varied from lower strength levels combined with higher ductility, given by the softer ferritic matrix, through intermediate ferritic-pearlitic combinations up to higher strength levels but with lower ductility, governed by the harder pearlite.

2 Shortcoming of ferritic-pearlitic grades

The main drawback for all of the 1st generation of ductile irons, and probably most pronounced in the most popular ferritic-pearlitic grade ISO 1083/500-7 that combines an intermediate strength with an intermediate ductility, is that the ferrite-pearlite proportions are very sensitive to the local cooling rate of the initially austenitic matrix, and also to variations in the amounts of pearlite-stabilizing elements such as manganese and copper. The first austenitic areas to be transformed into ferrite during cooling are usually located close to the graphite nodules, since the carbon solubility...
is much lower in ferrite than that in austenite, forcing the carbon atoms to diffuse through the growing ferrite into the nodules, at a rapidly decreasing rate due to the increasing distance. However, as soon as the first pearlite has nucleated, it grows much faster due to the much shorter diffusion distance between the pearlitic constituents ferrite and cementite, and therefore pearlite dominates the remaining transformation. The result is often a “bulls-eye” microstructure, where black graphite nodules are surrounded by a ring of white ferrite in a continuous matrix of grey pearlite.

The most obvious consequence of the detrimental sensitivity for variations in cooling rate and composition is that the hardness range can easily vary by at least 30–40 Brinell hardness units, with a corresponding variation in strength and ductility. This is the case both at different locations within a cast component due to various thicknesses of both the metal itself and the sand form surrounding it, but also at the same location in components of the same batch, and in an even more pronounced way between different batches.

The microstructures and range of mechanical properties within the 1st generation of ferritic-pearlitic ductile irons as standardized in ISO 1083:2004 are shown in Fig. 1.

![Microstructures and mechanical properties for the 1st generation of ferritic-pearlitic ductile irons](image)

The hardness interval allowed for ISO 1083/500-7 is 170-230 HBW [2], which actually corresponds to a decrease by 50% in machinability. This causes both reduced machining parameter settings and increased wear of the cutting tools, but also a direct reduction of the cutting depth by about 10 µm in the harder (230 HBW) material compared to the softer (170 HBW). This makes it almost impossible to optimize the machining parameters, or to consistently produce parts of this 1st generation ductile iron by cost-effective machining to narrow dimensional tolerances in the 20 µm range, since hardness variations may then consume the whole allowance, with nothing left for tool wear, backlash, temperature variations, etc.

Further, there is no cost-effective method available to reduce the hardness variations by post heat treatment. Some foundries with limited cooling space are deliberately alloying with more copper producing an almost pearlitic microstructure, followed by a ferritizing heat treatment. This may, however, rather increase the hardness variation between parts of the same batch, since the outer parts in a stack will cool faster (especially the ones in corners), while the inner parts will cool slower.

This restricting situation seems to have been accepted by most foundrymen and customers as a “law of nature”. To summarize, the intrinsic hardness variations in ferritic-pearlitic ductile irons make it very difficult to realize Lean Production in machining operations.

### 3 Alternatives to strengthening by pearlite

Fortunately some researchers [3-6] have sought alternatives to conventional structural strengthening by pearlite.

The most promising alternative mechanism to obtain ductile irons with intermediate strength and ductility is by solid solution strengthening. Among the different elements providing solution strengthening in ferritic irons, the most obvious choice is a further increase of the Si content in the fundamental ternary Fe-C-Si system, in order to raise the tensile strength to 500 MPa from the level 350 – 400 MPa that is obtained in fully ferritic ductile irons at a conventional silicon level of 2.3wt%–2.7wt% Si.

In the Fig. 1 above, **bold green** characters indicate property levels desired for an optimized ductile iron with a tensile strength level of 500 MPa, while undesired properties (in all three grades) are written in **red italics**. Hardness variations in all three and the very low $A_r$ value are mainly related to pearlite. Contrary to this, all of the favorable property levels can be concurrently attained by using ISO 1083/JS/500-10, a fully ferritic ductile iron solid solution strengthened by 3.7 wt%–3.8 wt% silicon!

### 3.1 Si-solution strengthening: misconceptions

Unfortunately this path has been considered doubtful due to two (2) prevailing misconceptions regarding the silicon influence on brittleness and chunky graphite.

The first and major misconception seems to have its origin back in the first ductile iron US Patent by Millis et al in 1949 [1], where it was stated that “… increasing the silicon content over these amounts (>2.5%) apparently lowers the mechanical properties, especially toughness, tensile strength and/or ductility …”. This
has often been summarized as “silicon makes the ductile iron brittle”.

However, all iron alloys containing ≥2.5 wt% Si in their Table V-VI (6 out of 54 alloys) concurrently contained ≥0.8 wt% Mn! This makes the conclusion about Si doubtful, especially since it was also stated that “It is more preferred that the manganese content not exceed about 0.3%, particularly when good ductility and/or high impact properties are desired”. These high manganese levels (≥0.8 wt%) were probably also responsible for the low ADI ductilities obtained in the pioneering austempering trials described in Table XIV of the same patent.

There is no doubt that solution strengthening by silicon has negative effects on increasing the notch-impact transition temperature and reducing the impact energy of ferrite, according to a thorough study by Leslie [1].

This fact has commonly been presumed to represent a serious limitation, obviously without considering that the conventional path to reach higher strengths, namely to have a substantial amount of harder but brittler pearlite in the matrix, also must reduce the notch-impact properties! For ductile irons of the same tensile strength (500 MPa), it was found [2] that matrices of solution strengthened ferrite and of conventional ferrite-pearlite show similar Charpy behavior and energy levels (while ferritic irons with lower Si and strength show higher impact energies).

Furthermore, the significance of notch ed impact energy testing for load-bearing structures has been questioned in a thorough study by Babu et al [3], because the elastic strain rate under plane stress conditions during CVN testing is ≥540 s⁻¹, being at least 5000 times higher than the strain rates experienced in service under plane strain conditions in earth-moving equipment. They conclude that if impact energy at very high strain rates would be a significant property for load-bearing structures, microalloyed steel forgings could not be commonly used as a cost-effective alternative to quenched & tempered steel. This limited relevance may also be the case in most ductile iron components, especially in larger cross-sections.

Fracture toughness determination by JIC and by instrumented Charpy testing indicates that Si-solution strengthened ferritic ductile iron is indeed slightly tougher than ferritic-pearlitic ductile iron of the same tensile strength [4,5]. Finally, the ductility (fracture elongation A₅) at the same strength is considerably improved, see Figs. 5–8.

The second misconception concerns the increased risk of formation of chunky graphite in the interior of larger cross-sections at higher silicon levels. The chunky graphite shape (instead of nodular) may reduce fracture elongation and fracture toughness by up to 50% and reduce tensile and fatigue strength by 20%–25%, while yield strength and hardness are hardly affected at all.

However, thicker castings are usually loaded in bending, leading to lower stress levels in the interior where chunky graphite may form. Further, according to a recent thesis [6] on the subject, the main reason for the formation of chunky graphite is not the high silicon content as such, but low available oxygen content locally in the melt during solidification. That can also be caused by other strong oxide formers like Ce and other rare earths, by Al & Ca.

To prevent chunky formation, it is recommended to aim for a high nodule count, to use chills (increased silicon content reduces the risk of white solidification!), to avoid large risers (partly possible to be substituted by less expensive chills!), and to have balanced Ce + Sb contents.

### 3.2 Si-solution strengthening: real obstacles

During the development and promotion of ferritic Si-solution strengthened ductile irons, four (4) real obstacles have emerged:

The first obstacle is caused by foundry process logistics where very large holding furnaces are fed by smaller smelting furnaces. This makes it complicated to make large adjustments in silicon content between different batches. However, the improvements in compositional control offered by modern larger smelting furnaces makes it possible to manage without a holding furnace.

The second obstacle is the increased need for low-alloyed scrap, especially regarding manganese level but also regarding other pearlite-stabilizing elements. Pig iron may be used in larger amounts, but to keep the CEL close to eutectic when Si level is raised, there may also be a need for hypoeutectic pig iron (and/or sponge iron).

The third obstacle is a common lack of insight that the total production cost for manufacturing by machining of castings is the sum of three (3) categories [7]:

A. Purchase price (model cost + running cost);
B. Production cost (strongly governed by consistent hardness and machinability);
C. Cost of nonconformance (interruptions, rejects).

Production cost at the customer of the castings may often be several times higher than the purchase price, and the cost of nonconformance may also be substantial when the cast material shows inconsistent behavior.

The fourth obstacle is some general conservatism & lack of knowledge at both foundries and their customers.

## 4 Motives for changing iron at Indexator AB

As soon as we became aware of the development work on solution strengthening of ferritic ductile irons and the Swedish standardization 1998 of two new ferritic grades SS 140720 (with 3.2 wt% Si to attain $R_m = 450$ MPa) and SS 140725 (with 3.7 wt% Si to attain $R_m = 500$ MPa), we started to evaluate if the reduction of hardness scatter to ±15 HBW was realistic, and if it could help us to consistently reach our 20 μm dimensional tolerance in turning. It was soon evident that hardness scatter was even lower, usually reduced to about 25% of the previous level!

We were further encouraged by the ISO standardization of the 3.7 wt% Si grade in 2004 as ISO 1083/JS/500-10 (described
in normative Annex A).

Ductile iron castings in 500 MPa grade has always been the major structural material used in rotators designed and manufactured by Indexator for our OEM customers. The improvements offered by the superior and consistent properties of Si-solution strengthened ferritic ductile iron have encouraged us to speed up the conversion. Despite both the prevailing misconceptions and the real obstacles already described, we have now converted about a third of our total annual ductile iron purchase of about 1300 tons / year. To put the number into a global context, Indexator purchases about 1% out of 1% (=1/10 000) of the global annual production of ductile iron.

It is therefore important for Indexator AB to disseminate the positive experiences, in order to further improve the technical and economical advantages of ductile irons, and to promote the supply of superior material grades.

The first new casting design to be manufactured directly from the beginning in solution strengthened JS/500-10, was in 2005 an intricate swivel housing weighing 138 kg, as shown in Fig. 2.

The casting is the main body of an OEM rotator, integrated in a 5-claw recycling grapple for 20-ton loads. Since the introduction, we have successfully produced about 2000 rotators of this model.

Fig. 2: Swivel housing cast 2005 in ISO 1083/JS/500-10, enabling consistent properties and 20 μm tolerances.

(a) Author with machined swivel housing; (b) Recycling grapple in action, with our OEM rotator as upper part;
(c) CAD view showing the five hydraulic seal grooves (yellow) with 20 μm radial tolerance against the shaft.

5 Typical properties in Indexator castings

The hardness level and reduced scatter in our parts, as well as the mechanical properties (yield strength $R_{y0,2}$, tensile strength $R_m$, and fracture elongation $A_f$) offered by the Si-solution strengthened ferritic ductile iron ISO 1083/JS/500-10 are presented in the following Figs. 3-5.

The data is based on one year production of 933 parts, cast at 15 dates at a German foundry, with tensile testing bars

Fig. 3: Hardness for ISO 1083/JS/500-10 vs. Si content from one year production, compared with literature data
Fig. 4: Tensile & yield strengths for ISO 1083/JS/500-10 vs. Si content, from one year production & 1st samples

Fig. 5: Fracture elongation for ISO 1083/JS/500-10 vs. Si content, from one year production & 1st samples

Machined from Y-blocks, and for three of the casting dates supplemented by tensile testing bars machined from castings with wall thickness 55–65 mm.

As can be seen from the data, the silicon content varied in serial production between 3.71 wt%–3.94 wt%, while initial tests with 1st samples varied between 3.63 wt%–3.95 wt% Si. The manganese and carbon contents varied between 0.22 wt%–0.35 wt% and 3.39 wt%–3.57 wt%, respectively.

As expected, the hardness increased slightly with increasing of the solution hardening silicon content, in good agreement with the hardness slope in Björk gren, with an even better agreement in slope and also with the hardness level in Svensson et al [13]. The graph is deliberately drawn with its ordinate reflecting the whole hardness interval 170–230 HBW permitted for ferritic-pearlitic ductile iron ISO 1083/JS/500-7. This visualizes that even with some variation in composition, Si-solution strengthened ferritic ductile irons offer a drastically reduced scatter in hardness, usually being within 5–10 Brinell-units and always within 15 Brinell-units, as long as the silicon content is kept within 0.1 wt%.

We also have experience from scatter reductions around the circumference of cylindrical parts with 250 mm diameter. The hardness typically varied between 179–212 HBW when
cast in ISO 1083/JS/500-7 (due to high sensitivity of cooling rate on ferritic-pearlitic proportions), causing vibrations and dimensional scatter during turning, but the hardness was uniformly 187 HBW when cast in ISO 1083/JS/500-10.

The solution strengthening effect given by silicon on yield strength \( R_{p0.2} \) and tensile strength \( R_m \) in ferritic ductile iron can be seen in Fig. 4. This effect was larger on yield strength than on tensile strength, as can be noted both in comparison to the standardized minimum levels 320 MPa and 500 MPa for ferritic-pearlitic ISO 1083/JS/500-7 at a conventional silicon level of 2.3 wt%–2.7 wt%, but it can also be noted from the steeper slope with Si increase for the yield strength. This raised the \( R_{p0.2}/R_m \) ratio from about 0.6 in ferritic-pearlitic to about 0.8 in ferritic ductile iron solution strengthened by \(-3.8 \) wt% Si.

Both slopes were steeper than in the equations published by Björkegren[12], but his equations were calculated for lower silicon levels between 3.0 wt%–3.4 wt% (corresponding to a tensile strength of 450 MPa). Despite this, the equations resulted in approximate the same yield strength levels, whereas the tensile strength level was about 30–40 MPa higher for the current data.

Some of the strength in solution strengthened ductile iron can be attributed to the fact that in order to keep the CEL close to the eutectic level when the silicon content is raised by +1.2 wt%, the C level has to be decreased by about \(-0.3 \) wt%, which reduces the graphite nodule volume by about 1/10, and thus the volume of the load-carrying “steel” matrix is increased by some percent.

The influence on fracture elongation \( A_t \) from the solution strengthening effect by silicon on the ferritic ductile iron can be seen in Fig. 5. The high mean elongation value 22% obtained in Y-blocks from production was modestly reduced with increasing silicon content \((-0.323\% \text{ for an increase by } 0.1 \text{ wt\% Si})\). This relative reduction was actually smaller than the concurrent relative increase in tensile strength \((+9.29 \text{ MPa for an increase by } 0.1 \text{ wt\% Si, at a mean value of } 540 \text{ MPa})\), and much smaller than the relative increase in yield strength \((+13.8 \text{ MPa for an increase by } 0.1 \text{ wt\% Si at a mean value of } 400 \text{ MPa})\).

The elongation level was in general very high, usually between 20%–25% and never below 15%, as long as the iron had a spherical graphite structure (one exception from this is described below). This means that fracture elongation for Si-solution strengthened ductile iron, despite the concurrently increased yield strength, also reaches about twice the ductility usually obtained by ferritic-pearlitic ductile irons of the same tensile strength!

The effect from variations in manganese content was also compared with the effect of silicon. It was found that for each addition of 0.1 wt%, the increase in yield strength was about half as high for Mn \((+7.4 \text{ MPa})\) as it was for Si \((+13.8 \text{ MPa})\), and tensile strength was almost unaffected with increasing Mn \((-0.8 \text{ MPa})\).

However, the fracture elongation decreased about six times faster for manganese \((-1.83\% \text{ for } +0.1 \text{ wt\% Mn})\) than for an identical increase in silicon \((-0.323\% \text{ for } +0.1 \text{ wt\% Si})\).

This implies that although the low manganese levels may not cause any precipitation of pearlite at these high silicon levels, the segregation of manganese at intercellular areas during the later part of solidification may still have a harmful effect on ductility, although beginning from a very high level. The recommended limit in ISO 1083/JS/500-10, to allow maximum 0.3 wt% Mn (a level actually recommended already by Millis[13]), should therefore be obeyed, and any further reductions down to or even lower than 0.2 wt% Mn may be beneficial for ductility.

The only exception from fracture elongations above 15%, a 10% value obtained in one of the 1st samples (being more typical for ferritic-pearlitic JS/500-7), was found to be due to its mainly vermicular graphite (CGI) structure! This implies that also CGI materials may benefit from the superior mechanical properties offered by Si-solution strengthened ferrite, since this sample was actually very strong with 451 MPa in yield and 565 MPa in tensile strength. The hardness was 207 HBW, with the composition 3.30 wt% C, 3.92 wt% Si and 0.21 wt% Mn.

For applications where high heat conductivity is of prime importance, the graphite structure usually gets the most attention, although contribution from heat conductivity in the continuous “steel” matrix should not be neglected. It was shown by Sjögren et al[14] that “the thermal conductivity of especially compacted graphite iron and ductile iron can be improved considerably by replacing a pearlitic matrix with a ferritic matrix.” The thermal conductivity of ferrite is according to Holmgren et al[15] reduced by about 20% when the silicon content is raised from 2.5 wt% to 3.7 wt% Si, but at the latter silicon level the thermal conductivity 20 W/m-K is still slightly higher than the value for a pearlitic ductile iron with 2.5 wt% Si.

This implies that by replacing conventional pearlitic matrices with Si-solution strengthened ferritic matrices in various types of cast irons, there may be substantial improvements in mechanical and machining properties without sacrificing thermal properties.

### 5.1 Further examples of typical properties

Others have also experienced the advantages of Si-solution strengthened ductile irons. At least three patents have been granted in this field, applied for by Kovacs[16], Kehrer[17], and Menk[18].

In Fig. 6, a comparison of mechanical properties obtained by a Finnish foundry[19] is made between results for 408 Lynchburg tensile testing samples made of conventional ferrit-pearlitic iron JS/500-7 (unfilled symbols) and 116 Lynchburg tensile testing samples made of Si-solution strengthened ferritic ductile iron JS/500-10 (grey filled symbols). It is obvious that ductility was higher for the Si-solution strengthened ferritic ductile iron (10%–22% elongation) than for the conventional ferritic-pearlitic iron (6%–18.5%), and further that the former offer consistent strength while the latter does not!

The scatter in ductility for Si-solution strengthened ductile
Fig. 6: Tensile strength $R_m$ and yield strength $R_{p0.2}$ vs. fracture elongation $A_s$ for 408 Lynchburg tensile testing samples of ISO 1083/JS/500-7 (unfilled triangles and squares) and for 116 Lynchburg samples of ISO 1083/JS/500-10 (grey filled triangles and squares) [16].

Iron may be attributed to variations in nodularity, nodule size distribution and porosity, while these factors may in conventional ductile iron be more or less masked by the large variations in ferrite-pearlitic proportions.

In Fig. 7, the fracture elongation $A_s$ is plotted relative to the yield strength $R_{p0.2}$ for conventional ferritic-pearlitic iron (unfilled symbols) and Si-solution strengthened ferritic ductile iron (black filled symbols). The advantage given by the concurrent increase of yield strength and fracture elongation is evident.

Finally, Figure 8 shows the Material Quality Index, defined as $MQI = R_m^2 \times A_s \times 10^4$, plotted against yield strength. This

Fig. 7: Fracture elongation $A_s$ vs. yield strength $R_{p0.2}$ for ISO 1083/JS/500-7 (grey unfilled diamonds) and ISO 1083/JS/500-10 (black filled diamonds) [16].
5.2 Fatigue properties

The fatigue properties have been examined by Hamberg et al. They found that the behavior for Si-solution strengthened ductile iron SS 140725 (ISO 1083/JS/500-10) was at least as good as for ferritic-pearlitic ductile iron.

5.3 Properties at even higher silicon levels

The mechanical properties at higher solution strengthening levels have been examined by Kikkeri. He found that for a silicon content of 4.44 wt% Si, the yield strength $R_{p0.2}$ was 557 MPa, the tensile strength $R_m$ was 677 MPa, and the fracture elongation $A_f$ was 12.2 %. This material could easily replace pearlitic-ferritic ISO 1083/JS/600-3 (with $R_{p0.2} > 370$ MPa), and may with a slightly higher silicon content also be able to replace fully pearlitic ISO 1083/JS/700-2 (with $R_{p0.2} > 480$ MPa).

This should not be a surprise for those familiar with the properties of SiMo alloys, being ferritic ductile iron developed for high-temperature applications like exhaust manifolds and turbocharger housings. In a study by Li et al., the mechanical properties were evaluated for silicon levels between 3.7 wt% ~ 5.2 wt% and molybdenum levels between 0.6 wt% ~ 0.9 wt%. The yield strength $R_{p0.2}$ increased with increasing silicon content within the range 460 ~ 630 MPa, the tensile strength $R_m$ increased within 610 ~ 790 MPa, while the fracture elongation $A_f$ decreased within 16% ~ 7 %. Some of the increase in strength and decrease in ductility in these materials may be attributed to the precipitation of molybdenum carbides. However, it was stated that “elongation of up to 10 % can be steadily achieved for 4.95 wt% Si samples through controlling melt chemistry and inoculation practice.”

Summary

The advantages of fully ferritic ductile iron solutions strengthened by increased Si levels have been described. For an intended ultimate tensile strength of 500 MPa, the ductility (elongation at fracture $A_f$) is about twice as high in the solution strengthened ferritic ductile iron compared to conventional ferritic-pearlitic ductile iron, combined with a concurrent increase of the yield strength, raising the $R_{p0.2}/R_m$ ratio from about 0.6 to 0.8. The impact energy behavior is comparable and the fracture toughness is slightly better than for ferritic-pearlitic iron.

There is ample evidence that ferritic ductile iron solutions strengthened by silicon to various superior combinations of mechanical and machining properties, ought to be entitled to the designation “second generation of ductile iron”.

The approaching “paradigm shift” towards the 2nd generation of ductile irons will, together with the continuing development of ausferritic ductile irons (ADI), further strengthen the cost-effectiveness of ductile irons and facilitate Lean Production.

References


About the author

Male, Swedish citizen, born in Stockholm in 1961. Dr. Larker has a Masters degree in Mechanical Engineering and a Ph. D degree in Engineering Materials from Lulea University of Technology in Sweden.

He was appointed Associate Professor in Engineering Materials at the same University 1998. Since then, Richard is R&D Manager at the Swedish company Indexator AB, with a focus on materials selection, tribology and testing of their hydraulic products, where ductile irons have a major role.

He is chairman of the current Swedish development project “Weight- and Volume-Intelligent Cast Structures” during 2006-2009, with $3.2 M funding shared 50/50 between government and companies like Volvo Truck, Scania, Volvo Construction Engineering, Atlas Copco, Komatsu Forest and Indexator. The project aims at developing cost-effective cast structures in ausferritic (ADI) and solution strengthened ductile irons for trucks, buses, wheel loaders, excavators, cranes and forestry machines.

He is chairman of the Swedish standardization committee TK 130 “Cast iron and steel”. He is nominated expert in the current revision of European standards

- on Ductile iron (EN 1563), on ADI (EN 1564),
- on Technical conditions of delivery (EN 1559), and
- on Designation system for cast iron (EN 1560).

He is also active in the current development within ISO/TC25 on a Technical Report named “Understanding cast iron materials”, aimed to assist designers and engineers in understanding the possibilities and limitations of cast irons.

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